
**The Influence of Inelastic Damage on Creep,
Fatigue and Fracture Toughness**

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Abstract

A significant number of advanced gas cooled reactor (AGR) plant components with operating temperatures in the range of 500-650 °C undergo creep-fatigue loading conditions. Some of these components have previously operated in the creep regime and subsequently had their operating temperatures reduced. The aims of this project are to determine the effects of service exposure, thus prior creep damage, on fatigue crack growth and fracture behaviour of the material and to include these effects in fatigue and fracture assessments.

In laboratory testing, creep damage can be introduced into the material globally, where uniform creep damage will form through a section, or locally, where creep damage is confined to a small region local to the crack tip. The local creep damage (LCD) approach has been employed in this research and the global creep damage (GCD) tests will be performed in future work.

The material examined is an ex-service Type 316H stainless steel, which is widely used in AGR plant components. Creep damage local to the crack tip has been introduced into the material by interrupting creep crack growth (CCG) tests performed at 550 °C, which is the typical operating temperature of the plant components. Fatigue crack growth (FCG) and fracture toughness tests have been performed on these LCD specimens and the results are compared to those of conducted on the material with no creep damage.

The yield stress of Type 316H stainless steel is relatively low at 550 °C. In order to limit the extent of plasticity in CCG tests during loading and unloading, the material has been uniformly pre-compressed (PC) to 8% plastic strain at room temperature prior to specimen manufacturing. It has been shown that the pre-compression process increases the yield stress and reduces the tensile failure strain of the material compared to the as-received (AR) material's tensile response. A substantial drop in the rupture time and creep ductility has also been exhibited in the uniaxial creep behaviour of the PC material compared to AR, whereas the creep strain rate remains almost unchanged.

To interrupt CCG tests on LCD specimens at a desired normalised crack length of around $a/W = 0.5$, tests have been performed on nominally identical specimens to calibrate the CCG behaviour of the material. The results from these tests have been analysed in terms of the creep fracture mechanics parameter C^* and compared to the

available test data on AR material at 550 °C. It has been shown that the creep crack growth rate, da/dt , in PC material is around an order of magnitude larger than those of seen in short term tests on AR material and the data follow the trend of long term tests on AR material.

FCG and fracture toughness tests have been performed on LCD specimens at room temperature to investigate the potential effects of prior creep damage on subsequent fatigue and fracture behaviour of the material, outside the creep regime. The results from these tests have been compared to those of performed on PC material containing no creep damage. A considerable drop in the fracture energy of the material has been observed in LCD specimens in regions close to and away from the creep damage zone, compared to those of obtained from PC material. Assuming a sharp crack tip in LCD specimens, though discontinuous, the FCG rate per cycle, da/dN , data have been correlated with the stress intensity factor range, ΔK . A severe delay in the crack initiation and also a significant drop in the crack growth rate per cycle in the regions close to the creep damage zone have been observed in the FCG data on the LCD specimen compared to the PC material. However, further away from the creep damage zone the FCG data from LCD specimen converge with the trends attained from the PC material.

Further FCG and fracture toughness tests on LCD specimens need to be performed in future work to confirm the observed trends. Also to examine the influence of uniform creep damage in the absence of micro cracks (which form ahead of the crack tip when the local creep damage approach is employed) on the subsequent FCG and fracture behaviour of the material, GCD tests need to be conducted in future work and the results to be compared to those of obtained from LCD specimens.

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Declaration

I hereby declare that the work presented in this dissertation is original and has not been submitted for a degree or diploma at any other university or institution. Information derived from the published and unpublished work of others has been acknowledged in the text and references are given in the list of sources.

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Nomenclature

a	Crack length
a_0	Crack length
a_e	Effective crack length
a_f	Final crack length
Δa	Increment of crack growth
Δa_{max}	Maximum allowable crack extension in a fracture toughness test
$\dot{a}$, da/dt	Creep crack growth rate
$\dot{a}_0$	Initial creep crack growth rate
$\dot{a}_s$	Steady state creep crack growth rate
$\dot{a}_{NSW}$	Creep crack growth rate predicted by NSW model
$\dot{a}_s^{NSWA}$	Creep crack growth rate predicted by NSWA model
$\dot{a}_{NSW-MOD}$	Creep crack growth rate predicted by NSW-MOD model
da / dN	Crack growth rate per cycle in fatigue crack growth tests
A	Creep stress coefficient in minimum creep strain law
A_0	Initial cross sectional area in uniaxial specimen
A_A	Creep stress coefficient in average creep strain law
A_f	Final cross sectional area in uniaxial specimen
A_p	Coefficient in power law hardening expression for plasticity
B	Specimen thickness
B_e	Effective thickness
B_n	Specimen net thickness between side grooves
B_r	Coefficient of rupture law
c	Uncracked ligament of a specimen
C	Primary creep coefficient
C_0	The average of the elastic compliances in fracture toughness tests

C_c	Corrected elastic compliance
$C(t)$	Transient creep characterising parameter
C^*	Steady state creep characterising parameter
C^*_{exp}	Experimentally determined value of C^*
C^*_{ref}	Estimate of C^* from reference stress solution
C^*_{EPRI}	Estimate of C^* from EPRI solution
d	Instantaneous diameter in uniaxial specimen
d_0	Initial diameter in uniaxial specimen
$d_{0,hkl}$	The strain free value of the lattice plane spacing for {hkl} crystallographic plane
d_{hkl}	The lattice plane spacing for {hkl} crystallographic plane
D	Constant coefficient in creep crack growth correlation with C^*
D_f	Fatigue damage
D'	Constant coefficient in creep crack growth correlation with K
E	Young's Modulus
E_{hkl}	Young's Modulus for {hkl} crystallographic plane
E'	Effective Young's modulus
f_i	Original area fraction of voids
F_y	Pre-fatigue cracking force
g	Grain size in Cocks and Ashby model
G	Energy release rate
h	Half of the initial distance between the centres of the pin holes
h_I	Non-dimensional geometry function in EPRI solution for J
H	Non-dimensional function of specimen geometry and N or (n)
I_n	Function of n in RR solution
I_N	Function of N in HRR solution
J	Elastic-plastic fracture mechanics parameter
$J_{0,k}$	Fracture resistance at the k th interval

$J_{0.2}$	A line corresponding to the total crack extension of 0.2 mm
$J_{0.2/BL}$	A line parallel to the blunting line at the offset of 0.2 mm
J_e	Elastic contribution to J
J_{max}	Maximum allowable J in fracture toughness tests
J_p	Plastic contribution to J
J_{ref}	Estimate of J from reference stress solution
J_{EPRI}	Estimate of J from EPRI solution
J_{IC}	Critical value of J for fracture under Mode I loading conditions
ΔJ	Change in the J parameter in cyclic tests with significant plasticity
K	Stress intensity factor
K_c	Fracture toughness for brittle materials
K_{IC}	Material Mode I plane strain fracture toughness
K_{max}	Maximum stress intensity factor in cyclic tests
K_{min}	Minimum stress intensity factor in cyclic tests
ΔK	Stress intensity factor range in cyclic tests
l_0	Initial gauge length in uniaxial specimen
$2l$	Centre to centre void spacing in Cocks and Ashby model
m	Exponent in correlation of creep crack growth rate with K
n	Creep stress exponent in minimum creep strain law
n_p	Creep stress exponent in primary creep strain law
n_A	Creep stress exponent in average creep strain law
N	Ramberg-Osgood power law stress exponent
N_{fi}	Number of cycles to failure in cycles of type i .
N_i	Number of cycles of type i
p	Constant in primary creep strain and strain rate expression
P	Applied load
P_{max}	Maximum load level in cyclic or fracture toughness tests
P_{min}	Minimum load level in cyclic tests

P_{p0}	Geometry dependent normalising load based on σ_{p0}
P_{LC}	Plastic collapse load of cracked body
Q	Activation energy
r	Radial distance from the crack tip
r_c	Creep process zone
r_p	Plastic zone size
r_y	Plastic zone correction factor
$\dot{r}$	Void growth rate
R	Universal gas constant
R -ratio	The ratio of P_{min} to P_{min} in cyclic tests
s	Length of crack tip contour Γ
ds	Element along the crack tip contour Γ
S_{ij}	Deviatoric stress tensor
t	Time
$t_{0.2}, t_{0.5}$	Time for 0.2 mm and 0.5 mm crack extension
t_f	Time to failure under uniaxial stress conditions
t_f^*	Time to rupture under multiaxial stress conditions
t_i	Initiation time
t_n	Void nucleation time in Cocks and Ashby model
t_r	Time to rupture
t_{red}	Redistribution time
t_{red}^{eng}	Engineering definition of the redistribution time
t_T	Transition time from small scale creep to widespread creep conditions
T	Temperature
T_i	Traction vector
T_m	Melting temperature
u_d	The uncertainties in the d-spacing measurements

u_{d0}	The uncertainties in the stress free d-spacing measurements
u_i	Displacement vector
$\dot{u}_i$	Displacement rate vector
U	Potential energy
U_k	Area under the force vs. displacement curve up to the line of constant displacement at the k th interval
V, V_0, V_f	Instantaneous, initial and final potential drop (PD) readings
W	Specimen width
W_s	Strain energy density
$\dot{W}$	Strain energy rate density
x	Coordinate directly ahead of crack tip
y	Coordinate normal to the crack plane
Y	Non-dimensional shape function in expression for K
α	Constant in Ramberg-Osgood plasticity law
β	Constant in plastic zone size expression
δ_{ij}	Kronecker delta
Δ	Load line displacement
Δ_c	Creep load line displacement
Δ_e	Elastic load line displacement
Δ_p	Plastic load line displacement
Δ_{Total}	Total load line displacement
$\dot{\Delta}$	Load line displacement rate
$\dot{\Delta}^c$	Component of displacement rate directly associated with the accumulation of creep strains
$\dot{\Delta}^i$	Component of displacement rate directly associated with instantaneous (elastic and plastic) strains
$\dot{\Delta}_e^i$	Component of displacement rate directly associated with instantaneous elastic strains

$\dot{\Delta}_p^i$	Component of displacement rate directly associated with instantaneous plastic strains
ε	Strain
$\varepsilon_1, \varepsilon_2, \varepsilon_3$	Principal strain components
ε^c	Component of strain due to creep
ε^e	Elastic strain
ε^p	Plastic strain
ε_{crit}	Critical creep strain at failure
ε_f	Uniaxial creep strain at failure (creep ductility)
ε_{p0}	Ramberg-Osgood normalising strain
ε_p^c	Creep strain in the primary region
ε_{ref}	Strain at reference stress
ε_{ref}^c	Creep strain at reference stress
ε_{ref}^e	Elastic strain at reference stress
$\dot{\varepsilon}$	Minimum creep strain rate
$\dot{\varepsilon}_0$	Normalising strain in creep law and RR field
$\dot{\varepsilon}_A$	Average creep strain rate
$\dot{\varepsilon}^c$	Creep strain rate
$\dot{\varepsilon}^e$	Elastic strain rate
$\dot{\varepsilon}^p$	Plastic strain rate
ε_f^*	Multiaxial creep ductility
$\dot{\varepsilon}_p^c$	Creep strain rate in the primary region
$\dot{\varepsilon}_{ref}^c$	Creep strain rate at reference stress
$\tilde{\varepsilon}_{ij}(\theta; n)$	Non-dimensional strain in RR field solution
$\tilde{\varepsilon}_{ij}(\theta; N)$	Non-dimensional strain in HRR field solution
$\bar{\varepsilon}_{ij}^e$	Equivalent elastic strain tensor
$\bar{\varepsilon}^p$	Equivalent plastic strain
Γ	Anticlockwise integration contour around the crack tip

η	Factor relating J or C^* to load and displacement measurements
λ	Non-dimensional crack velocity
μ	Plane stress/strain parameter in expression for J_{ref}
ν	Poisson's ratio
ν_{hkl}	Poisson's ratio for $\{hkl\}$ crystallographic plane
ν_r	Creep rupture stress exponent
ω	Creep damage parameter
$\dot{\omega}$	Creep damage accumulation rate
ϕ	Exponent in correlation of creep crack growth rate with C^*
θ	Crack tip angle
$\theta_{0,hkl}$	The strain free value of Bragg angle for $\{hkl\}$ crystallographic plane
θ_{hkl}	The Bragg angle for $\{hkl\}$ crystallographic plane
σ	Stress
σ_0	Normalising stress in power law creep expression and RR field
$\sigma_{0.2}$, $\sigma_{0.2\%}$	0.2 % proof stress
σ_1 , σ_2 , σ_3	Principal stress components
σ_a	Stress amplitude in cyclic tests
σ_e	Equivalent stress
σ_m	Hydrostatic or mean stress
σ_{max}	Maximum stress in cyclic tests
σ_{mean}	Mean (average) stress in cyclic tests
σ_{min}	Minimum stress in cyclic tests
σ_{nom}	Nominal (engineering) stress
σ_{p0}	Ramberg-Osgood normalising stress
σ_{ref}	Reference stress
σ_{true}	True stress
σ_{UTS}	Ultimate tensile stress
σ_y	Yield stress

σ_{xx}	Stress component in x coordinate direction
σ_{yy}	Stress normal to a crack or notch
σ_{zz}	Stress component in z coordinate direction
σ_m / σ_e	Triaxiality
$\Delta\sigma$	Stress range in cyclic tests
$\tilde{\sigma}_{ij}(\theta; N)$	Non-dimensional stress in HRR field solution
$\tilde{\sigma}_{ij}(\theta; n)$	Non-dimensional stress in RR field solution
τ	Estimate of time when $C(t) \rightarrow C^* (= J/C^*)$ (redistribution time)
τ_{xy}	Shear stress in x - y plane
CCG	Creep crack growth
CCI	Creep crack initiation
CMOD	Crack mouth opening displacement
DSA	Dynamic strain aging
FCG	Fatigue crack growth
FE	Finite element
GCD	Global creep damage
HRR	Hutchinson-Rice-Rosengren elastic-plastic crack tip field solution
LCD	Local creep damage
LLD	Load line displacement
LVDT	Linear variable differential transformer
MSF	Multiaxial strain factor
ND	Neutron diffraction
NSW	Nikbin, Smith and Webster creep crack growth model
NSWA	Approximate NSW model
NSW-MOD	Modified version of the NSW model
PE	Plane strain conditions
PS	Plane stress conditions
ROA	Reduction of area

RR	Riedel and Rice creep crack tip field solution
RS	Residual stress
SIF	Stress intensity factor
SSC	Small scale creep
SSY	Small scale yield
UTS	Ultimate tensile strength

Chapter 1

Introduction

Component fabrication processes and in-service history effects can often lead to changes in a material's mechanical response and deformation behaviour. Typical examples of pre-service fabrication processes are welding, forging, bending and rolling whereas the service exposure effects may be exhibited in the form of creep damage. Type 316H stainless steel (SS) is of great interest, particularly due to its resistance to corrosion and widespread use in components in the UK's advanced gas cooled reactor (AGR) power stations, which operate at temperatures in the range of 500-650 °C. An important issue to be determined in the life assessments of AGR power plants is to investigate fatigue crack growth and fracture behaviour of defects in plant components which have previously operated in the creep regime (typically at temperatures of around 550 °C) and subsequently had their operating temperatures reduced to 470 °C.

Because of economic reasons it is desirable to optimise the components life times in high temperature plants. However, it must be ensured that a service exposed component, which may contain creep damage, can sustain the applied loads, with suitable safety margins against the lower temperature brittle/ductile fracture and fatigue crack growth (FCG) failure mechanisms and the components removal from service occurs prior to failure. A study of the influence of creep damage on FCG and fracture toughness is performed in this research by conducting a series of fatigue and fracture toughness tests on creep damaged material and comparing the results to those of obtained from specimens containing no creep damage.

Due to creep deformation in polycrystalline materials at elevated temperatures, intergranular voids nucleate, grow and coalesce to form micro cracks along the grain boundaries. This process is generally referred to as "creep damage". Creep damage is known to be a life limiting factor in components subjected to high stresses at elevated temperatures. For laboratory testing, the conventional approach to introduce creep damage into a fracture compact tension, C(T), specimen is to perform interrupted creep crack growth (CCG) tests in which creep damage exists local to the crack tip. These interrupted CCG tests are called local creep damage (LCD) tests in this work.

Type 316H SS has a relatively low yield stress at 550 °C. Therefore the development of some plastic strain during the specimen loading and unloading at this

temperature is inevitable. Interpreting the results from interrupted CCG test data obtained from as-received (AR) material will therefore be complicated due to the difficulties to distinguish between the influence of creep and plasticity on the material's fracture behaviour. However, it has been previously shown that the extent of plasticity on loading is almost eliminated when the material is pre-compressed to 8% plastic strain [1, 2]. Therefore, to examine the influence of creep damage alone on the fatigue and fracture behaviour of 316H, the material has been pre-conditioned through uniform pre-compression (PC) to 8% plastic pre-strain at room temperature, thus hardening the material. Another advantage of material pre-compression taken place in this project is to study the pre-straining effects induced due to fabrication processes on the subsequent deformation and fracture behaviour of the material.

In order to understand the pre-compression effects on the fracture behaviour of the material, uniaxial creep, CCG, FCG and fracture toughness tests are performed on the AR and PC materials and comparisons have been made between the results from these tests. FCG and fracture toughness tests have also been repeated on LCD specimens, which are made of the PC material, and the results are compared to those of obtained from AR and PC materials. This way the influence of combined pre-straining (and thus prior plastic damage) and creep damage, which can be referred to as “inelastic damage”, on the fracture behaviour of the material has been examined.

1.1 Aims and Objectives

The overall aim of this research is to determine the influence of prior inelastic damage (creep damage introduced into the pre-strained material) on subsequent fatigue crack growth and fracture toughness of the material through numerical and analytical methods. The main objectives of this work are

1. To investigate pre-compression effects on the tensile response, creep deformation, CCG, FCG and fracture behaviour of 316H SS
2. To examine the influence of pre-compression on the micro structural changes in the material
3. To develop experimental techniques to introduce local creep damage into fracture specimens at a desired crack length
4. To examine the influence of prior creep damage on the fracture toughness and FCG behaviour of the material

1.2 Structure of Thesis

The elastic, plastic and creep fracture mechanics concepts are reviewed in Chapter 2 and the theoretical backgrounds relevant to the creep damage, fatigue deformation and fracture toughness analyses performed in this research are presented. Furthermore, a summary of creep crack initiation and growth prediction models available in the literature are given in Chapter 2. Materials and pre-conditioning details are explained in Chapter 3. Further presented in this chapter is a brief literature review of the pre-compression and creep damage effects on the subsequent fracture behaviour of the material. A proposed method to introduce creep damage into laboratory fracture specimens is explained and the specimen geometries, dimensions and orientations are also given in Chapter 3. Experimental testing and analysis procedures for creep crack growth, fatigue crack growth, fracture toughness and residual stress (RS) measurements applied in this work are introduced in Chapter 4.

The tensile and uniaxial creep test results on the AR and PC materials are described in Chapter 5 and the pre-compression effects on the uniaxial fracture behaviour of the material have been examined. The creep crack initiation and growth test data on the PC material are shown in Chapter 6 and the results obtained are compared to the short/long term data on the AR material and also to those of predicted using NSW creep crack initiation and growth models. A numerical modelling technique to predict the CCG behaviour of AR and PC materials at 550 °C is presented in Chapter 7 and the results are assessed through comparisons with the experimental data previously shown in Chapter 6. The creep properties employed in this model are based on the trends estimated for the AR material in Chapter 5, assuming that the PC material follows the same trends. The results and observations from a micro structural study on the AR and PC materials before and after tensile and creep tests are included in Chapter 8. These results are discussed based on the experimental data presented in Chapter 6. The influence of inelastic damage on fatigue and fracture behaviour of the material is examined through experiments and the results are shown in Chapter 9. The pre-compression effects on the FCG and fracture toughness of the material have been investigated by comparing the results from the tests on the AR and PC materials. Also to examine creep damage effects on the fatigue and fracture behaviour of the material, the experimental results from the tests on LCD specimens are compared to those of obtained from PC material in this chapter. Further included in Chapter 9 is the comparison of the CCG test data on LCD specimens to those of obtained from the CCG calibration tests. Also presented in this chapter are the results from residual

stress measurements on an interrupted CCG specimen with nominally the same dimensions and loading conditions as LCD samples. Furthermore, the dominant cracking modes under creep and fatigue deformation conditions and the interaction between these two failure mechanisms have been studied through fractography on LCD specimens in Chapter 9 and the observations are discussed in terms of the creep damage effects on subsequent fatigue and fracture behaviour of the material. Finally, conclusions and suggested future work are stated in Chapter 10.

Appendices have been included at the end of the thesis. The test matrices detailing the specimen geometries, testing temperatures and material's pre-straining conditions are summarised in Appendix A. Details of the numerical results to evaluate accurate solutions of the η parameter for crack lengths in the range of $0.35 \leq a/W \leq 0.45$ are presented in Appendix B and comparisons have been made to the solutions available in the literature. The sensitivity of the η parameter to the power law exponent, N , and also the stress state has also been shown in this Appendix. The comparisons of the experimental creep crack initiation and growth data on the PC material to the NSW prediction models were previously described in Chapter 6, by employing the average creep strain rate properties and the creep ductility evaluated based on the reduction of area (ROA) in NSW models calculations. More comparisons have been provided in Appendix C by employing the minimum creep strain rate power law constants and the axial creep ductility in NSW models calculations.

Chapter 2

Deformation and Fracture Concepts

2.1 Elastic-Plastic Deformation

The total strain accumulated in a body can be expressed as the sum of time independent strain and time dependent creep, $\varepsilon^c(t)$, components whereas the time-independent component can be divided into elastic ε^e and plastic ε^p parts:

$$\varepsilon_{total} = \varepsilon^e + \varepsilon^p + \varepsilon^c(t) \quad (2.1)$$

This expression is usually known as the “theory of total creep deformation” which in particular describes the behaviour of the material in the steady state creep region [3].

The elastic component of total strain under uniaxial conditions is represented by

$$\varepsilon^e = \sigma / E \quad (2.2)$$

where σ is the applied stress and E is the elastic (Young's) modulus of the material.

The plastic strain may be described using a power law equation as

$$\varepsilon^p = \alpha \varepsilon_{p0} (\sigma / \sigma_{p0})^N = A_p \sigma^N \quad (2.3)$$

where N is the power law plasticity stress exponent, ε_{p0} and σ_{p0} are the normalising strain and stress for plasticity, respectively, and α and A_p are constants which can be

related by $A_p = \frac{\alpha \varepsilon_{p0}}{\sigma_{p0}^N}$.

It may be assumed that under stresses less than yield strength, σ_y , the material behaves linearly while a power law equation relates stress and strain at stresses beyond yield. The linear and non-linear behaviour of material under uniaxial stress can be written as

$$\frac{\varepsilon}{\varepsilon_y} = \frac{\sigma}{\sigma_y} ; \text{ for } \sigma < \sigma_y \quad (2.4)$$

$$\frac{\varepsilon}{\varepsilon_y} = \left(\frac{\sigma}{\sigma_y} \right)^N ; \text{ for } \sigma > \sigma_y \quad (2.5)$$

where $\varepsilon_y = \frac{\sigma_y}{E}$ is the yield strain. Note the yield stress in Eqns (2.4) and (2.5) is often

taken as 0.2% proof stress, $\sigma_{0.2}$, of the material.

2.1.1 Ramberg-Osgood Model

The tensile behaviour of a strain hardening material may be described using Ramberg-Osgood material model by [4]

$$\varepsilon = \frac{\sigma}{E} + A_p \sigma^N \quad (2.6)$$

in uniaxial form and

$$\frac{\varepsilon}{\varepsilon_{p0}} = \frac{\sigma}{\sigma_{p0}} + \alpha \left(\frac{\sigma}{\sigma_{p0}} \right)^N \quad (2.7)$$

in non-dimensional form. In Eqns (2.6) and (2.7), the first term on the RHS represents linear elastic deformation and the second term is to describe non-linear plastic deformation of the material.

The normalising stress, σ_{p0} , in Ramberg-Osgood model, Eqn (2.7), is often taken as 0.2% proof stress of the material, $\sigma_{0.2}$. Ramberg-Osgood material model defines non-linearity to exist from the beginning of the tensile response which continuously increases by increasing the stress. Thus, no absolute yield point is considered in this material model. However, under stresses much smaller than 0.2% proof stress, the non-linear term in equation (2.6) is found almost negligible and a linear relationship between stress and strain can be inferred. On the other hand as the stress increases, the linear component of the Ramberg-Osgood equation becomes negligible and the power law term is magnified.

2.1.2 Dynamic Strain Aging Effects

The concept of dynamic strain aging (DSA), which is equivalent to Portevin-Le Chatelier (PLC) effect, was first introduced in [5] and further examined in [6-8]. This mechanism can be described as follows:

The discontinuous yielding behaviour of the material which is referred to as DSA, occurs due to rapid diffusion of the solute atoms in the specimen at a rate faster than the speed of the dislocations. Therefore the dislocations are caught and locked by the solute atoms. A load increase is thus needed to take apart dislocations from solute atoms and once the dislocations are torn away from the solute atoms, a sudden load drop occurs. The amount of stress required to unpin the solute atoms from the dislocations increase with increasing the time that dislocations are arrested. When the dislocations are released, they start moving at a large speed until they are arrested by the solute atoms again. The process of pinning and unpinning is repeated continuously and leads to the serrations in the stress-strain curve. It has been noted that Nitrogen

plays an important role in the strain aging of steels compared to Carbon due to higher solubility and diffusion coefficient [9].

During the DSA process, the competition between the slow increase of the stress levels and sudden unpinning of the dislocations, leads to a negative strain rate in a macroscopic level [10]. It has been suggested in [10] that the appearance of PLC at high temperatures can be attributed to bulk diffusion. DSA is observed and reported in a variety of f.c.c (*e.g.* [11, 12]) and b.c.c materials (*e.g.* [13]) and constitutive models are developed to describe DSA effects which generally depend on the strain rate, grain size and temperature (an example is given in [13]). Having known that the crystal structure of 316H austenitic stainless steel is f.c.c, DSA effects are expected to be observed in the tensile behaviour of this material at elevated temperatures as shown and described in [14, 15].

2.1.3 Multiaxial Deformation

The Ramberg-Osgood equations (2.6) and (2.7) describe the tensile behaviour of a power law hardening material under uniaxial stress conditions. However under multiaxial stress states a yielding criterion is required to specify when the material starts to deform plastically. The common yielding criteria in engineering applications are von-Mises and Tresca. According to Tresca yield criterion, the yielding of a material begins when the maximum shear stress attains a critical value. von-Mises yield criterion suggests that the yield occurs when the shear strain energy reaches a critical value. In high temperature fracture mechanics the von-Mises criterion is often applied.

2.1.3.1 Hydrostatic (Mean) Stress

The hydrostatic stress (also known as the mean stress) is the uniform stress which is equal in all directions. Experimental observations prove that only deviatoric (shear) stresses control the yielding process and the mean stress has no contribution in deformation and yielding of a material. Having known the principal stresses ($\sigma_1 > \sigma_2 > \sigma_3$) for a given material and loading conditions, the mean stress can be calculated by

$$\sigma_m = \frac{\sigma_1 + \sigma_2 + \sigma_3}{3} \quad (2.8)$$

2.1.3.2 von-Mises Equivalent Stress

The critical stress level at which yielding occurs under multiaxial stress conditions is known as the equivalent stress, σ_e (or $\bar{\sigma}$). According to the von-Mises yield criterion, the equivalent stress can be calculated from

$$\sigma_e = \frac{1}{\sqrt{2}} \sqrt{[(\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_1 - \sigma_3)^2]} \quad (2.9)$$

where σ_1 , σ_2 and σ_3 are components of principal stress. It can be seen in Eqn (2.9) that under uniaxial stress conditions (*i.e.* $\sigma_2 = \sigma_3 = 0$) the equivalent stress is equal to the principal stress ($\sigma_e = \sigma_1$).

2.1.3.3 Equivalent Strain

Under multiaxial stress conditions, the equivalent elastic strain can be obtained by

$$\bar{\varepsilon}_{ij}^e = \frac{(1+\nu)}{E} S_{ij} + \frac{(1-2\nu)}{E} \delta_{ij} \sigma_m \quad (2.10)$$

where δ_{ij} is the Kronecker delta and the deviatoric stress tensor, S_{ij} , is given by

$$S_{ij} = \sigma_{ij} - \delta_{ij} \sigma_m \quad (2.11)$$

The plastic component of the equivalent strain can be calculated using deformation plasticity theory [16] as

$$\bar{\varepsilon}^p = \sqrt{\frac{2}{3} \varepsilon_{ij}^p \varepsilon_{ij}^p} \quad (2.12)$$

where the plastic strain tensor, ε_{ij}^p , can be written as

$$\varepsilon_{ij}^p = \frac{3}{2} \frac{\bar{\varepsilon}^p}{\sigma_e} S_{ij} \quad (2.13)$$

where equivalent stress, σ_e , and deviatoric stress tensor, S_{ij} , are given by Eqns (2.9) and (2.11), respectively. Assuming Ramberg-Osgood material model, the equivalent plastic strain can be defined by

$$\bar{\varepsilon}^p = \alpha \frac{\sigma_0}{E} \left(\frac{\sigma_e}{\sigma_0} \right)^N = A_p \sigma_e^N \quad (2.14)$$

and the total strain can be written as the sum of elastic and plastic components of the equivalent strain [17] as

$$\bar{\varepsilon}_{ij} = \frac{(1+\nu)}{E} S_{ij} + \frac{(1-2\nu)}{E} \delta_{ij} \sigma_m + \frac{3}{2} \alpha \frac{\sigma_0}{E} \left(\frac{\sigma_e}{\sigma_0} \right)^{N-1} \frac{S_{ij}}{\sigma_0} \quad (2.15)$$

2.2 Creep Deformation

In many industrial applications (*e.g.* aero engines and electric power plants) components operate at elevated temperatures. Creep deformation and crack growth is

an important failure mode commonly encountered at high temperatures. Creep is a time dependent deformation which may eventually cause failure. Therefore design procedures are needed to avoid excessive creep deformation and prevent fracture in industrial applications. Components that operate under high stresses at elevated temperatures are considerably expensive and it's not economical to replace them once damage or crack initiation is observed. Therefore components are not changed until they significantly pass their life time and become unsafe to use. It is an important challenge to predict a component's life time and keep the large scale component under operation for as long as possible [18].

2.2.1 Introduction to Creep

When metals are subjected to a constant stress σ at a constant temperature T which is greater than approximately 30% of the absolute melting temperature of the material, T_m , for an extended period, time dependent deformation may occur. This process, which is the predominant failure mode in many industrial applications, is generally known as creep.

There are two main creep mechanisms in polycrystalline materials; diffusion creep and dislocation creep (also known as power law creep). The creep mechanism for a given material under a constant stress at a given temperature can be determined using a deformation mechanism map. A deformation mechanism map for 316 stainless steel has been shown in Figure 2.1 for example. As seen in this figure, at relatively high temperatures and stresses dislocation creep is the dominant creep mechanism due to the glide and climb of dislocations. However, at lower stresses diffusion creep is expected to take place. Figure 2.1 also shows that dominant plastic deformation is expected at very high stresses.

All the creep tests in this work have been performed on 316H stainless steel material at 550 °C under relatively high stresses. Therefore, power law creep is expected to be the dominant mechanism for all the tests examined.

2.2.1.1 Creep Deformation Stages

The conventional way to perform a uniaxial creep test is to subject a specimen to a steady tensile load while it is maintained at a constant temperature. Time independent elastic and plastic strains (also known as the loading strain) are measured during the load up process. Subsequent to the specimen load up, the time dependent strain which is referred to as creep strain is often plotted against time as schematically shown in Figure 2.2. As seen in this figure, the first stage of the creep curve is generally referred

to as primary or transient creep, the second stage as secondary or steady state creep and the third stage as the tertiary or accelerating creep region.

2.2.1.1.1 Primary Creep

During the time-dependent creep deformation, there is a challenge between the strain hardening and thermal recovery processes. Thermal recovery processes assist deformation to be continued while work hardening processes reinforce the resistance of material against deformation. In the primary region, the strain hardening dominates and therefore the creep strain rate decreases.

2.2.1.1.2 Secondary Creep

In the secondary region the creep strain rate remains constant which in fact indicates a balance between work hardening and thermal softening [19]. The secondary stage is the most important creep region for most of the engineering materials because a component often spends the largest proportion of its life time in this region. The time-dependent strain variation during the secondary creep region is linear (see Figure 2.2) and the slope of the line is referred to as minimum (or steady state) creep strain rate, $\dot{\epsilon}_{min}$.

2.2.1.1.3 Tertiary Creep

The creep strain rate increases in the tertiary region and deformation continues until the fracture occurs. The creep strain at the rupture time, t_r , is often referred to as the uniaxial creep ductility, ϵ_f . The uniaxial creep fracture in a specimen can be caused by an increase in local stresses under a constant load, formation of a neck, voiding and/or cracking and over-aging [18]. Nucleation and coalescence of voids on the grain boundaries are generally expected to take place in the tertiary creep region. The voids link up to form micro cracks and subsequently the existing cracks on the grain boundaries join together to cause failure by creep rupture.

2.2.1.2 Creep Dependency

The shape of the creep curve varies by changing the stress and temperature. The instantaneous creep strain rates in all creep regions increase by raising the temperature or stress. As shown in Figure 2.3, the rupture time and creep ductility are also changed by increasing the temperature and/or stress.

2.2.1.2.1 Temperature Effects

As shown in Figure 2.3, under a given stress level the minimum creep strain rate increases by an increase in the temperature. However, a decrease in the rupture time is generally expected when the temperature increases. At higher temperatures ($T > 0.5T_m$) it is more likely that secondary and tertiary creep will be most pronounced. Though at lower temperatures ($T < 0.5T_m$) primary creep will predominate and fracture may occur at smaller values of creep ductility [18].

2.2.1.2.2 Stress Effects

By increasing the applied stress at a constant temperature, the creep strain rate increases whilst a drop in rupture time is generally observed. Depending on the material, testing temperature and the examined stress range, the creep strain at failure can vary by changing the stress level and pronounced tertiary creep regions may be observed at higher loads.

2.2.1.3 Creep under Variable Stresses

The rate of the creep deformation depends on the applied stress, temperature and the “state” of the specimen which is defined as the position of the specimen on the creep curve and may be given by either instantaneous time or creep strain at that moment [18]. Therefore, the creep strain rate can be written in one of the following forms:

$$\text{Strain hardening law } \dot{\varepsilon}^c = f(\sigma, T, \varepsilon^c) \quad (2.16)$$

$$\text{Time hardening law } \dot{\varepsilon}^c = f(\sigma, T, t) \quad (2.17)$$

In Eqn (2.16), which is known as strain hardening (SH) law, the state of the material is defined by creep strain whilst the time hardening (TH) law in Eqn (2.17) describes the creep strain rate as a function of time. Thus, depending on the way to identify the state of the material, the creep deformation rate can be defined by time hardening or strain hardening law.

Since the creep strain rates obtained from these two approaches are not exactly the same, the equation to best describe the behaviour of the material is required to be chosen. Time hardening law may be appropriate when material ageing is significant (large tertiary creep region) whilst strain hardening approach is suitable in cases that work hardening is dominant (primary creep region).

2.2.1.3.1 Strain and Life Fraction Rules

Strain hardening and time hardening laws can be written in the normalised form and referred to as strain fraction (SF) and life fraction (LF) rules, respectively.

$$\text{Strain fraction rule } \dot{\varepsilon}^c = f\left(\sigma, T, \frac{\varepsilon^c}{\varepsilon_f}\right) \quad (2.18)$$

$$\text{Life fraction rule } \dot{\varepsilon}^c = f\left(\sigma, T, \frac{t}{t_r}\right) \quad (2.19)$$

where ε_f is the creep ductility and t_r is the time to rupture. In the current work, the tests have been run at a constant temperature and therefore the effect of temperature change in SF and LF equations can be neglected. Thus strain fraction and life fraction rules can be used to predict the rupture time and creep strain at failure under variable stresses. Under uniaxial stress conditions, the prediction can be made using the following formulae

$$\sum_{\sigma, T} \frac{\varepsilon^c(\sigma, T)}{\varepsilon_f(\sigma, T)} = 1 \quad (2.20)$$

$$\sum_{\sigma, T} \frac{t(\sigma, T)}{t_r(\sigma, T)} = 1 \quad (2.21)$$

Eqn (2.20) suggests that the failure happens when the sum of creep strain to creep ductility ratios attains unity. Similarly the prediction of the rupture time can be made using the life fraction rule in Eqn (2.21).

2.2.2 Power Law Representations of Creep

For materials at the range of stresses and temperatures in the dislocation creep regime, the stress and temperature dependency of the steady state creep strain rate, $\dot{\varepsilon}_s$, can be represented by a power law equation written as

$$\dot{\varepsilon}_s = C e^{\frac{Q}{RT}} \sigma^n \quad (2.22)$$

where C is a constant, Q is the creep activation energy, R is the universal gas constant, T is the absolute temperature, σ is the applied stress (which can be considered as the equivalent stress) and n is the power law creep stress exponent. This power law relationship is valid for the steady state creep region only and the subscript “s” in the formula indicates that these constants have been defined for secondary creep conditions. Assuming that a material is creeping at a constant temperature under uniaxial stress conditions, the steady state creep strain rate for the power law creeping material can be expressed as

$$\dot{\varepsilon}_s = \dot{\varepsilon}_0 \left(\frac{\sigma}{\sigma_0} \right)^n = A \sigma^n \quad (2.23)$$

where $\dot{\epsilon}_0$ and σ_0 are the normalising strain rate and stress for creep, respectively, and A is a temperature dependent material constant which can be defined as $A = \dot{\epsilon}_0 / \sigma_0^n$. The value of $\dot{\epsilon}_0$ in Eqn (2.23) is usually assumed to be 1 h^{-1} [20]. If the minimum creep strain rate is plotted versus nominal stress in log-log axes, the values of the steady state power law constants A and n in Eqn (2.23), which is also known as the Norton creep law [21], can be provided by the equation of the line of best fit to the data.

2.2.2.1 Primary Creep

At a constant stress, the creep strain in the primary region, ϵ_p^c , may be represented by a power law relation

$$\epsilon_p^c = C \sigma^{n_p} t^p \quad (2.24)$$

where C , n_p and p are material constants which are temperature dependent. Using strain hardening and time hardening laws described above, the primary creep strain rate can be found by differentiating Eqn (2.24) as shown in Eqns (2.25) and (2.26), respectively

$$\dot{\epsilon}_p^c = p C^{1/p} \sigma^{n_p/p} \epsilon^{(1-1/p)} \quad (2.25)$$

$$\dot{\epsilon}_p^c = p C \sigma^{n_p} t^{p-1} \quad (2.26)$$

2.2.2.2 Total Creep Deformation

The total strain accumulated in a body can be calculated by adding time-independent and time-dependent strain terms as previously shown in Eqn (2.1). Similarly, total creep strain rate can be obtained by

$$\dot{\epsilon}_{total} = \dot{\epsilon}^e + \dot{\epsilon}^p + \dot{\epsilon}^c(t) \quad (2.27)$$

The elastic $\dot{\epsilon}^e$ and plastic $\dot{\epsilon}^p$ strain rates may be assumed negligible under a constant stress level and therefore the total strain rate in a body is considered identical to the creep strain rate.

2.2.3 Creep Failure Mechanisms

When a material is subjected to a tensile stress under creep conditions, the extended necking may lead to ultimate failure. The creep ductility and rupture time can be found at this point where the failure takes place. Intergranular and intragranular (transgranular) microstructural mechanisms can be observed in creep deformation of a metallic material. Intergranular creep deformation mechanism may exist on the grain boundaries in polycrystals whereas intragranular mechanism can occur within

individual grains of a polycrystalline material. One or both of these microstructural deformation modes may be observed during creep deformation of a material. Nucleation of voids and formation of micro cracks can be described by the creep failure mechanism. Intergranular deformation is usually observed under low stress levels at high temperatures whilst transgranular failure is associated with high stress levels and low temperatures. The experimental results from previous investigations on ex-service 316H stainless steel show that intergranular failure is the pre-dominant deformation mechanism in uniaxial creep deformation and creep crack growth tests on this material at 550 °C [22-24].

2.2.3.1 Creep Rupture Time

Similar to the power law representation of the steady state creep strain rate in Eqn (2.23), the time to rupture of the material under a range of stress levels at a constant temperature may be approximated by a power law relationship written as

$$t_r = \frac{\varepsilon_{f0}}{\dot{\varepsilon}_0} \left(\frac{\sigma}{\sigma_0} \right)^{-v_r} = B_r \sigma^{-v_r} \quad (2.28)$$

where σ is the applied stress, ε_{f0} is the uniaxial failure strain at stress σ_0 , v_r is the rupture life stress exponent and B_r is the temperature dependant rupture constant which can be defined as $B_r = \varepsilon_{f0} / (\dot{\varepsilon}_0 \sigma_0^{-v_r})$. The power law stress exponent v_r and coefficient B_r in Eqn (2.28) can be found using the equation of the line of best fit to the creep test data.

2.2.3.2 Average Creep Strain Rate

For a power law creeping material the minimum creep strain rate can be described using Eqn (2.23) [21]. In order to account for all three creep regions, an average creep strain rate for a given stress and temperature is described as

$$\dot{\varepsilon}_A^c = \varepsilon_f / t_r \quad (2.29)$$

where ε_f is the creep ductility and t_r is the time to rupture. The stress dependency of the average creep strain rate, $\dot{\varepsilon}_A^c$, can be described using a simple power law relationship similar to the one for minimum creep strain rate

$$\dot{\varepsilon}_A^c = \dot{\varepsilon}_0 \left(\frac{\sigma}{\sigma_0} \right)^{n_A} = A_A \sigma^{n_A} \quad (2.30)$$

where A_A and n_A are the power law stress coefficient and exponent, respectively, and subscript "A" is to emphasize that the constants in the power law are determined based on the average creep strain rate and are not associated with secondary creep

properties. Similar to the steady state creep strain rate in Eqn (2.23), $\dot{\epsilon}_0$ is often taken as 1 h^{-1} and the power law constants n_A and $A_A = \dot{\epsilon}_0 / \sigma_0^{n_A}$ can be determined from a regression fit to the uniaxial creep test data. A schematic illustration of the minimum and average creep strain rates is shown in Figure 2.4. As seen in the figure, the average creep strain rate is generally greater than or equal to the minimum creep strain rate in the secondary creep region.

Combining Eqns (2.28)–(2.30) the creep strain at failure for a given stress and temperature can be defined as

$$\epsilon_f = A_A B_r \sigma^{n_A - v_r} \quad (2.31)$$

Eqn (2.31) shows that the creep ductility can be stress independent when the average creep strain rate and rupture time power law stress exponents are identical (*i.e.* $n_A = v_r$). However, increasing creep ductility by an increase in the applied stress is expected when $n_A > v_r$.

2.3 Creep Damage Mechanism and Models

The processes which may lead to material degradation as a result of micro structural changes during deformation are generally known as damage. In metallic materials at high temperatures creep damage corresponds to intergranular void nucleation, growth and coalescence which lead to the formation of micro cracks along the grain boundaries. Although the voids can be nucleated in primary and secondary creep regions, creep damage accumulation is mainly noticeable in tertiary creep region which may eventually lead to creep rupture.

Creep damage in polycrystalline materials is usually represented by void nucleation and growth along the grain boundaries. There are some damage models which can be applied to predict the time to rupture as a result of damage accumulation. These predictions can be made under uniaxial and multiaxial stress conditions.

2.3.1 Coupled Damage

The concept of damage parameter, ω , was originally used in the principle of continuum damage mechanics which was suggested by Kachanov [25] and Rabotnov [26, 27]. This parameter provides a measure of the extent of damage in a body and has a value ranging from zero, for undamaged body, to unity for fully damaged material. According to Rabotnov's formulation, damage accumulation and strain evolution are assumed to be a function of applied stress and the level of existing damage. For a

power law creeping material, instantaneous creep strain rate and damage accumulation rate can be defined as

$$\dot{\varepsilon}_{\omega}^c = \dot{\varepsilon}_0 \left(\frac{\sigma}{\sigma_0} \right)^n \frac{1}{(1-\omega)^m} = \frac{a\sigma^n}{(1-\omega)^m} \quad (2.32)$$

$$\dot{\omega} = \dot{\omega}_0 \left(\frac{\sigma}{\sigma_0} \right)^{\varphi} \frac{1}{(1-\omega)^{\psi}} = \frac{c\sigma^{\varphi}}{(1-\omega)^{\psi}} \quad (2.33)$$

where $\dot{\varepsilon}_{\omega}^c$ is the instantaneous creep strain rate which takes the effect of damage into account, and $m, n, \varphi, \psi, a, c$ and $\dot{\omega}_0$ are material constants [27].

The time to rupture and creep ductility can be obtained by integrating equations above and using the limits $t = 0, \omega = 0, \varepsilon = 0$ and $t = t_r, \omega = 1$ and $\varepsilon = \varepsilon_f$.

$$t_r = \frac{1}{\dot{\omega}_0} \left(\frac{\sigma}{\sigma_0} \right)^{-\varphi} \frac{1}{1+\psi} = \frac{\sigma^{-\varphi}}{c(1+\psi)} \quad (2.34)$$

$$\varepsilon_f = \frac{\dot{\varepsilon}_0}{\dot{\omega}_0} \left(\frac{\sigma}{\sigma_0} \right)^{n-\varphi} \frac{1}{1+\psi-m} = \frac{a\sigma^{n-\varphi}}{c(1+\psi-m)} \quad (2.35)$$

The rate of creep damage accumulation can be written in a general form as [18]

$$\dot{\omega} = \frac{\dot{\varepsilon}^c(\sigma, \omega)}{\varepsilon_f} \quad (2.36)$$

where $\dot{\varepsilon}^c$ is the instantaneous creep strain rate. Eqn (2.36) suggests that the instantaneous damage accumulation rate is a function of the level of existing damage and is generally referred to as coupled damage model. By integrating Eqn (2.36), damage parameter can be expressed as

$$\omega = \frac{\varepsilon^c}{\varepsilon_f} \quad (2.37)$$

Under multiaxial stress conditions, the equations above can be used by applying equivalent creep strain, $\bar{\varepsilon}^c$, and strain rate, $\dot{\bar{\varepsilon}}^c$, which are corresponding to the equivalent stress, σ_e . Also to account for multiaxial stress conditions in Eqn (2.37), multiaxial creep ductility, ε_f^* , must be considered instead of uniaxial creep strain at failure ε_f . Kachanov type [28] and Lui-Murakami [29] continuum damage material behaviour models are the coupled damage models which are often utilised to quantify creep damage in steel weldments [30].

2.3.2 Uncoupled Damage

In uncoupled damage model the assumption has been made that the instantaneous creep strain rate depends on the stress only and is independent of the existing level of damage, ω . In this model, primary-secondary creep deformation is used to characterise tertiary creep region where creep ductility for a uniaxial creep rupture test at a given stress is obtained at the creep rupture time, t_r (see Figure 2.5). In the uncoupled damage approach, the effect of damage on creep deformation is ignored and therefore tertiary creep region is not considered in calculations. In this approach failure happens when creep strain reaches a critical value ε_{crit} (strain based method) or the test duration attains the creep rupture time, t_r (rupture time based method) [31].

Using strain-based uncoupled damage approach, the rate of damage accumulation in a creeping body can be written as

$$\dot{\omega} = \frac{\dot{\varepsilon}^c(\sigma)}{\varepsilon_{crit}} \quad (2.38)$$

Alternatively the creep damage rate using rupture time-based uncoupled damage approach is given by

$$\dot{\omega} = \frac{1}{t_r(\sigma)} \quad (2.39)$$

This equation may be written in another form by applying the definition of average creep strain rate

$$\dot{\omega} = \frac{\dot{\varepsilon}_A}{\varepsilon_f} \quad (2.40)$$

As shown in Figure 2.5, if the creep deformation rate in Eqn (2.38) is described by the average creep strain rate, the creep strain at failure will be attained at the rupture time and the critical creep ductility will become the same as the creep strain at failure $\varepsilon_{crit} = \varepsilon_f$. Therefore Eqns (2.38) and (2.40) will become identical.

The equations above are applied for uniaxial stress conditions, however under multiaxial stress conditions the same formulation can be used by replacing critical multiaxial creep strain (which would be identical to multiaxial creep ductility using the definition of average creep strain rate) and the equivalent stress in all calculations.

Comparing both damage models it can be realised that the uncoupled creep damage approach is easier to use. However, the coupled damage is a more accurate method because of taking into account the effect of damage parameter on creep strain rate and consequently letting a large proportion of stress to be redistributed [32].

2.4 Void Nucleation and Growth Models

Creep damage is a process which consists of the nucleation and growth of voids along the grain boundaries. A brief study of the void nucleation process is presented in this section and the most common stable void growth models are described.

The nucleation of intergranular defects, which may be in the form of wedge cracks or voids (cavities), is associated with the sliding of grains and thus the subsequent stress concentration [33]. Stress concentration which leads to void nucleation or wedge crack formation usually occurs under one of the following conditions [34]

- i. Slip bands impinging on a grain boundary
- ii. Grain boundary ledges which can be occurred by slip bands in the vicinity of grains. Ledges may exist in either tension or compression but only the ones under tension can lead to cavity nucleation.
- iii. Triple junctions which are the points where three grain boundaries meet and wedge cracks are most likely introduced.
- iv. Grain boundary particles which usually slow down sliding and thus stress concentration may be introduced.

According to experimental observations in austenitic stainless steels, cavities are mainly nucleated where Carbides are present on the grain boundaries [35]. The maximum principal stress normal to the grain boundary has also been found to play a key role in cavity nucleation [36]. In service exposed materials, pre-nucleated voids are most likely to exist in the samples tested under high loads at elevated temperatures. Therefore, the nucleation time can generally be ignored in failure time calculations [35]. At low stresses diffusive processes may control the void growth and creep deformation would be relatively small, however a void can grow by creep deformation of the surrounding material at high stresses. Note that if creep deformation is described by a power law creep model for a given material, then mathematical void growth models derived for perfectly plastic materials can also be related to creep [33].

2.4.1 Ductility Exhaustion Approach

Under multiaxial stress conditions creep damage can be calculated using

$$\omega = \int_0^t \frac{\dot{\varepsilon}^c}{\varepsilon_f^*(\dot{\varepsilon}^c, \sigma)} dt \quad (2.41)$$

where $\dot{\varepsilon}^c$ is the instantaneous equivalent creep strain rate and ε_f^* is the multiaxial creep ductility which is a function of stress and equivalent creep strain rate [37]. Eqn (2.41) is known as the “ductility exhaustion” approach and is currently being used in R5

“assessment procedure for the high temperature response of structures” [38]. In order to calculate creep damage, $\dot{\varepsilon}^c$ and ε_f^* need to be known. The multiaxial creep ductility may be related to the uniaxial creep ductility using a multiaxial strain factor (MSF), in such a way that

$$\varepsilon_f^* = \varepsilon_f \times \text{MSF} \quad (2.42)$$

Having known the creep ductility from uniaxial creep rupture tests, multiaxial creep strain at failure can be estimated using void growth models. The MSF in these models is often a function of stress triaxiality, which can be defined as the ratio of the mean stress to the equivalent von-Mises stress [39, 40]. Using the existing void growth models the effect of stress state on ductility can be determined.

Creep damage development at grain boundaries is generally expressed in terms of void nucleation and void growth mechanisms. A number of models are available in the literature to describe void nucleation and/or void growth, which may be used to predict the time to failure or strain at failure under both uniaxial and multiaxial stress conditions. Some of these cavity growth estimation approaches are based on a single model whilst others consider the transition between different mechanisms of void growth. A summary of the existing void growth models are described next.

2.4.2 Void Growth Models

The common void growth models which are often used to calculate creep damage in engineering materials in conjunction with the ductility exhaustion approach are Manjoine [41, 42], Rice and Tracey [39], Cocks and Ashby [43, 44] and Hales [45]. Rice and Tracy, and Cocks and Ashby models are usually employed in damage calculations for stainless steels.

2.4.2.1 Rice and Tracey Plastic Damage Model

Rice and Tracey [39] proposed the expression below for void growth by rigid plastic deformation of the surrounding matrix

$$\frac{\dot{r}}{r\dot{\varepsilon}} = 0.283e^{(3\sigma_m/2\sigma_e)} = e^{(3\sigma_m/2\sigma_e - \sqrt{2})} \quad (2.43)$$

where r is the void radius, $\dot{\varepsilon}$ is the strain rate and σ_m / σ_e is the triaxiality. Note that the voids are considered to have spherical shapes in this model. By integrating Eqn (2.43) assuming that initial and final void sizes are independent of triaxiality, the ratio of multiaxial to uniaxial strain at failure can be written as

$$\frac{\varepsilon_f^*}{\varepsilon_f} = 1.65e^{(-3\sigma_m/2\sigma_e)} = e^{(1/2-3\sigma_m/2\sigma_e)} \quad (2.44)$$

This relationship has a good agreement with the numerical simulation results at high triaxialities ($\sigma_m / \sigma_e > 80\%$). Rice and Tracey also proposed a more general void growth expression for the full range of triaxialities.

$$\frac{\dot{r}}{r\dot{\varepsilon}} = 0.558\sinh\left(\frac{3\sigma_m}{2\sigma_e}\right) + 0.008\cosh\left(\frac{3\sigma_m}{2\sigma_e}\right) \quad (2.45)$$

The second term in Eqn (2.45) is much smaller than the first term and thus can be neglected. By integrating both sides of the equation and using the limits of $\varepsilon = 0$, $r = r_i$ and $\varepsilon = \varepsilon_f$, $r = r_f$, new relationships for calculating uniaxial and multiaxial strain at failure can be found and expressed as

$$\varepsilon_f = \frac{1.792\ln(r_f / r_i)}{\sinh(3\sigma_m / 2\sigma_e)} \quad (2.46)$$

$$\frac{\varepsilon_f^*}{\varepsilon_f} = \frac{0.521}{\sinh(3\sigma_m / 2\sigma_e)} \quad (2.47)$$

Using Needleman's assumption that the ratio of multiaxial to uniaxial strains is not significantly influenced by void interaction effects [46], Rice and Tracey model can be considered valid to describe cavity growth in a power law creeping material [47].

2.4.2.2 Cocks and Ashby Creep Damage Model

Cocks and Ashby proposed a model [43, 44] for intergranular nucleation, growth and coalescence of an array of spherical voids in a power law creeping material at elevated temperature (as shown in Figure 2.6). In this model the time to failure, t_f , under constant stress is defined as

$$t_f = t_n + \frac{\lambda}{(n+1)\dot{\varepsilon}_s} \ln\left(\frac{1}{(n+1)f_i}\right) \quad (2.48)$$

where t_n is the void nucleation time, $\dot{\varepsilon}_s$ is the steady state creep strain rate, n is steady state creep stress exponent, f_i is the original area fraction of voids and λ is a triaxiality dependent function which can be calculated using

$$\lambda = \frac{1}{\sinh\left(\frac{2(n-1/2)\sigma_m}{(n+1/2)\sigma_e}\right)} \quad (2.49)$$

In Cocks and Ashby model the creep strain at failure, ε_f , can be calculated by

$$\varepsilon_f = t_f \dot{\varepsilon}_s + \frac{0.2l}{g} \quad (2.50)$$

where g is the grain size and $2l$ is the centre to centre void spacing at fracture.

According to Cocks and Ashby model, the ratio of multiaxial to uniaxial failure times can be described by the expression below assuming that the nucleation time, t_n , is negligible compared to the void growth time, taking the original area fraction of voids to be the same under both uniaxial and multiaxial stress states, and also neglecting the second term of Eqn (2.50) which reflects the effect of tertiary creep.

$$\frac{t_f^*}{t_f} = \frac{\sinh \left[\frac{2(n-1/2)}{3(n+1/2)} \right]}{\sinh \left[\frac{2(n-1/2)}{(n+1/2)} \frac{\sigma_m}{\sigma_e} \right]} \quad (2.51)$$

Knowing that the average creep strain rate under multiaxial and uniaxial stress conditions can be defined as [33],

$$\dot{\varepsilon}_A = \frac{\varepsilon_f}{t_f} = \frac{\varepsilon_f^*}{t_f^*} \quad (2.52)$$

the multiaxial strain factor from Cocks and Ashby model is given by

$$MSF = \frac{t_f^*}{t_f} = \frac{\varepsilon_f^*}{\varepsilon_f} = \frac{\sinh \left[\frac{2(n-1/2)}{3(n+1/2)} \right]}{\sinh \left[\frac{2(n-1/2)}{(n+1/2)} \frac{\sigma_m}{\sigma_e} \right]} \quad (2.53)$$

As seen in Eqn (2.53), the multiaxial strain factor calculated using Cocks and Ashby model is a function of triaxiality and creep stress exponent. The variation of MSF against triaxiality calculated for a given power law stress exponent of $n = 7.5$ using this model is shown, as an example, in Figure 2.7. Cocks and Ashby model is commonly used in analytical and numerical studies of the creep damage accumulation and crack growth behaviour in Type 316H stainless steel at elevated temperatures [48-50].

2.4.3 Stress Modified Ductility Exhaustion Approach

As shown in Eqn (2.41), the multiaxial creep strain at failure in ductility exhaustion approach is assumed to be a function of the stress and equivalent creep strain rate. The “stress modified ductility exhaustion” approach is an improved method developed by Spindler based on the recent studies of creep damage calculations in experimental tests. The validation of this approach under multiaxial and uniaxial stress conditions

using the available test data for 316H stainless steel can be found in [51]. Using this approach, the creep damage predictions have been improved for the tests in which the creep dwells are at an intermediate stress level. The current R5 approach has been found to be extremely conservative for intermediate dwell cycles, however longer life time predictions could be made by applying the new stress modified ductility exhaustion approach [52-56]. In parallel to stress modified ductility exhaustion model validation, a new approach was developed to evaluate creep damage from constant stress and strain rate experimental data. This new approach which is called “reverse ductility modelling” (RDM) is based on the test data obtained from [54] and [55] on Type 347 weld metal and 316H stainless steel, respectively. Using the test data on Type 316H [55] it was observed that the best prediction of creep damage at failure was obtained by applying “stress modified ductility exhaustion approach” in conjunction with RDM and total inelastic strain.

It was shown by Spindler in [51] that comparing the creep damage calculated from creep strain and inelastic strain, the latter leads to considerably longer life time predictions. Using “stress modified ductility exhaustion” approach, creep damage can be calculated using

$$\omega = \int_0^{t_h} \frac{\dot{\bar{\epsilon}}^c}{\epsilon_f^*(\dot{\bar{\epsilon}}^c, \sigma_1, T)} dt \quad (2.54)$$

where t_h is the dwell time and $\epsilon_f^*(\dot{\bar{\epsilon}}^c, \sigma_1, T)$ is the multiaxial inelastic strain at failure which is a function of temperature, maximum principal stress and von-Mises inelastic strain rate in the creep dwell. In order to calculate creep damage using this formulation, values of the equivalent inelastic strain rate in the creep dwell and inelastic strain at failure are required.

2.4.3.1 Inelastic Multiaxial Strain at Failure

As shown and discussed in [51], there are three failure mechanisms observed in uniaxial test data for Type 316H stainless steel at a (constant) high temperature; plastic hole growth, diffusion controlled cavity growth and constrained cavity growth. A combined model consists of all these three mechanisms has been proposed by Spindler [51] to calculate the multiaxial strain at failure by

$$\varepsilon_f^* = \text{Max} \left\{ \begin{array}{l} \text{MIN} \left[\varepsilon_U e^{\left(\frac{1}{2} \frac{3\sigma_m}{2\sigma_e} \right)}, A_1 e^{\left(\frac{AG_{AC}}{RT} \right)} \left(\dot{\varepsilon}^c \right)^{n_1} (\sigma_1)^{-m_1} \frac{\sigma_e}{\sigma_1} e^{\left(\frac{1}{2} \frac{3\sigma_m}{2\sigma_e} \right)} \right] \\ , \text{MIN} \left[\varepsilon_U e^{\left(\frac{1}{2} \frac{3\sigma_m}{2\sigma_e} \right)}, A_2 e^{\left(\frac{AG_C}{RT} \right)} (\sigma_1)^{-m_2} \frac{\sigma_e}{\sigma_1} e^{\left(\frac{1}{2} \frac{3\sigma_m}{2\sigma_e} \right)} \right] \end{array} \right\} \quad (2.55)$$

The empirical failure strain coefficients in Eqn (2.55) have been determined for Type 316H at temperatures ranging from 500 °C to 600°C in [51]. By replacing the values of coefficients in Eqn (2.55), the multiaxial strain at failure in Type 316H SS can be defined as

$$\varepsilon_f^* = \text{Max} \left\{ \begin{array}{l} \text{MIN} \left[0.5764 e^{\left(\frac{1}{2} \frac{3\sigma_m}{2\sigma_e} \right)}, e^{\left(\frac{15975}{T} \right)} \left(\dot{\varepsilon}^c \right)^{0.9316} (\sigma_1)^{-2.495} \frac{\sigma_e}{\sigma_1} e^{\left(\frac{1}{2} \frac{3\sigma_m}{2\sigma_e} \right)} \right] \\ , \text{MIN} \left[0.5764 e^{\left(\frac{1}{2} \frac{3\sigma_m}{2\sigma_e} \right)}, 197930 (\sigma_1)^{-2.495} \frac{\sigma_e}{\sigma_1} e^{\left(\frac{1}{2} \frac{3\sigma_m}{2\sigma_e} \right)} \right] \end{array} \right\} \quad (2.56)$$

2.4.3.2 Stress Modified Ductility Exhaustion Damage

The stress modified ductility exhaustion damage for 316H at 550 °C can be calculated by [51]

$$\omega = \sum \delta \varepsilon_c^* \times \text{Max} \left\{ \begin{array}{l} \text{MIN} \left[0.5764 e^{\left(\frac{1}{2} \frac{3\sigma_m}{2\sigma_e} \right)}, e^{\left(\frac{15975}{T} \right)} \left(\dot{\varepsilon}^c \right)^{0.9316} (\sigma_1)^{-2.495} \frac{\sigma_e}{\sigma_1} e^{\left(\frac{1}{2} \frac{3\sigma_m}{2\sigma_e} \right)} \right] \\ , \text{MIN} \left[0.5764 e^{\left(\frac{1}{2} \frac{3\sigma_m}{2\sigma_e} \right)}, 197930 (\sigma_1)^{-2.495} \frac{\sigma_e}{\sigma_1} e^{\left(\frac{1}{2} \frac{3\sigma_m}{2\sigma_e} \right)} \right] \end{array} \right\}^{-1} \quad (2.57)$$

Another simplified method has also been proposed by Spindler to calculate creep damage using a lower bound of the total inelastic strain at failure which is independent of the strain rate. This model can be expressed as

$$\omega = \frac{\varepsilon_{f,observed}^*}{\varepsilon_{f,predicted}^*} = \varepsilon_{f,observed}^* \left[\varepsilon_{f,LB} \frac{\sigma_e}{\sigma_1} e^{\left(\frac{1}{2} \frac{3\sigma_m}{2\sigma_e} \right)} \right]^{-1} \quad (2.58)$$

where $\varepsilon_{f,LB}$ is the uniaxial lower bound total inelastic strain at failure and $\varepsilon_{f,observed}^*$ is the observed equivalent inelastic strain at failure. It has been shown in [51] that conservative predictions of creep damage can be obtained by employing the lower bound total inelastic strain at failure, which is independent of the stress and strain rate, in calculations.

2.5 Linear Elastic Fracture Mechanics (LEFM)

In a cracked or notched body the stress local to the crack tip is considerably higher than the remote stress. Stress intensity factor (SIF), K , is the fracture mechanics parameter which describes stress distribution in the vicinity of a sharp crack under linear elastic conditions [57]. In the majority of fracture mechanics tests, the crack growth occurs under mode "I", where the stress is normal to the crack plane. Therefore the stress intensity factor is designated K_I . However for simplicity, the subscript "I" is often ignored and the stress intensity factor is written as K .

2.5.1 Stress Intensity Factor (SIF), K

By defining a cylindrical coordinate system for an infinite plate made of a linear elastic material with a centre crack length of $2a$ subjected to a uniform-biaxial state of stress, σ , (see Figure 2.8) and taking the crack tip as the origin, the normal and shear stress fields local to the sharp crack can be written as functions of the radial distance from the crack tip, r , and the crack tip angle, θ , [18]

$$\sigma_{xx} = \left(\frac{\sigma\sqrt{\pi a}}{\sqrt{2\pi r}} \right) \times \cos(\theta/2) \times (1 - \sin(\theta/2) \times \sin(3\theta/2)) \quad (2.59)$$

$$\sigma_{yy} = \left(\frac{\sigma\sqrt{\pi a}}{\sqrt{2\pi r}} \right) \times \cos(\theta/2) \times (1 + \sin(\theta/2) \times \sin(3\theta/2)) \quad (2.60)$$

$$\tau_{xy} = \left(\frac{\sigma\sqrt{\pi a}}{\sqrt{2\pi r}} \right) \times \cos(\theta/2) \times \sin(\theta/2) \times \cos(3\theta/2) \quad (2.61)$$

In linear elastic fracture mechanics (LEFM) the expressions for the stress fields are obtained by assuming negligible plasticity in the first instance. Assuming LEFM conditions in a cracked body, the effects of remote loading, crack size and geometry can be described by a single parameter which is termed stress intensity factor, K . The stresses around a sharp crack tip in a linear-elastic material can be defined as

$$\sigma_{ij} = \frac{K}{\sqrt{2\pi r}} f_{ij}(\theta) \quad (2.62)$$

This equation is valid under both plane stress ($\sigma_{zz} = 0$) and plane strain ($\sigma_{zz} = \nu(\sigma_{xx} + \sigma_{yy})$) conditions [58].

For an infinite plate subjected to uniform stresses normal to a crack of size $2a$, the stress intensity factor can be expressed as [18]

$$K = \sigma\sqrt{\pi a} \quad (2.63)$$

In general the stress intensity factor for a sharp crack in a finite body can be written as

$$K = Y(a/W)\sigma\sqrt{a} \quad (2.64)$$

where σ is the applied stress, a is the crack length and $Y(a/W)$ is a dimensionless shape function. The solution of $Y(a/W)$ for conventional fracture geometries can be found in fracture mechanics textbooks such as [59]. For a side grooved compact tension, C(T), specimen the nominal stress can be defined as

$$\sigma = \frac{P}{BW} \sqrt{\frac{B}{B_n}} \quad (2.65)$$

where P is the applied load, W is the width of the sample, B is the specimen thickness and B_n is the net thickness between the side grooves. The empirical solution of the shape function for a standard C(T) specimen with $a/W > 0.45$ is given in [60] and is defined as

$$Y(a/W) = \sqrt{W/a} \left[\frac{2 + a/W}{(1 - a/W)^{3/2}} \right] f(a/W) \quad (2.66)$$

where a is the crack length, W is the width and $f(a/W)$ is calculated using

$$f(a/W) = 0.886 + 4.64(a/W) - 13.32(a/W)^2 + 14.72(a/W)^3 - 5.6(a/W)^4 \quad (2.67)$$

2.5.2 Energy Release Rate, G

Energy release rate, G , is a fracture mechanics parameter which characterises the crack tip region in a linear elastic material. The physical meaning of G can be defined as the amount of energy per unit area required to extend a crack. Assuming that dU is the required energy to extend the crack by da , the energy release rate for a linear elastic material can be expressed as [18]

$$G = \frac{dU}{Bda} \quad (2.68)$$

where B is the specimen thickness. The linear elastic fracture mechanics parameters, K and G , are related by

$$G = \frac{K^2}{E'} \quad (2.69)$$

where E' is the effective Young's modulus, which is equal to E and $E/(1-\nu^2)$ under plane stress (PS) and plane strain (PE) conditions, respectively.

2.5.3 Small Scale Yielding (SSY)

An estimated measure of the plastic zone size, r_p , in a cracked body for a power law hardening material can be calculated by

$$r_p = \frac{1}{\beta_1 \pi} \left(\frac{N-1}{N+1} \right) \left(\frac{K}{\sigma_y} \right)^2 \quad (2.70)$$

where K is the stress intensity factor, N is the power law hardening exponent, σ_y is the yield stress, $\beta_1=1$ for plane stress and $\beta_1=3$ for plane strain conditions. It can be seen in Eqn (2.70) that the plastic zone size depends on the power law hardening plasticity exponent, N , and is three times larger under plane stress conditions.

Under small scale yielding (SSY) conditions, the linear elastic fracture mechanics parameter, K , may still be used to characterise the stress distribution outside the plastic zone. The stress intensity factor under SSY conditions can be calculated by considering an effective crack length defined as [61]

$$a_e = a + \frac{1}{1 + (P/P_{p0})^2} r_y \quad (2.71)$$

in Eqn (2.64), where P_{p0} is a geometry dependent normalising load and r_y is the plastic zone correction factor which can be assumed to be half of the plastic zone size (*i.e.* $r_p = 2r_y$).

2.6 Non-Linear Fracture Mechanics

Linear elastic fracture mechanics is suitable for conditions that the plastic zone around the crack tip is sufficiently small. When the plastic zone correction factor becomes significantly large, the stress intensity factor cannot characterise stress distribution at the crack tip under SSY conditions any longer. For an elastic-plastic material with relatively large plastic zone size, the non-linear fracture mechanics parameter, J , may describe the stress and strain distribution around the crack tip.

2.6.1 J -Integral

For a non-linear elastic cracked body J -integral can be defined as [62]

$$J = \int_{\Gamma} (W_s dy - T_i \frac{\partial u_i}{\partial x} ds) \quad (2.72)$$

where Γ is an anticlockwise integration contour around the crack tip, T_i is the traction vector, u_i is the displacement vector, s is the arc length along Γ and W_s is the strain energy density which can be expressed as

$$W_s = \int_0^{\varepsilon_{ij}} \sigma_{ij} d\varepsilon_{ij} \quad (2.73)$$

where σ_{ij} and ε_{ij} are the stress and strain tensors, respectively. As shown in Figure 2.9, the traction vector can be defined as the product of the stress tensor and the outward normal vector to the contour Γ .

$$T_i = \sigma_{ij} n_j \quad (2.74)$$

The starting point of the anticlockwise contour Γ is at the lower surface of the crack face and the ending point is on the upper crack face. It has been shown in [62] that J -integral is time and path independent. The J -integral for a non-linear elastic cracked body can also be defined in terms of the energy release rate as [62]

$$J = -\frac{\partial U}{\partial A_c} \quad (2.75)$$

where U is the potential energy and A_c is the crack area.

2.6.2 HRR Stress and Strain Distribution Fields

For a non-linear power law hardening material the stress and strain distribution fields near the crack tip can be defined in terms of J -integral by HRR (Hutchinson [63, 64] and Rosengren&Rice [65]) equations as

$$\frac{\sigma_{ij}}{\sigma_{p0}} = \left[\frac{J}{\alpha \varepsilon_{p0} \sigma_{p0} I_N r} \right]^{\frac{1}{N+1}} \tilde{\sigma}_{ij}(\theta; N) \quad (2.76)$$

$$\frac{\varepsilon_{ij}}{\varepsilon_{p0}} = \alpha \left[\frac{J}{\alpha \varepsilon_{p0} \sigma_{p0} I_N r} \right]^{\frac{N}{N+1}} \tilde{\varepsilon}_{ij}(\theta; N) \quad (2.77)$$

where $\tilde{\sigma}_{ij}$ and $\tilde{\varepsilon}_{ij}$ are functions of crack tip angle, θ , and power law hardening stress exponent, N . A table of HRR singular field quantities for $\tilde{\sigma}_{ij}$ and $\tilde{\varepsilon}_{ij}$ has been provided by Shih in [66]. In Eqns (2.76) and (2.77), α , σ_{p0} and ε_{p0} are Ramberg-Osgood material model constants (see Eqn (2.7)), r is the radial distance from the crack tip and I_N is a dimensionless constant which can be estimated using [18]

$$\text{Plane stress: } I_N = 7.2 \sqrt{0.12 + \frac{1}{N} - \frac{2.9}{N}} \quad (2.78)$$

$$\text{Plane strain: } I_N = 10.3 \sqrt{0.13 + \frac{1}{N} - \frac{4.6}{N}} \quad (2.79)$$

It can be observed in the HRR equations that the relationship between the stress σ_{ij} and strain ε_{ij} fields, and also angular stress $\tilde{\sigma}_{ij}$ and strain $\tilde{\varepsilon}_{ij}$ constants, can be represented by a power law (*i.e.* $\tilde{\varepsilon}_{ij} = \tilde{\sigma}_{ij}^N$).

2.6.3 *J*-Integral Estimation Methods

The non-linear fracture mechanics parameter, J , can be partitioned into elastic J_e and plastic J_p terms, written as

$$J = J_e + J_p \quad (2.80)$$

where the elastic term can be calculated using the stress intensity factor, K , and the effective Young's modulus, E' , by

$$J_e = \frac{K^2}{E'} \quad (2.81)$$

The existing methods to estimate the plastic J_p term are described and shown below.

2.6.3.1 Experimental Displacement Method

The plastic J_p term can be experimentally calculated by

$$J_p = \frac{A_p}{B_n(W-a)} \eta \quad (2.82)$$

where B_n is the net thickness between the side grooves, W is the width, a is the crack length, η is a geometry dependent function and A_p is the area corresponding to the plastic displacement under load against displacement curve, as shown in Figure 2.10. It has been shown in [67] that by integrating the area under the curve for a power law hardening material with the stress exponent N , J_p can be written as

$$J_p = \frac{P \Delta_p}{B_n(W-a)} H \eta \quad (2.83)$$

where P is the applied load, Δ_p is the plastic load line displacement and $H = N/(N+1)$ for a C(T) specimen. It has been shown in [67] that for a C(T) specimen, same solutions of J_p can be found by employing load line displacement (LLD) and crack mouth opening displacement (CMOD) along the load line. Also shown in [67] is that for a C(T) specimen with a standard crack length in the range of $0.45 < a/W < 0.7$, $\eta = 2.2$ under both plane stress and plane strain conditions.

2.6.3.2 EPRI Solution

For a power law hardening material, J -integral can be estimated using EPRI solution. Considering a standard C(T) specimen, the plastic component of J -integral, J_p , can be expressed as

$$J_p = \alpha \sigma_{p0} \varepsilon_{p0} c h_1(a/W, N) (P/P_{p0})^{N+1} \quad (2.84)$$

where α , σ_{p0} and ε_{p0} are Ramberg-Osgood material model constants (see Eqn (2.7)), P is the applied load per unit thickness, P_{p0} is the limit load per unit thickness, a is the crack length, W is the specimen width, c is the uncracked ligament, N is the power law hardening exponent and h_1 is a dimensionless function of the normalised crack length and stress exponent. The numerical solutions of h_1 have been tabulated in [61].

The plastic normalising load, P_{p0} , in Eqn (2.84) is given by

$$\text{Plane stress: } P_{p0} = 1.071 \xi c \sigma_{p0} \quad (2.85)$$

$$\text{Plane strain: } P_{p0} = 1.455 \xi c \sigma_{p0} \quad (2.86)$$

where ξ is defined as

$$\xi = \sqrt{\left[(2a/c)^2 + 2(2a/c) + 2 \right]} - \left[(2a/c) + 1 \right] \quad (2.87)$$

An estimate of the total J , which is the sum of elastic and plastic terms, can therefore be obtained by

$$J_{EPRI} = J_e + J_p = \frac{K^2(a_e)}{E'} + \alpha \sigma_{p0} \varepsilon_{p0} c h_1(a/W, N) (P/P_{p0})^{N+1} \quad (2.88)$$

where K is the stress intensity factor, a_e is the effective crack length calculated using Eqn (2.71) and E' is the effective Young's modulus.

2.6.3.3 Reference Stress Approach

Due to the difficulties encountered in estimation of J -integral using EPRI solution, it was proposed in [68] to develop a new technique for approximation of J -integral using the reference stress, σ_{ref} , parameter. The normalised reference stress, $\sigma_{ref}/\sigma_{0.2}$, (which is usually denoted as L_r in the literature) indicates the level of plasticity in a fracture specimen and can be expressed as

$$\frac{\sigma_{ref}}{\sigma_{0.2}} = \frac{P}{P_{LC}} \quad (2.89)$$

where P is the applied load, $\sigma_{0.2}$ is 0.2% proof stress (which is often taken as the yield stress of the material) and P_{LC} is the plastic collapse load in an elastic-perfectly plastic

material. If the normalised reference stress is much larger than unity, it implies that there is widespread plasticity in the specimen. If this ratio is much less than unity, a limited amount of plasticity is expected in the specimen. It must be noted that the normalised reference stress under plane stress conditions is generally larger than plane strain. The plastic collapse load in Eqn (2.89) can be approximated for a C(T) specimen by [69]

$$P_{LC} = \left[\sqrt{\left(1 + \frac{2}{\sqrt{3}}\right) \left(1 + \frac{2}{\sqrt{3}} (a/W)^2\right)} - \left(1 + \frac{2}{\sqrt{3}} (a/W)\right) \right] B_n W \sigma_{0.2} \quad (2.90)$$

$$P_{LC} = \left[\sqrt{2.702 \left(1 + 1.702 (a/W)^2\right)} - (1 + 1.702 (a/W)) \right] \frac{2}{\sqrt{3}} B_n W \sigma_{0.2} \quad (2.91)$$

for plane stress and plane strain conditions, respectively.

By applying the definition of reference stress in EPRI solution (Eqn (2.88)) and using the Ramberg-Osgood material model, a new estimate of J using the reference stress method can be provided by

$$J_{ref} = \frac{K^2(a_e)}{E'} + \mu \sigma_{ref} \left(\varepsilon_{ref} - \frac{\sigma_{ref}}{E} \right) \left(\frac{K(a)}{\sigma_{ref}} \right)^2 \quad (2.92)$$

where ε_{ref} is the total reference strain corresponding to the reference stress, which can be obtained by Ramberg-Osgood material model (Eqn (2.6)). Note that in Eqn (2.92) $\mu = 1$ for plane stress and $\mu = (1-\nu^2)$ for plane strain conditions. It has been shown in [68] that using the reference stress approach, J_{ref} provides an overestimated value of J for a fully plastic material, however under linear elastic conditions J_{ref} is expected to give an accurate solution of J .

2.7 Creep Fracture Mechanics

When a cracked body is subjected to a stress at elevated temperatures, the crack tip fields may be described by fracture mechanics parameter K under small scale yielding or J under fully plastic conditions. However, for a power law creeping body K and J parameters may not characterise the crack tip fields any longer.

When a cracked body is subjected to creep deformation, the creep zone starts to build up at the crack tip. Under small scale creep (SSC) conditions that the creep zone is small compared to the specimen size and crack length, the stress distribution may still be described by fracture mechanics K and J parameters. However, when the cracked body remains under creep deformation conditions for a longer time, the creep zone size starts to increase and consequently stress distribution changes with time. In

this situation that stress and strain fields vary until a steady state condition is reached, transient creep condition is expected to exist at the stationary sharp crack tip. As time proceeds, creep zone continues to enlarge at the crack tip until material reaches steady state conditions at which widespread creep is attained and no further change occurs in the stress fields around the crack tip. Different stages of the creep zone development ahead of the crack tip are schematically shown in Figure 2.11.

2.7.1 Steady State Creep Crack Growth

At long times, where a steady state of creep deformation and damage has developed at a crack tip, the CCG rate, $\dot{a}$ (or da/dt), may be described by the crack tip fracture mechanics parameter C^* using a power law relationship,

$$\dot{a} = DC^{*\phi} \quad (2.93)$$

where D and ϕ are material constants which may be temperature and stress state dependent and can be determined by a regression fit to the experimental data. The $\dot{a}$ and C^* data are often presented in mm/h and MPam/h, respectively. Under steady state conditions, $\dot{a}$ vs. C^* data appear as a straight line on log-log axes [18]. However, prior to the steady state conditions being achieved, the data will appear as a “tail” region.

The C^* creep fracture mechanics parameter, which is analogous to the non-linear fracture mechanism parameter J , can be defined as

$$C^* = \int_{\Gamma} (\dot{W} dy - T_i \frac{\partial \dot{u}_i}{\partial x} ds) \quad (2.94)$$

where Γ is an integration contour around the crack tip, $\dot{u}_i$ is the displacement rate vector and $\dot{W}$ is the strain energy rate density which is given by

$$\dot{W} = \int_0^{\dot{\epsilon}_{ij}^c} \sigma_{ij} d\dot{\epsilon}_{ij}^c \quad (2.95)$$

where $\dot{\epsilon}_{ij}^c$ is the creep strain rate. For a power law creeping material (described by the Norton law Eqn (2.23)), the strain energy rate density can also be written as

$$\dot{W} = \frac{n}{n+1} \sigma_{ij} \dot{\epsilon}_{ij}^c \quad (2.96)$$

where n is the creep stress exponent.

As seen in Eqns (2.94) and (2.95), C^* is a function of strain rate and thus is considered as a time dependent parameter. However, similar to the J -integral parameter shown in Eqn (2.72), C^* is path independent under widespread creep

conditions. Note that prior to steady state creep conditions (under SSC and transient creep conditions), C^* is path and time dependent and is conventionally denoted by $C(t)$.

2.7.2 RR Stress and Strain Rate Distribution Fields

Under widespread steady state creep conditions, the stress and strain rate distribution fields ahead of the crack tip can be characterised by C^* parameter using RR (Riedel and Rice [70]) equations which can be expressed as

$$\frac{\sigma_{ij}}{\sigma_0} = \left[\frac{C^*}{\dot{\epsilon}_0 \sigma_0 I_n r} \right]^{\frac{1}{n+1}} \tilde{\sigma}_{ij}(\theta; n) \quad (2.97)$$

$$\frac{\dot{\epsilon}_{ij}}{\dot{\epsilon}_0} = \left[\frac{C^*}{\dot{\epsilon}_0 \sigma_0 I_n r} \right]^{\frac{n}{n+1}} \tilde{\epsilon}_{ij}(\theta; n) \quad (2.98)$$

where σ_{ij} and $\dot{\epsilon}_{ij}$ are the stress and strain rate tensors, respectively, σ_0 is the normalising stress, $\dot{\epsilon}_0$ is the normalising strain rate and r is the radial distance from the crack tip. The stress and strain angular functions, which have been derived from numerical studies, can be correlated using

$$\tilde{\epsilon}_{ij}(\theta; n) = \left(\tilde{\sigma}_{ij}(\theta; n) \right)^n \quad (2.99)$$

The dimensionless constant I_n in RR equations can be evaluated by

$$\text{Plane stress: } I_n = 7.2 \sqrt{0.12 + \frac{1}{n} - \frac{2.9}{n}} \quad (2.100)$$

$$\text{Plane strain: } I_n = 10.3 \sqrt{0.13 + \frac{1}{n} - \frac{4.6}{n}} \quad (2.101)$$

2.7.3 C^* Estimation Methods

Considering the analogy between plastic and creep deformation, steady state C^* creep fracture mechanics parameter can be estimated for a power law creeping material by similar methods employed to provide approximate solution of J -integral. These methods are explained and discussed in this section.

2.7.3.1 Experimental Displacement Rate Approach

The C^* creep fracture mechanics parameter, can be determined experimentally from the load line displacement rate measurements, $\dot{\Delta}$, using the relation [67]

$$C^*_{exp} = \frac{P\dot{\Delta}}{B_n(W-a)} H\eta \quad (2.102)$$

where H and η are geometry dependent constants. For a C(T) specimen with the normalised crack length in the valid range specified in ASTM E1457 standard [71], $H = n/(n+1)$ (where n is the power law creep stress exponent) and $\eta=2.2$ [67].

The total load line displacement rate, $\dot{\Delta}^T$, in a creep crack growth test can be partitioned into an instantaneous, $\dot{\Delta}^i$, and a time-dependent part which is directly associated with the accumulation of time dependent creep strains, $\dot{\Delta}^c$, such that,

$$\dot{\Delta}^T = \dot{\Delta}^i + \dot{\Delta}^c \quad (2.103)$$

The instantaneous displacement rate, $\dot{\Delta}^i$, can be further divided into an elastic $\dot{\Delta}^i_e$ and a plastic $\dot{\Delta}^i_p$ part, calculated using [71]

$$\dot{\Delta}^i_e = \frac{\dot{a}B_n}{P} \left[\frac{2K^2}{E'} \right] \quad (2.104)$$

$$\dot{\Delta}^i_p = \frac{(N+1)J_p B_n \dot{a}}{P} \quad (2.105)$$

In Eqn (2.104) $\dot{a}$ is the crack growth rate, B_n is the net thickness between the side grooves, K is the stress intensity factor, P is the applied load and E' is the effective elastic modulus (which is equal to $E/(1-\nu^2)$ for plane strain and E for plane stress conditions). In Eqn (2.105) N is the power law stress exponent and J_p is the plastic component of the J -Integral.

According to the ASTM E1457 standard [71], the plastic term of the displacement rate is generally small and thus can be neglected in creep displacement rate calculations. Furthermore, where the crack growth is pre-dominantly due to creep, the elastic part of the instantaneous load line displacement rate $\dot{\Delta}^i_e$ is often found to be small compared to $\dot{\Delta}^c$ and hence usually neglected when calculating C^* . Therefore the total displacement rate is usually employed in C^* calculations using Eqn (2.102).

2.7.3.2 EPRI Solution

The EPRI solution, which was used to evaluate J -integral for a power law hardening material (see Eqn (2.84)), can be applied to estimate C^* for a power law creeping material. Since creep deformation is a time dependent process, ε_{p0} , σ_{p0} and P_{p0} in Eqn (2.84) have to be replaced by $\dot{\varepsilon}_0$, σ_0 and P_0 , respectively. Also the plastic hardening exponent N in Eqn (2.84) must be replaced by the creep stress exponent n . Therefore, the EPRI solution of the C^* parameter can be written as

$$C^*_{EPRI} = \sigma_0 \dot{\epsilon}_0 c h_1 (a/W, n) (P/P_0)^{n+1} \quad (2.106)$$

2.7.3.3 Reference Stress Method

An alternative method to estimate C^* parameter is to apply reference stress approach, which was previously employed to estimate J -integral in Eqn (2.92), in conjunction with the creep strain rate. By implementing the reference stress method, C^* can be estimated by

$$C^*_{ref} = \sigma_{ref} \dot{\epsilon}_{ref}^c \left(\frac{K(a)}{\sigma_{ref}} \right)^2 \quad (2.107)$$

where K is the stress intensity factor and $\dot{\epsilon}_{ref}^c$ is the reference creep strain rate corresponding to the reference stress σ_{ref} (i.e. $\dot{\epsilon}_{ref}^c = A \sigma_{ref}^n$).

2.7.4 Non-Steady State Creep Crack Growth

The C^* creep fracture mechanics parameter may characterise the CCG behaviour of a material under widespread steady state creep conditions. However, a model is required to predict creep crack growth behaviour during transient creep, where creep damage is building up from SSC to a steady state condition.

The creep process zone, r_c , can be divided to small elements with the same width of dr . The assumption has been made that creep stress distribution occurs under steady state conditions and therefore the fracture mechanics parameter C^* can characterise the crack tip field using RR equations. The elements width, dr , are considered to be small enough to have constant stress distribution through the elements. The creep damage starts to build up and the crack grows discretely through elements in sequence. Therefore when one element reaches a critical value of damage, it fails and the creep damage accumulates in the next element. This process continues until the whole specimen fails.

It has been shown in [18] that the initial CCG rate, $\dot{a}_0$, which may happen at the time that damage is building up in the first element, can be related to the steady state crack growth rate $\dot{a}_s$ by

$$\dot{a}_0 = \frac{1}{n+1} \left(\frac{dr}{r_c} \right)^{1/(n+1)} \dot{a}_s \quad (2.108)$$

Knowing that the power law creep stress exponent n is much larger than unity, an approximate solution of the initial CCG rate can be expressed as

$$\dot{a}_0 \approx \frac{\dot{a}_s}{n+1} \quad (2.109)$$

This prediction is based on the assumption that stress redistribution has reached the steady state conditions before crack starts to grow.

Considering the average creep strain rate properties, another prediction of the initial crack growth rate may be made using [18]

$$\dot{a}_0 \approx \frac{n_A + 1 - v_r}{n_A + 1} \dot{a}_s \quad (2.110)$$

where n_A and v_r are the average creep strain rate and rupture time power law stress exponents, respectively. As seen in Eqn (2.110), when $n_A = v_r$ the same estimates of the initial CCG rate can be provided by Eqns (2.109) and (2.110) assuming that $n_A = n$.

Prior to steady state creep conditions, the crack tip fields may be described by the time and path dependent $C(t)$ fracture mechanics parameter which can be defined as

$$C(t) = \frac{K^2 / E'}{(n+1)t} \quad (2.111)$$

$C(t)$ can be related to the steady state C^* parameter using

$$C(t) = C^* \left[\frac{(1 + t / \tau)^{n+1}}{(1 + t / \tau)^{n+1} - 1} \right] \quad (2.112)$$

where τ is the redistribution time which is defined as the time required for the widespread steady state creep conditions to be attained at the crack tip (thus the stress distribution fields to be defined by RR equations) and can be estimated by

$$\tau = \frac{J}{C^*} \quad (2.113)$$

As seen in Eqn (2.112), $C(t)$ tends to C^* when time t is infinitely long.

For situations where the crack tip deformation is predominantly elastic (*e.g.* short times and/or high crack velocity) it is expected that crack growth will be controlled by the linear elastic stress intensity factor, K [18]. Under such conditions, the CCG rate may be estimated by

$$\dot{a} = D' K^m \quad (2.114)$$

where D' and m are material constants.

2.8 NSW CCG and CCI Prediction Models

Under steady state creep conditions, the creep crack initiation and growth behaviour of a power law creeping material can be described by NSW (after Nikbin, Smith,

Webster) prediction models which are based on the experimental uniaxial creep rupture properties [72-74]. Different versions of NSW models are presented and discussed in this section.

2.8.1 NSW CCG Models

As explained previously, under steady state creep conditions a power law relation can be inferred when the experimental CCG rate, $\dot{a}$, is plotted against the C^* creep fracture mechanics parameter (see Eqn (2.93)). The NSW models have been proposed in [72-74] to predict the steady state CCG rate from the uniaxial creep data. In NSW models, the creep crack growth is predicted to occur when a critical quantity of damage, measured by creep ductility exhaustion approach, is attained at a characteristic distance, r_c , ahead of the crack tip. The CCG rate using NSW model, $\dot{a}_{NSW}$, can be predicted by

$$\dot{a}_{NSW} = (n+1) \frac{\dot{\epsilon}_0}{\epsilon_f^*} \left(\frac{C^*}{\dot{\epsilon}_0 \sigma_0 I_n} \right)^{\frac{n}{n+1}} (r_c)^{\frac{1}{n+1}} \quad (2.115)$$

or

$$\dot{a}_{NSW} = \frac{(n+1)}{\epsilon_f^*} (Ar_c)^{\frac{1}{n+1}} \left[\frac{C^*}{I_n} \right]^{\frac{n}{n+1}} \quad (2.116)$$

where $\dot{\epsilon}_0$ and σ_0 are material constants, n is the power law creep stress exponent, r_c is the creep process zone (which is usually taken as the material grain size) and the multiaxial creep ductility, ϵ_f^* , is taken as the uniaxial creep ductility ϵ_f for plane stress conditions and $\epsilon_f/30$ under plane strain conditions [75]. The dimensionless constant I_n in Eqns (2.115) and (2.116) can be evaluated by Eqn (2.100) and (2.101) for plane stress and plane strain conditions, respectively. Note that in NSW equations $\dot{\epsilon}_0$ is often taken as 1 h^{-1} and the predicted CCG rate is found relatively insensitive to r_c due to the small fraction power.

A modified version of the NSW model, known as NSW-MOD, has been derived in [76] to predict the CCG rate under steady state creep conditions by considering the dependency of creep strain on the crack tip angle, θ , and the power law creep stress exponent, n , in addition to the stress state. The CCG rate using NSW-MOD model, $\dot{a}_{NSW-MOD}$, is predicted by

$$\dot{a}_{NSW-MOD} = (n+1) \left(Ar_c \right)^{\frac{1}{n+1}} \left[\frac{C^*}{I_n} \right]^{\frac{n}{n+1}} \frac{\tilde{\varepsilon}_e(\theta, n)}{\varepsilon_f^*(\theta, n)} \bigg|_{max} \quad (2.117)$$

As seen in Eqn (2.117) the crack growth is predicted using the maximum value of the angular function $\tilde{\varepsilon}_e / \varepsilon_f^*$, where $\tilde{\varepsilon}_e$ is a non-dimensional function of θ and n , values of which are tabulated in [66]. In Eqn (2.117), the multiaxial creep ductility, ε_f^* , for a range of stress triaxialities may be estimated from the uniaxial creep strain at failure, ε_f , using an appropriate model, such as the Cocks and Ashby void growth and coalescence model [44] (see Eqn (2.53)). In NSW model calculations, the characteristic distance r_c is generally taken as the grain size of the material, thus can be considered 0.05 mm for 316H stainless steel [50].

It has been suggested in [73] that for a range of steels, the CCG rate is most sensitive to the multiaxial creep ductility ε_f^* . Hence the steady state CCG rate may be reasonably approximated by

$$\dot{a}_s^{NSWA} = \frac{3C^{*0.85}}{\varepsilon_f^*} \quad (2.118)$$

where $\dot{a}_s^{NSWA}$ and C^* have the units mm/h and MPam/h, respectively. Similar to the original NSW model, the multiaxial creep ductility ε_f^* in Eqn (2.118) is taken as the uniaxial creep strain at failure, ε_f , for plane stress and $\varepsilon_f/30$ for plane strain conditions [75]. Eqn (2.118) is often referred to as the approximate NSW (NSWA) model.

2.8.2 NSW CCI Models

In the static creep crack growth tests, a substantial amount of time is often required for creep strain/damage to build up and stress redistribution to take place ahead of the crack tip prior to crack extension from a pre-existing defect. This period of time is usually referred to as creep crack initiation (CCI) or incubation time, t_i . CCI is generally defined as the period of time for some small measureable amount of crack extension to occur. The initiation time in CCG tests is usually reported for 0.2 or 0.5 mm of crack extension, Δa , and denoted as $t_{0.2}$ and $t_{0.5}$, respectively [71]. Assuming that CCI occurs under predominant steady state creep conditions and C^* is the controlling crack tip field parameter, t_i can be correlated with the fracture mechanics parameter C^* for creep ductile materials using a power law relation

$$t_i = \frac{\Delta a}{\dot{a}_s} = \frac{\Delta a}{DC^*\phi} \quad (2.119)$$

where D and ϕ are the material constants defined in Eqn (2.93).

The CCI time may be estimated by assuming that the crack grows from the point of initial loading at a constant rate. The lower and upper bound estimates of the CCI time may be obtained using the steady state CCG rate, $\dot{a}_s$, and the initial CCG rate (CCG rate at the initial crack length, a_0), $\dot{a}_0$, respectively. As shown in Eqn (2.109) the initial CCG rate, $\dot{a}_0$, for a given value of C^* may be estimated to be $(n+1)$ times less than the steady state value of CCG rate [18]. Hence the CCI time is estimated by

$$\frac{\Delta a}{\dot{a}_s} \leq t_i \leq \frac{\Delta a}{\dot{a}_s} (n+1) \quad (2.120)$$

Using the steady state CCG rate predictions made by NSW-MOD model, the CCI time may be estimated as

$$\frac{\Delta a}{(n+1)(Ar_c)^{\frac{1}{n+1}} \left(\frac{I_n}{C^*} \right)^{\frac{n}{n+1}} \frac{\varepsilon_f^*}{\tilde{\varepsilon}_e} \Big|_{max}} \leq t_i \leq \frac{\Delta a}{(Ar_c)^{\frac{1}{n+1}} \left(\frac{I_n}{C^*} \right)^{\frac{n}{n+1}} \frac{\varepsilon_f^*}{\tilde{\varepsilon}_e} \Big|_{max}} \quad (2.121)$$

Assuming that $\tilde{\varepsilon}_e \Big|_{max} = 1$ in Eqn (2.121), the upper and lower bound CCI predictions can be obtained by the original NSW model.

Combining Eqns (2.118) and (2.120), the CCI time estimated by the NSW model can be defined as

$$\frac{\Delta a \varepsilon_f^*}{3C^{*0.85}} \leq t_i \leq \frac{\Delta a (n+1) \varepsilon_f^*}{3C^{*0.85}} \quad (2.122)$$

2.9 Fatigue

In engineering applications, fatigue is known as an accumulative damage process which can occur in a component subjected to cyclic loading conditions. The maximum stress during fatigue deformation is often less than the ultimate tensile strength (UTS) of the material but may be either below or above the yield stress. Most of high temperature industrial components are subjected to cyclic loading during operation (*i.e.* due to start ups and shut downs). In the early 1960s, it was shown by Paris that fatigue crack growth can be characterised by fracture mechanics parameters [77, 78]. Since then, fracture mechanics have been frequently applied to describe fatigue deformation in materials. Fundamental concepts of fatigue deformation and crack growth are summarised and described in this section.

2.9.1 Introduction to Fatigue

The failure of industrial components subjected to high stress levels at elevated temperatures is often due to creep, fatigue, environmental processes or interaction between these mechanisms. Under service conditions, where more than one mechanism may contribute to a component's failure, the dominant failure mode can be identified by investigating material composition, heat treatment, cyclic to mean load ratio, frequency, temperature and the operating environment [18].

Laboratory fatigue deformation tests are usually conducted on uncracked round bar specimens whereas fatigue crack growth behaviour of a material is conventionally investigated by performing tests on C(T) specimens at a constant temperature. Note that at higher temperatures creep deformation is expected to have contributions to specimen's failure. To study the fatigue deformation behaviour of a material alone, in the absence of creep deformation, fatigue tests can be performed either at room temperature where creep is not the dominant failure mechanism or at high temperatures with no hold time considered at maximum and minimum load levels.

2.9.2 Fatigue Parameters

Fatigue failure is mainly favoured by high tensile stress, large difference between maximum and minimum stress levels and large number of stress cycles. Other factors which may influence fatigue failure are residual stresses, coating, corrosion and surface hardness. In all fatigue tests, the specimen is subjected to a maximum load, σ_{max} , followed by a minimum load, σ_{min} , in each cycle. For a specimen subjected to cyclic loading conditions, fatigue parameters can be defined and expressed as

$$\Delta\sigma = \sigma_{max} - \sigma_{min} \quad (2.123)$$

$$\sigma_{mean} = \frac{\sigma_{max} + \sigma_{min}}{2} \quad (2.124)$$

$$\sigma_a = \frac{\sigma_{max} - \sigma_{min}}{2} \quad (2.125)$$

where $\Delta\sigma$ is the stress range, σ_{mean} is the mean (average) stress and σ_a is the stress amplitude. In order to describe a fatigue cycle, the stress ratio, R , and amplitude ratio, A , can be defined as

$$R = \frac{\sigma_{min}}{\sigma_{max}} = \frac{P_{min}}{P_{max}} = \frac{K_{min}}{K_{max}} \quad (2.126)$$

$$A = \frac{\sigma_a}{\sigma_{mean}} = \frac{1-R}{1+R} \quad (2.127)$$

where P_{min} and K_{min} are the load and stress intensity factor corresponding to the minimum stress σ_{min} . Similarly P_{max} and K_{max} are the corresponding load and stress intensity factor for the maximum stress σ_{max} . From the definition of R ratio in Eqn (2.126) it can be realised that for tensile-compressive fatigue cycles R -ratio is negative. However, for tensile-tensile cycles the R -ratio is always positive. Eqn (2.126) shows that $R = 0$ when the minimum stress level is $\sigma_{min} = 0$. Note that for tensile-compressive cycles where the magnitude of σ_{min} and σ_{max} are the same, $R = -1$ and $\sigma_{mean} = 0$. A cycle with such loading conditions is known as fully reversed cycle.

2.9.3 Fatigue Damage

Most engineering components experience various types of loading cycles. Under dominant fatigue conditions, damage is initiated and accumulated during each cycle. Individual cyclic damage is not sufficient alone to cause failure, however the cumulative fatigue damage may eventually lead to fracture in a component. Miner's cumulative damage law [79] suggests to calculate the fatigue damage, D_f , in multiple types of cycles by

$$D_f = \sum \frac{N_i}{N_{fi}} \quad (2.128)$$

where N_i is the number of cycles of type i and N_{fi} is the number of cycles to failure in cycles of type i . According to Miner's rule, a component fails when the sum of fractional fatigue damages reaches unity [18].

2.9.4 Fatigue Behaviour of Cracked Components

In large industrial components, in particular welded structures such as pressure vessels, the possibility of observing a defect in form of a crack is fairly high. A key point in the component life assessments is to measure the initial crack length and estimate the remaining life time of the structure during which the material can sustain the applied loads prior to reaching a catastrophic crack length [80]. Fatigue deformation and crack growth is a common failure mechanism in industrial components. Hence, it is important to predict the life time of cracked geometries deforming under fatigue conditions, in industrial applications.

Fatigue crack growth data are usually plotted in the form of crack propagation rate per cycle, da/dN , against the change in stress intensity factor, ΔK , in log-log axes. Following the definition of K in Eqn (2.64), for a given crack length the change in stress intensity factor due to cyclic loading can be defined as

$$\Delta K = K_{max} - K_{min} = \Delta \sigma Y \sqrt{a} \quad (2.129)$$

where $\Delta \sigma = (\sigma_{max} - \sigma_{min})$. A schematic illustration of the fatigue crack growth curve in log scale is shown in Figure 2.12. As seen in this figure, there are generally three regions observed in a fatigue crack growth trend of a material:

- Region I (threshold): The crack propagation rate, da/dN , decelerates rapidly as ΔK increases. A threshold value of ΔK , denoted ΔK_{th} , can be observed in this region below which insignificant or no fatigue crack propagation can be detected in a test.
- Region II: A power law relationship between da/dN and ΔK can be inferred in this region.
- Region III: The crack growth rate increases with increasing ΔK . The crack propagation in this region is unstable and in the case of brittle material the brittle fracture may occur when K_{max} approaches the fracture toughness of the material, K_c .

2.9.4.1 Empirical Fatigue Crack Growth Equations

As shown in Figure 2.12, the fatigue crack growth behaviour becomes steady at intermediate values of ΔK . For an isotropic material, the crack growth rate per cycle da/dN in the steady state regime (Region II) can be defined by the power law Paris equation [77], as

$$\frac{da}{dN} = C \Delta K^m \quad (2.130)$$

where C and m are empirical material constants which can be found by a regression fit to the test data. It has been shown in recent studies that constant m ranges between $2 < m < 7$ for various materials [59]. As seen in Eqn (2.130), the formulation of crack growth rate per cycle in Paris law is insensitive to the R -ratio. However, another equation has been developed by Forman [81] to relate da/dN to R -ratio in fatigue regions II and III by

$$\frac{da}{dN} = \frac{C \Delta K^m}{(1-R)K_c - \Delta K} \quad (2.131)$$

Forman's equation can also be written as

$$\frac{da}{dN} = \frac{C \Delta K^{m-1}}{K_c / K_{max} - 1} \quad (2.132)$$

This equation shows that the FCG rate per cycle tends to infinity as K_{max} approaches K_c . Note that the values and units of material constants C and m in Forman's equation are different from those of obtained from Paris law in Eqn (2.130). Another semi-empirical equation to describe crack growth behaviour in fatigue regions II and III was proposed by Weertman [82] as

$$\frac{da}{dN} = \frac{C \Delta K^4}{K_c^2 - K_{max}^2} \quad (2.133)$$

Note that Eqns (2.130)–(2.133) are empirical relations. However, a number of researchers have employed physical models to develop equations showing fatigue crack growth behaviour of different materials. An example of a physical-based formulation to describe da/dN in all three fatigue regions is McEvily's relationship [83] which can be defined as

$$\frac{da}{dN} = C(\Delta K - \Delta K_{th})^2 \left[1 + \frac{\Delta K}{K_c - K_{max}} \right] \quad (2.134)$$

2.9.4.2 Fatigue Life Estimation

The FCG equations explained above can be integrated to evaluate the fatigue life for a given material and test conditions. Paris law is the main empirical equation often used to estimate fatigue life of a component in the steady state regime. Combining Eqns (2.129) and (2.130) Paris law can be rewritten as

$$\frac{da}{dN} = C(\Delta \sigma Y \sqrt{a})^m \quad (2.135)$$

Eqn (2.135) can be integrated

$$\frac{1}{C(\Delta \sigma Y)^m} \int_{a_0}^a a^{-m/2} da = \int_0^N dN \quad (2.136)$$

to evaluate the number of cycles required to reach the crack extension of $\Delta a = (a - a_0)$ in the steady state secondary fatigue region by

$$N = \frac{2}{(2-m)} \frac{1}{C(\Delta \sigma Y)^m} \left[a^{1-m/2} - a_0^{1-m/2} \right] \quad (2.137)$$

where a_0 is the initial crack length. Eqn (2.137) may describe the fatigue behaviour in the steady state power law region, though by employing the final crack length (which

corresponds to the third fatigue region) in this equation the number of cycles to failure N_f may be approximated using a linear extrapolation.

2.10 Fracture Toughness

Fracture toughness is a material property which describes the resistance of a cracked body against failure. This material property can be identified by the stress intensity factor K (using Eqn (2.64)) for brittle materials or J (using Eqns (2.80)–(2.82)) for ductile materials. The study of the fracture toughness of a material is substantially crucial for design applications.

The majority of failures in cracked components take place under Mode "I" where the crack opening occurs by applying a stress normal to the crack plane. Therefore, the critical value of stress intensity factor K (for brittle materials) or J (for ductile materials) under Mode "I" loading conditions is designated as K_{IC} or J_{IC} , respectively, and is known as the fracture toughness. 316H stainless steel is a ductile material, hence its fracture toughness can be described by J_{IC} .

In ductile materials, three stages are often observed prior to the final fracture [59]:

- i. Formation of a free surface at a particle/inclusion and nucleation of voids
- ii. Void growth due to plastic strain and hydrostatic stress
- iii. Void coalescence and formation of the micro-cracks

In service exposed ductile metals, pre-existing voids can usually be observed in the material. Therefore the failure of a component is mainly controlled by void growth and coalescence stages. There are various models to calculate the void growth and coalescence in ductile materials. Rice and Tracey model [39] is the one which is commonly used to assess the void growth rates in ductile materials.

The material failure may occur by cleavage fracture or ductile crack growth. A brittle material fails immediately due to cleavage however the failure in a ductile material occurs with a stable crack extension process. At present, various standard test methods are available for measuring fracture toughness of materials by performing tests on fracture mechanics specimens. The ones which are most widely used are the ASTM standard E1820 [84] and ESIS P2-92 [85]. The details of the testing and analyses procedures employed in fracture toughness tests are given in Chapter 4.

2.11 Figures

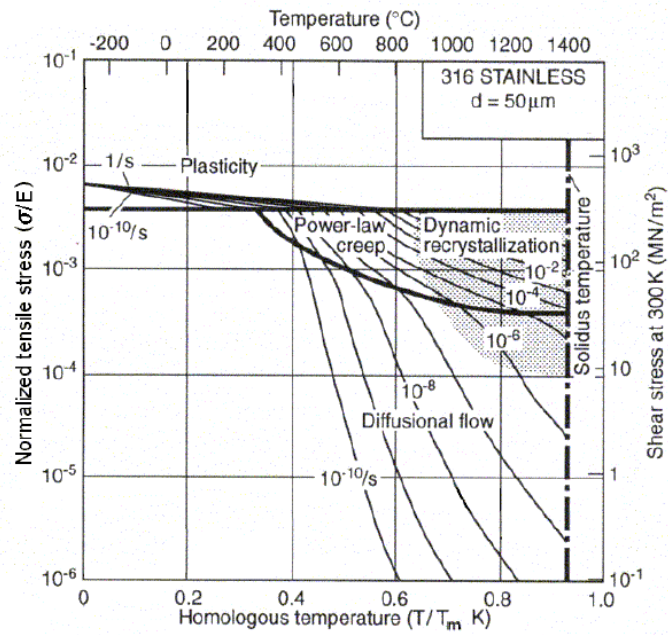


Figure 2.1: Deformation mechanism map for 316 stainless steel (taken from [86])

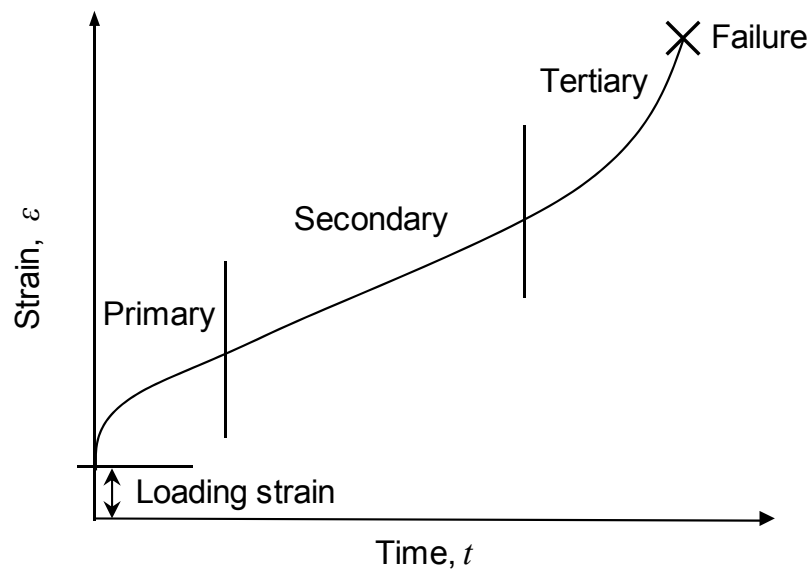


Figure 2.2: Schematic illustration of a uniaxial creep curve

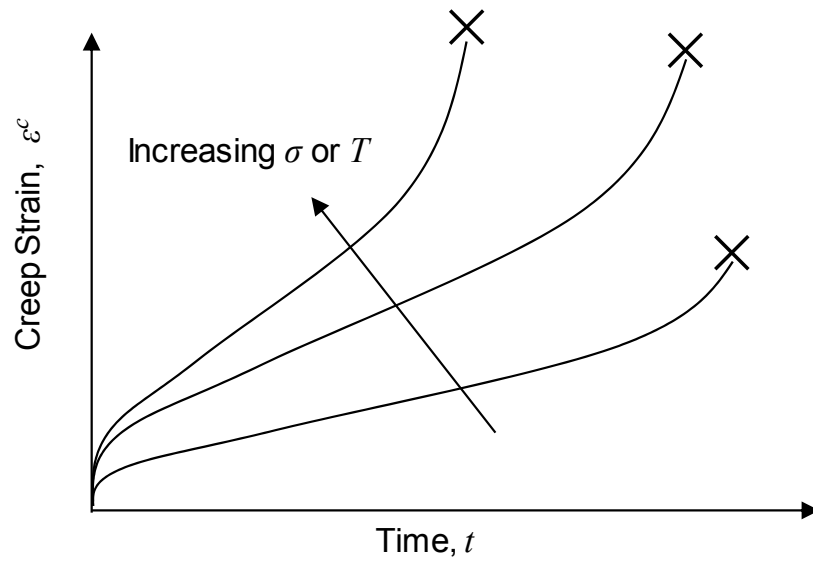


Figure 2.3: Schematic illustration of the creep dependency on stress and temperature

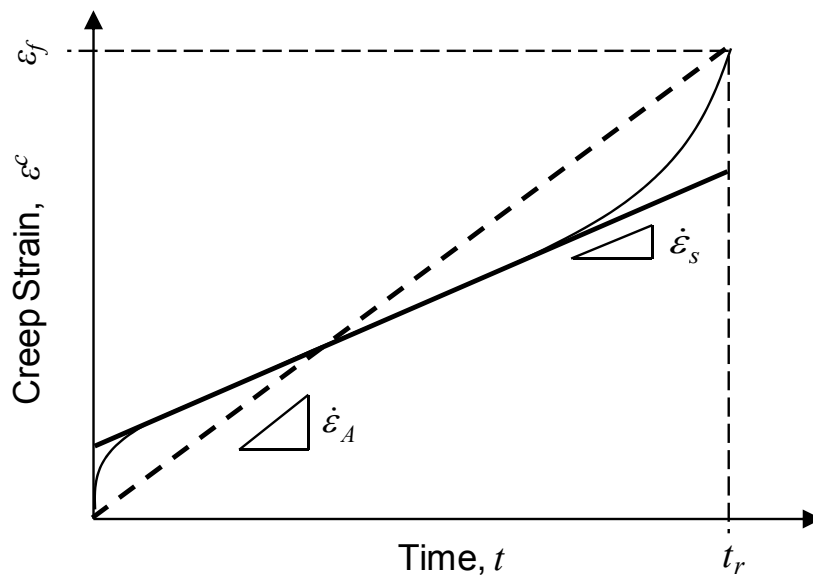


Figure 2.4: Schematic illustration of the average and minimum creep strain rate

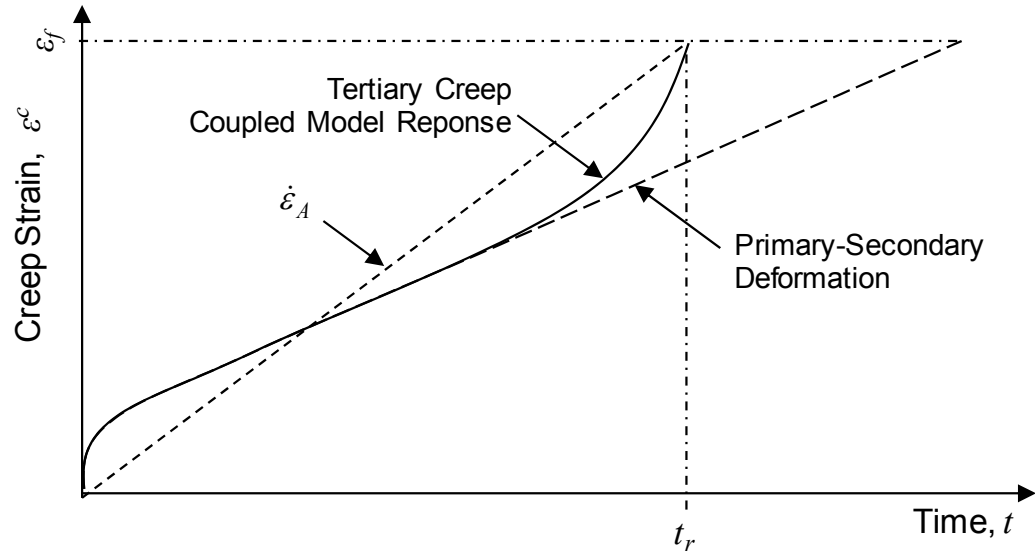


Figure 2.5: Coupled and uncoupled damage models

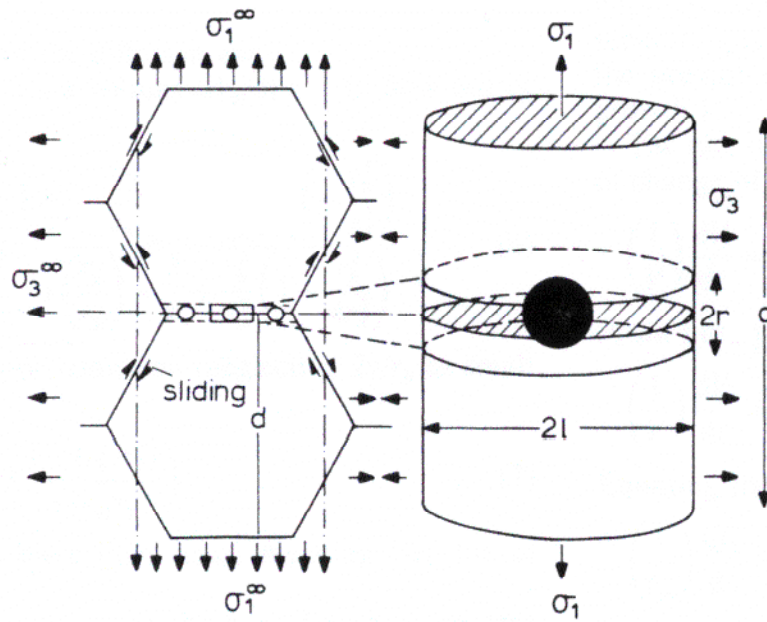


Figure 2.6: Spherical void growth in Cocks and Ashby model (taken from [44])

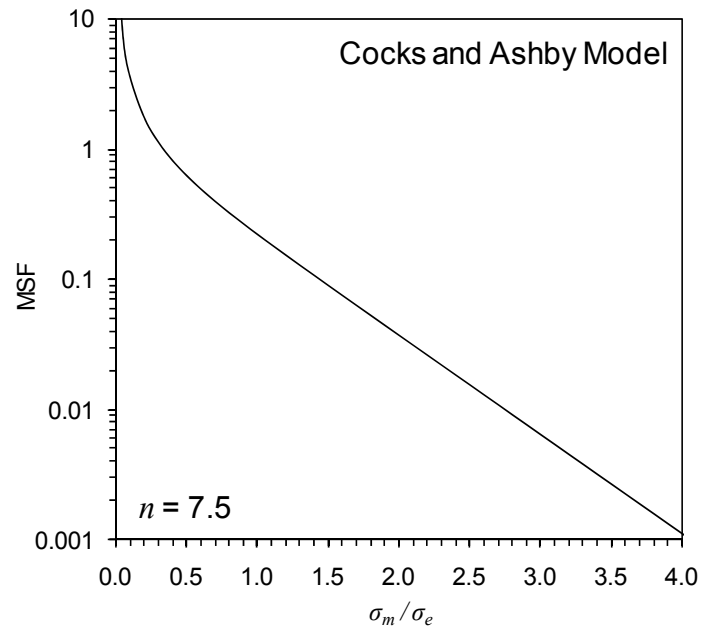


Figure 2.7: Variation of MSF against triaxiality calculated using Cocks and Ashby model

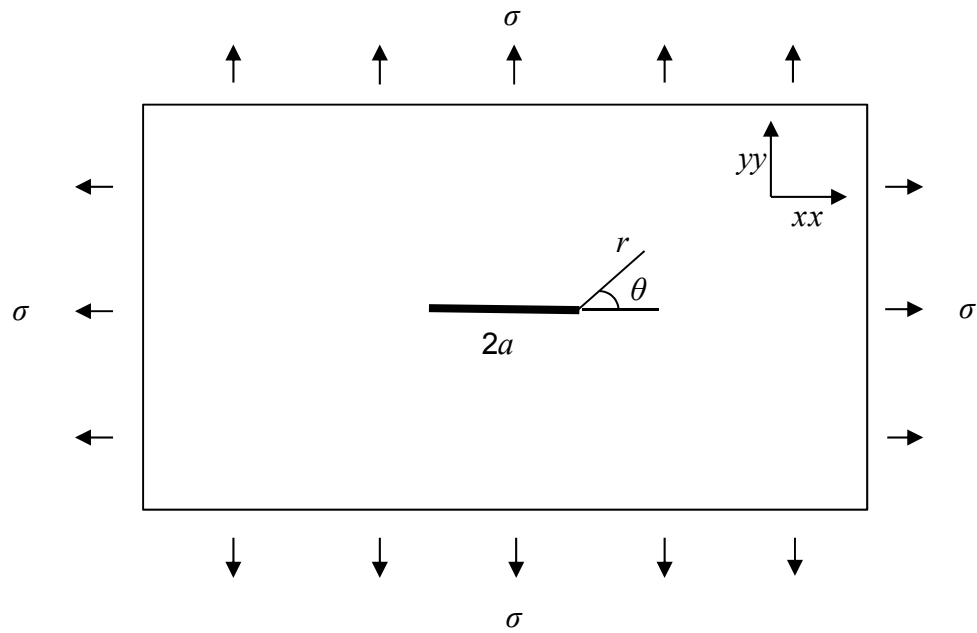


Figure 2.8: Infinite plate subjected to a uniform state of biaxial stress

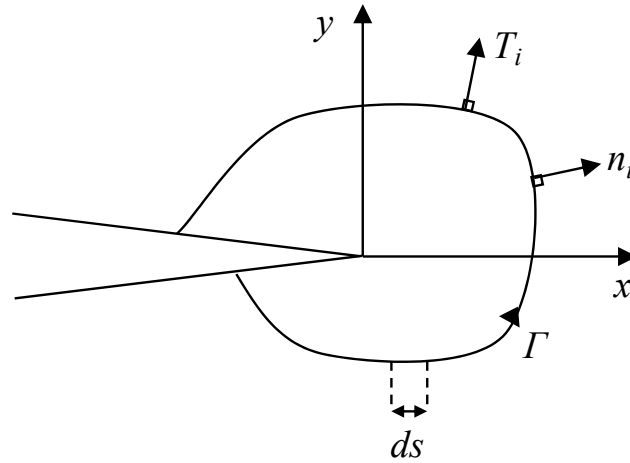


Figure 2.9: Integration contour around a crack tip

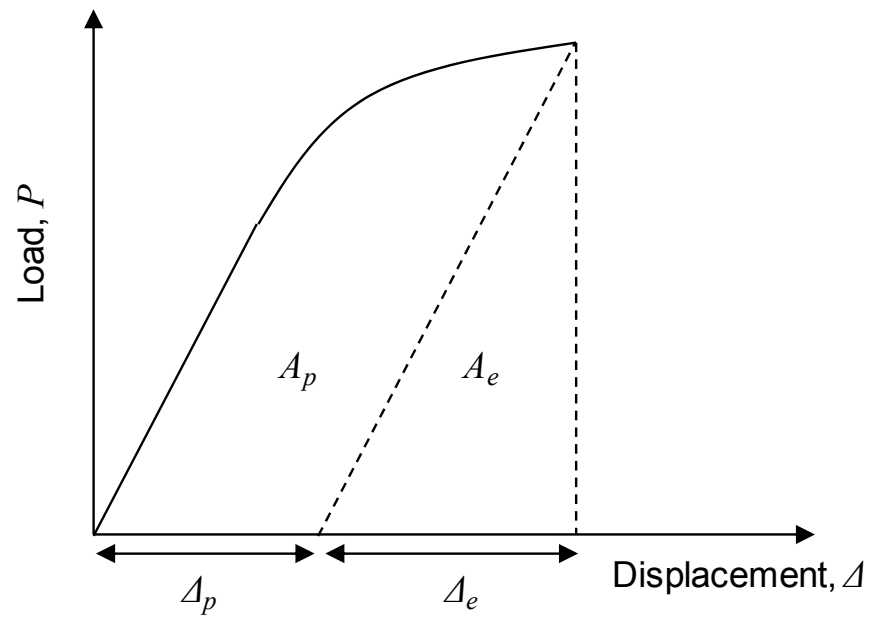


Figure 2.10: Illustration of the elastic and plastic area under load vs. displacement curve

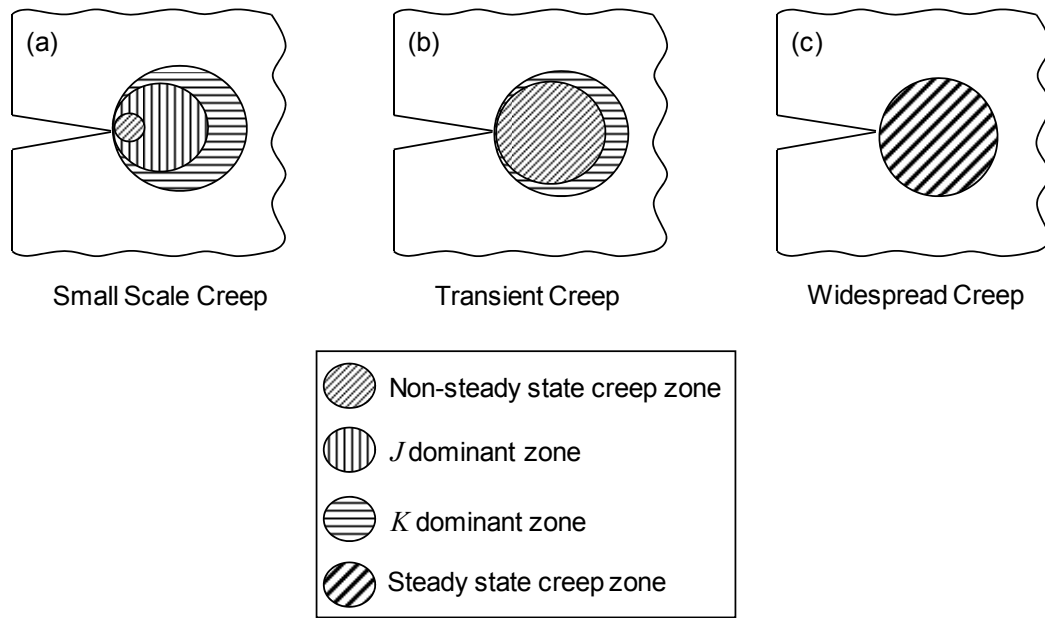


Figure 2.11: Different stages of creep zone development ahead of the crack tip
(a) small scale creep, (b) transient creep (c) steady state creep conditions

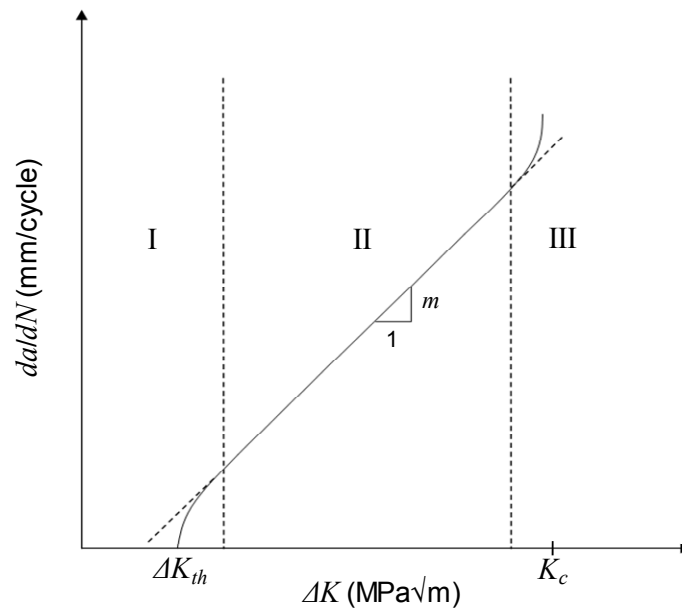


Figure 2.12: Fatigue crack propagation curve

Chapter 3

Materials and Pre-Conditioning

3.1 Pre-Conditioning Effects on Creep, Fatigue and Fracture Toughness

3.1.1 Pre-Straining Effects

Material pre-straining may occur during component fabrication processes. The pre-straining of 316H steel can alter the material's grain size, as discussed in [87]. Components can also be cold worked prior to welding, leading to a significant variation in the material's mechanical properties [88]. Warm rolling of 316H plates, which introduces pre-strain, was found to reduce creep strain rate and ductility compared to material in its annealed condition [14]. Also shown in [14] is that the yield stress of the material increases as a result of pre-straining. Subjecting 316H steel to a uniform compressive plastic pre-strain of 8% at room temperature has however been found to increase the creep strain rate and reduce the creep ductility [1, 89]. The influence of prior plastic pre-straining, introduced in tension, on the subsequent creep behaviour has been examined in [90-92] and smaller creep ductilities are found in the pre-conditioned material. Larger reduction in creep strain at failure has been found in [90] by increasing the level of prior plastic pre-strain. Previous studies on the effect of prior pre-straining on the subsequent creep deformation of 316 stainless steel have also shown that in the materials which are pre-compressed to a percentage of plastic strain smaller than the loading strain on uncompressed material, the plastic strain during load up is reduced by an extent of approximately equal to the compressive plastic pre-straining [93]. More investigations on the prior pre-straining effects on subsequent creep behaviour of stainless steels can be found in [93-98].

The influence of pre-straining (introduced by cold rolling or pre-tension) on the room temperature fracture toughness of an austenitic type stainless steel is investigated in [99]. It has been shown in [99] that the fracture toughness of the austenitic stainless steel continuously decreases as the level of pre-straining increases from 0% to 15%. A similar reduction in the fracture toughness of the pre-strained 316 material is also reported in [100].

No data were found available in the literature for the pre-straining effects on the fatigue crack growth behaviour of austenitic stainless steels. Though it has been shown in [101] that pre-straining effects are small in 304L stainless steel when fatigue tests are conducted under strain controlled mode, whereas in load control small effects were found in low cycle fatigue but extended lives were obtained in high cycle fatigue tests on the pre-strained material. Also shown in [102] is that at high strain amplitudes the fatigue deformation behaviour of the cold worked 316L(N) stainless steel is similar to the material with the as-received condition, however longer fatigue lives were found in the cold worked material at lower strain amplitudes.

3.1.2 Creep Damage Effects

The influence of service exposure in the form of global creep damage has been examined on the subsequent fatigue crack initiation resistance of CrMoV in [103] and it has been observed that the applied stress, the testing temperature, and the creep ductility play important roles in the fatigue deformation behaviour of this material. It has been shown in [103] that prior global creep damage generally has detrimental effects on the fatigue initiation resistance of the material.

The influence of prior global creep damage on the subsequent tensile and fracture properties of 316 stainless steel has been studied in [104]. In [104] the material has been crept at 750 °C under 103 MPa and it has been found that a moderate increase in the yield stress, severe reduction in tensile strain at failure and a rapid drop in the fracture energy (found by Charpy tests) are exhibited in the material after introduction of global creep damage. The increase in the yield stress of the crept material may be attributed to the increase in dislocation density by subjecting the material to creep deformation conditions at elevated temperatures. Further global creep damage experiments have been performed on 316H at 550 °C under 300 MPa in [105], where the tests on round bar specimens were interrupted at different regions of the creep curve (*i.e.* primary, secondary or tertiary creep regions) and subsequent tensile tests were conducted on the crept material under different strain rates. It has been shown in [105] that the tensile strain at failure of a crept material depends on the strain rate and the amount of creep damage introduced into the material (which has been quantified by the life fraction rule in [105]).

The influence of creep damage on the fatigue and fracture toughness behaviour of pre-compressed 316H stainless steel is to be investigated in this research. All creep damage tests are to be performed at 550 °C, which is the typical operating temperature of the material under study.

3.2 Ex-Service Component Details

3.2.1 Material Acquisition

The material utilised in this research is Type 316H stainless steel, provided by EDF Energy, taken from a service exposed steam header (Header 1B1/1, Cast 53415 taken from Heysham 1 station). All specimens tested in this project have been extracted from this header, which is referred to as Header “A” from hereafter. The details and dimensions of Header A are shown in Figure 3.1. As seen in this figure the cylindrical part of the header (denoted section “C”), which had the internal and external diameters of 300 mm and 430 mm, respectively, was separated from the welded dome piece. This section was further divided into three pieces denoted section C1, C2 and C3. As seen in Figure 3.1, a nozzle welded to the cylindrical part of the header was located in section C1. Thus section C2 and C3, which were far from the weld region, were employed in this research to extract specimens from the as-received material.

3.2.2 Service Conditions and Chemical Composition

Header A was under service for 87790 hours at the operating temperature of 523 °C. The header component was heat treated, after removal, for three hours at 1050 °C and then water quenched. The chemical composition (by % weight) of Header A is given in Table 3.1.

3.3 Pre-Compression

As discussed in Section 3.1, pre-straining of 316H to 8% compressive plastic strain has been found to reduce creep ductilities and thus accelerating the creep crack growth behaviour of the material. Therefore, in order to perform creep crack growth tests in the allocated time scale for this research and also to reduce level of plasticity in creep deformation and crack growth tests, the material has been uniformly pre-compressed to 8% plastic true strain at room temperature prior to specimen manufacturing.

Blocks of approximately $63 \times 26 \text{ mm}^2$ in cross-section, taken from Header A, were uniformly pre-compressed to 8% plastic true strain at room temperature and a number of uniaxial tensile specimens or a single compact tension, C(T), specimen was manufactured from them. The pre-compression was performed along the axial direction of the header component. Specimens with the loading direction parallel to the pre-compression axis have been extracted from the pre-strained blocks.

3.4 Available AR and PC Creep Data on 316H

Uniaxial creep data on the as-received and pre-strained material to 8% compressive plastic strain are available at 550 °C in [1] for specimens extracted from Header 1C2/3, called Header B from hereafter, and also Header 2D2/2, called Header C from hereafter. These two headers are both from Cast 55882. The uniaxial creep data from the tests performed on Header A material are compared to those of previously available from Header B and C in Chapter 5.

Creep crack growth data on 316H stainless steel at 550 °C are available for the 8% PC material extracted from Header 1D2/2, called Header D from hereafter, long term AR data from Header C and short term AR data from Header 1A2/1, called Header E from hereafter [24, 89]. The creep crack growth behaviour of the PC material extracted from Header A is compared to those of previously performed on the AR and PC materials extracted from Header C, D and E in Chapter 6. Headers C, D and E are all from Cast 55882. Further comparisons have also been made in Chapter 6 to a short term CCG data set obtained from different casts and headers provided by EDF Energy.

3.5 Introduction of Creep Damage into the Material

After compressively pre-straining the material, creep damage will be introduced prior to fracture toughness and fatigue crack growth testing. For comparison purposes, fracture toughness and fatigue tests will be repeated on uncompressed material and also the pre-compressed material with no creep damage. There are generally two approaches to replicate creep damage of a service exposed component in laboratory tests. These techniques are known as local and global creep damage approaches which are described next.

3.5.1 Local Creep Damage (LCD)

A practical way to introduce creep damage into the material in a laboratory environment as required for this research, is to manufacture fracture specimens from the pre-strained material and perform creep crack growth tests for a pre-defined period of time to obtain creep damage local to the crack tip. This approach is generally referred to as local creep damage (LCD) method. Note that in this method the extent of crack length at which a CCG test is to be interrupted must be considered and predicted. In this research, the creep crack growth tests on C(T) specimens are started at the initial normalised crack length of $a_0/W = 0.35$, the crack propagation gradually continues until the normalised crack length of around $a/W \approx 0.5$ is attained and finally the CCG tests are interrupted and unloaded.

3.5.1.1 Proposed Method to Interrupt CCG Tests

To provide an approximate measure of the instantaneous crack length during CCG tests in LCD specimens, CCG tests on nominally identical specimens, with the same dimensions and initial normalised crack length of $a_0/W = 0.35$, must be performed at 550 °C for calibration purposes. Subsequent to CCG calibration tests completion, the samples are broken open to calibrate the PD data and the CCG data are then analysed. The PD and LLD rates corresponding to the instantaneous normalised crack length of around $a/W \approx 0.5$ can therefore be estimated from these calibration tests. The obtained PD and LLD rates can be used as indicators to interrupt CCG tests on the LCD specimens at the desired normalised crack length of around $a/W \approx 0.5$.

Note that due to the material variability, specimen machining and some possible experimental errors during CCG tests, it is challenging to interrupt all LCD tests at the exact normalised crack length of 0.5. However, interrupting the LCD tests at the approximate normalised crack length of $a/W \approx 0.5$ doesn't influence the subsequent fracture toughness and fatigue crack growth test results on these local creep damage specimens as long as the final normalised crack length is in range of $0.45 < a/W < 0.7$ which is considered valid for performing fatigue and fracture toughness tests according to the standard test methods.

3.5.2 Global Creep Damage (GCD)

An alternative approach to introduce creep damage into the material is to globally, uniformly creep damage a large block of material (*i.e.* large uniaxial tensile creep sample) and subsequently manufacture fracture specimens from the creep damaged sample. Note that due to the large size of the global creep damage (GCD) samples, these tests must be performed on special hydraulic machines with sufficiently high load carrying capacities. In order to introduce global creep damage into the material, the samples can be pulled in tension at 550 °C until a defined creep strain rate is obtained. To conduct tests in a reasonable time, a relatively high stress range of around 260–300 MPa is required to be applied on the GCD specimens made of PC material. To save material and reduce costing, it is recommended to weld the gauge region of large specimens, made of the desired material, to extension pieces from cheaper stainless steels. To ensure that no significant damage or crack forms in the weld region or extension pieces, the cross sectional area of the gauge region could be designed smaller than other parts so that the applied stress would be higher on the material located in the gauge region.

It must be noted that the temperature and load stability, extensometry design and also data acquisition are significant challenges for running such high load tests on large specimens. Therefore, it was decided to keep the focus of this project on the local creep damage approach and keep the global creep damage as an optional test method to be done in the future.

3.6 Test Matrix

To perform experiments on AR and PC materials, test matrices were developed as shown in Appendix A. As seen in Table A.1 and Table A.2, uniaxial creep rupture tests at 550 °C and also tensile tests at both room temperature and 550 °C were performed on the AR and PC material. In addition, to examine the creep crack growth behaviour of the material subsequent to the introduction of the 8% plastic pre-strain, a few CCG tests have been performed on the PC material as detailed in Table A.2. Fracture toughness and fatigue crack growth tests have also been conducted on the material with and without prior plastic pre-strain. These tests have been repeated on the specimens made of PC material, subjected to a few millimetres of creep crack extension. The details of the specimens containing creep damage local to the crack tip (denoted local creep damage, LCD, samples) are provided in Table A.3. In order to avoid any complications and ambiguity in the data due to *e.g.* creep strain accumulation, all fracture toughness and fatigue tests are performed at room temperature.

3.7 Specimen Design and Orientation

3.7.1 Uniaxial Creep and Tensile Specimens

The dimensions of the room temperature round bar tensile specimens have been defined according to the ASTM standard E8M [106]. For the case of high temperature tensile and uniaxial creep rupture tests, similar specimens have been designed by adding ridges to the top and bottom of the gauge region for extensometry purposes as shown in Figure 3.2. The diameter of all tensile and uniaxial creep specimens in this research is considered 8 mm. The uniaxial samples have been extracted from Header A in such a way that the loading axis is parallel to the axial direction of the header component. When uniaxial specimens are extracted from the pre-compressed blocks, the samples are machined in such a way that the loading axis is also parallel to the pre-straining direction. A schematic illustration of the uniaxial specimen orientation is shown in Figure 3.3.

3.7.2 Fracture Specimens

Following the standard test methods for creep crack growth, fatigue crack growth and fracture toughness [71, 85, 107], the fracture specimens have been manufactured with the thickness to width ratio of $B/W = 0.5$. C(T) samples were manufactured from uncompressed and pre-strained materials with the specimen width of $W = 50$ mm. The fracture specimens have been machined from the header component with the axial-radial orientation (loading axis parallel to the axial direction and the crack plane along the radial direction). For the case of C(T) samples extracted from the pre-compressed blocks, the loading direction of the machined specimens is also parallel to the pre-straining axis. The orientation of C(T) specimens extracted from Header A is schematically shown in Figure 3.3. Note that due to service exposure, the material employed for testing may not be completely homogeneous and isotropic. Thus in order to obtain comparable results from different fracture tests it is important to machine the specimens with the same orientations.

The C(T) specimens for fracture toughness testing were side grooved by 10% of the thickness at each side (20% side groove in total). The rest of the specimens have the total side groove of 30% of the thickness. Note that due to the maximum load carrying capacity of the creep machines, the LCD specimens for fracture toughness testing were also side grooved by 30%. The starter crack in C(T) specimens made for CCG testing was introduced by an EDM notch of root diameter 0.25 mm. However, for fracture toughness and fatigue crack growth specimens, starter cracks with sufficiently sharp crack tips were introduced by pre-fatigue cracking.

3.8 Tables

Table 3.1: Chemical composition in wt% for the 316H material taken from Header A

C	Si	Mn	S	P	Ni	Cr	Mo	Co	B	N
0.052	0.56	1.66	0.06	0.017	11.68	17.28	2.33	0.039	0.0045	0.036

3.9 Figures

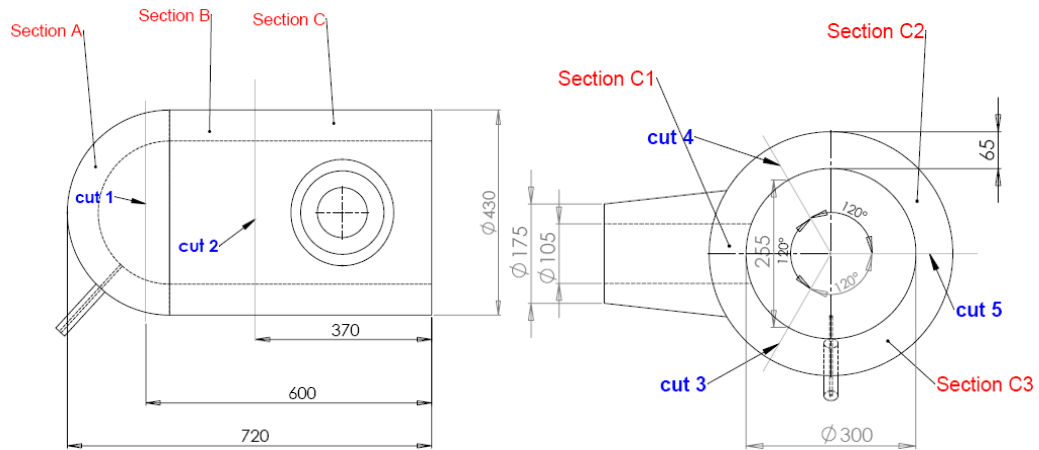


Figure 3.1: Header A dimensions and details

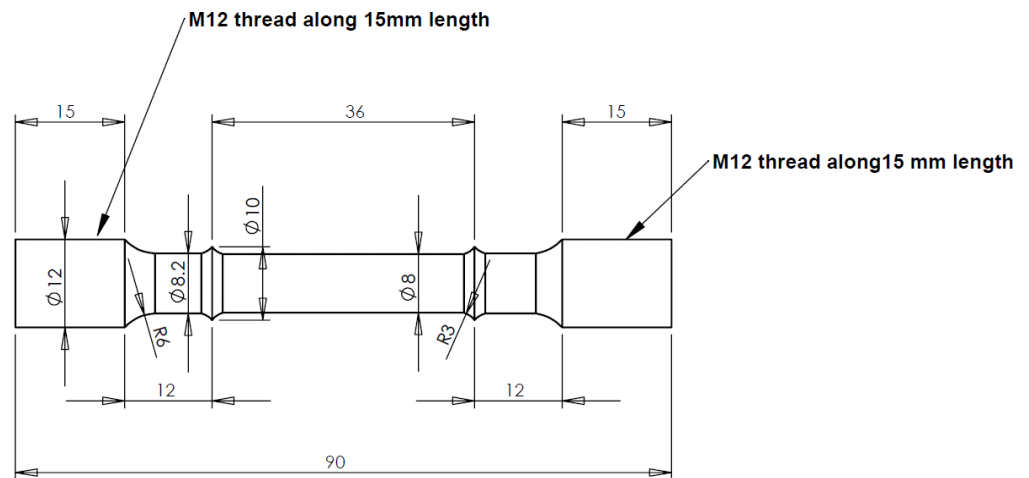


Figure 3.2: Round bar specimen dimensions for high temperature tensile and uniaxial creep rupture tests

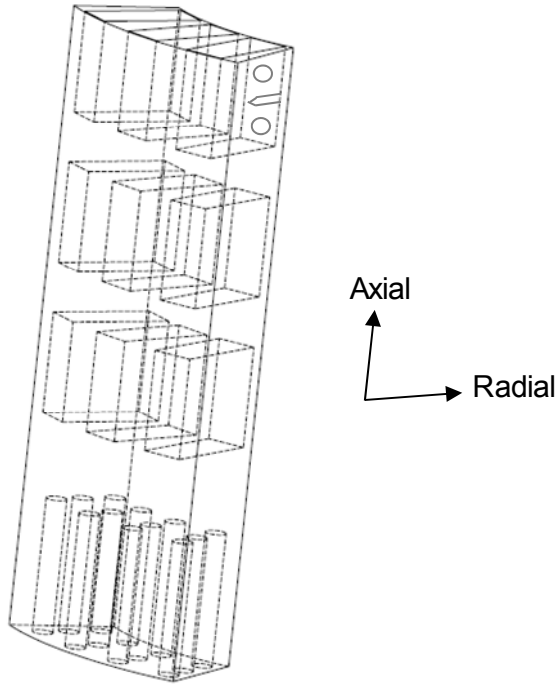


Figure 3.3: Schematic illustration of specimen orientations

Chapter 4

Experimental Testing and Analysis Procedures

A summary of the standard test methods which have been employed for running creep crack growth, fatigue crack growth and fracture toughness tests are detailed in this chapter. In addition, the methods to measure and analyse residual stresses in fracture specimens have been described and the required equations are defined and explained.

4.1 Creep Crack Growth

Creep crack growth (CCG) tests have been performed according to ASTM E1457 [71] and the procedures given in this standard have been followed. The analysis methods for a number of fracture mechanics specimen types have been described in ASTM E1457. All creep crack growth tests in this research have been performed on compact tension, C(T), specimens.

4.1.1 Starter Crack

It has been suggested in ASTM E1457 to introduce a pre-crack in specimens prior to CCG testing. Therefore, all C(T) specimens were pre-cracked by an EDM notch of root radius 0.125 mm. The dimensions of the CCG test specimens are tabulated in Chapter 6.

4.1.2 Crack Length Monitoring

The potential drop (PD) technique is often used to measure the crack length during CCG tests. Subsequent to test completion, samples are broken open and the initial a_0 and average final a_f crack lengths are accurately measured using digital imaging methods. Knowing the initial V_0 and final V_f PD voltage, the instantaneous crack length, a , during the test can be calculated using a linear correlation between PD and crack length given in ASTM E1457 [71] as

$$a = a_0 + \frac{V - V_0}{V_f - V_0} (a_f - a_0) \quad (4.1)$$

4.1.3 Load Line Displacement Monitoring

The load line displacements during CCG tests can be monitored using linear variable differential transformers (LVDT) outside the furnace by designing an external extensometry system. The other method to record LLD data is to accommodate a capacitance gauge at the crack mouth of the specimen. The LVDT extensometry approach was employed for LLD measurements in this research.

4.1.4 Crack Growth and Load Line Displacement Rate Calculations

After measuring the instantaneous crack length and LLD using PD and LVDT, respectively, the crack growth rate and load line displacement rate were calculated using the seven point incremental polynomial technique. Due to insufficient number of data points available for the first and last three points to calculate crack growth and LLD rate using this method, the secant method was applied for rate calculations at these points. More details of the employed rate calculation approaches are given in [108].

4.1.5 Data Analysis

After calculating the crack growth and LLD rates, a data reduction technique was applied to smooth the data trends by reducing the noise and scatter. Reducing the number of data points for each test also facilitates da/dt vs. C^* data analysis by providing a clear trend when the CCG rate is correlated with the C^* parameter. More details of the data reduction approach are given in [108].

4.1.6 C^* Calculation

Under steady state creep conditions, C^* fracture mechanics parameter can be utilised to characterise the creep crack growth behaviour of the material. The C^* parameter corresponding to each data point was experimentally calculated using Eqn (2.102). As suggested in ASTM E1457, the total load line displacement rates were employed in Eqn (2.102) [71].

4.1.7 C^* Validity Criteria

In order to exclude the non-steady state tail regions from da/dt vs. C^* data trends and also to identify the valid regions for the CCG rate to be described by C^* crack tip correlating parameter, the existing C^* validity criteria in the literature were applied on each data set. A summary of these criteria employed in CCG data analyses is given below.

4.1.7.1 ASTM E1457 Validity Criteria

In order to describe the CCI and CCG behaviour of a material using the C^* creep fracture mechanics parameter, the validity criteria specified in ASTM E1457 [71] must be satisfied. These criteria are as follows:

- i. The creep load line displacement rate must constitute at least half of the total load line displacement rate *i.e.* $\dot{\Delta}^c / \dot{\Delta}^T \geq 0.5$. When this criterion is satisfied the material is referred to as creep ductile and the CCG rate can be described using C^* parameter. (For creep brittle materials with $\dot{\Delta}^c / \dot{\Delta}^T \leq 0.25$, K is the recommended parameter to describe the crack growth behaviour [71]). Note that the creep load line displacement rate can be calculated using Eqns (2.103) and (2.104), considering negligible plasticity in load line displacement.
- ii. The second criterion is to ensure that the transition time, t_T , from an elastic crack tip field to a C^* controlled creep crack tip field is exceeded. According to the ASTM E1457, the transition time can be estimated using

$$t_T = \text{Max} \left[\frac{K^2}{E'(n+1)C^*(t_T)} \right] \quad (4.2)$$

As seen in the equation above, the calculation of t_T depends on K and $C^*(t_T)$ which vary during the test. In order to calculate the transition time, it has been suggested in ASTM standard [71] to calculate t_T for each data point using the experimental C^* value and the instantaneous K at time t . Then the maximum value of t_T in the entire data set is selected as the transition time and all the data points prior to the calculated transition time are excluded in da/dt vs. C^* plot.

- iii. The third validity criterion is to exclude all the data points prior to $t_{0.2}$ which is the time corresponding to 0.2 mm creep crack extension. The data points for the time smaller than $t_{0.2}$ are considered to be in the transient crack growth regime where damage is building up to a steady state at the crack tip.
- iv. The fourth validity criterion is to exclude the data points with the load line displacement, Δ^{LLD} , greater than $0.05W$, where W is the specimen width.

4.1.7.2 R5 Additional Validity Criterion

An additional validity criterion has been defined in R5 “assessment procedure for the high temperature response of structures” [38] which suggests that the non-dimensional

crack velocity, λ , must be less than 0.5 in order to use C^* as a valid parameter to characterise the creep crack growth rate.

$$\lambda = \frac{\dot{a}(\sigma_{ref})^2}{EC^*_{exp}} \quad (4.3)$$

Note that the reference stress in λ calculations can be evaluated by Eqn (2.89)–(2.91).

4.1.7.3 EDF Energy Proposed Validity Criteria

A number of modifications to the existing C^* validity criteria specified in ASTM E1457 [71] have been proposed by EDF Energy in [24]. These suggested amendments are as follows:

- i. The initial and final crack length to be measured on the fracture surface using high resolution digital imaging system and image analysis softwares. In this method, the amount of crack extension due to creep is determined by area measurements instead of the nine-point average procedure mentioned in [71]. It has been claimed in [24] that the area measurement technique provides a more accurate estimate of the average crack length compared to the nine-point average technique.
- ii. The analytical predictions of the elastic and plastic displacement rates in ASTM E1457 [71] are considered unreliable as usually lead to negative values of the creep displacement rate. Therefore, the total displacement rates are recommended to be employed in C^* calculations.
- iii. Due to possible scatter and noise in the recorded crack length (from the PD data) and load line displacement data, numerical data smoothing can be performed using a theta equation [109].
- iv. In ASTM E1457 [71], the transition time, t_T , is calculated using Eqn (4.2) and is used to check whether the widespread creep conditions are attained in the fracture specimen or not. The formula given in [71] is valid only for the materials following Norton's deformation law. Even for the materials deforming according to the Norton's creep law, the $C(t)$ transient crack tip parameter is estimated to be approximately $1.6 \times C^*$ at the transition time. Considering the variation of K and C^* throughout the test, it has been suggested in [71] to calculate t_T at each time using Eqn (4.2) and take the maximum value as the transition time. However this method, overestimates the transition time in the materials which don't follow the Norton's law, especially when the primary creep is considerably large. Having known that there is significant primary creep behaviour in the

creep deformation of AR 316H stainless steel, it can be concluded that the transition time calculated using Eqn (4.2) is not very accurate for this material. Thus it has been suggested in [24] to use the experimental redistribution time, t_{red} , instead of transition time, to check widespread creep conditions in creep fracture specimens.

The redistribution time can be estimated based on the creep to elastic displacement ratio. It has been shown in [24] that for a C(T) specimen, the relationship between reference strains and load line displacements can be generally expressed as

$$\frac{\varepsilon_{ref}^c}{\varepsilon_{ref}^e} = 2.4 \frac{\Delta_c}{\Delta_e} \quad (4.4)$$

where Δ_c and Δ_e are the creep and elastic load line displacements, respectively. In R5 [38], the redistribution time has been defined in terms of the reference stress as

$$\varepsilon_{ref}^c[t_{red}] = \varepsilon_{ref}^e = \sigma_{ref} / E \quad (4.5)$$

where ε_{ref}^e is the elastic strain at reference stress σ_{ref} and $\varepsilon_{ref}^c[t]$ is the creep strain corresponding to the reference stress at time t . Combining Eqns (4.4) and (4.5), the redistribution time is given by

$$\frac{\Delta_c(t_{red})}{\Delta_e(t_{red})} = \frac{1}{2.4} = 0.417 \quad (4.6)$$

The use of the experimental redistribution time defined by Eqn (4.6) may be unnecessarily restrictive as it removes the data points which have $C(t)$ values of only a small percentage larger than C^* . Therefore an alternative engineering definition of the redistribution time, t_{red}^{eng} , has been proposed in [24] which can be expressed as

$$\varepsilon_{ref}^c[t_{red}^{eng}] = 0.5 \varepsilon_{ref}^e = \sigma_{ref} / 2E \quad (4.7)$$

Using Eqns (4.4) and (4.7), the engineering redistribution time can be defined as

$$\frac{\Delta_c(t_{red}^{eng})}{\Delta_e(t_{red}^{eng})} = \frac{0.5}{2.4} = 0.208 \quad (4.8)$$

Eqn (4.8) suggests that t_{red}^{eng} is attained when the ratio of creep to elastic load line displacement reaches 0.208. It has been proposed in [24] to use t_{red}^{eng} parameter, instead of the transition time specified in ASTM E1457 [71], to check

the achievement of widespread creep conditions in CCG tests on AR 316H stainless steel.

4.2 Fatigue Crack Growth

The fatigue crack growth (FCG) tests in this work were performed on C(T) specimens according to ASTM E647 [107] and the procedures given in this standard were followed to analyse the test data. The results from these tests are expressed in term of the stress intensity factor range, ΔK , which can be defined by LEFM theories. According to ASTM E647, FCG tests are conducted by applying cyclic loads on cracked specimens and subsequently expressing the crack extension rate per cycle, da/dN , as a function of stress intensity factor range. A summary of the FCG testing and analysis procedures are given below.

4.2.1 Starter Crack

In order to introduce sufficiently sharp crack tips into the test pieces, C(T) specimens were pre-fatigue cracked prior to FCG testing. Pre-fatigue cracking removes the effects of the machined Chevron notch in C(T) specimens and also eliminates possible subsequent crack growth rate effects which may be caused by the change in the crack front shape. It has been suggested in ASTM E647 to ensure that the final maximum stress intensity factor K_{max} during pre-cracking does not exceed the initial K_{max} in the actual FCG tests. Specimens can be heat tinted after pre-fatigue cracking to mark the crack extent due to pre-cracking. Subsequent to FCG test completion, samples are broken open and the initial crack length in FCG tests can be accurately measured. The dimensions of all the FCG specimens are given in Chapter 9.

4.2.2 Frequency and R -Ratio

The FCG tests in this research have been conducted at a relatively high frequency of $f = 10$ Hz under sinusoidal cyclic loading conditions. Although pre-fatigue cracking was performed by the decreasing- K approach (with variable R -ratio) as suggested in ASTM E647, the actual FCG tests were conducted under a constant R -ratio of $R = 0.1$. To keep the R -ratio constant throughout the experiments, all FCG tests were performed under load-controlled mode and P_{min} and P_{max} were kept constant during the test.

4.2.3 Crack Length Measurement

It has been suggested in ASTM E647 to polish the crack growth region prior to test set up and measure the instantaneous crack length during FCG tests by means of visual microscopy techniques (*i.e.* travelling microscope) using indirect lighting on the

sample. However, potential drop (PD) and unloading compliance techniques can also be applied for crack length measurements in FCG tests. The unloading compliance technique has been employed in this research to monitor the crack length for a given number of cycles. More details of the unloading compliance technique can be found in Section 4.3.7. The number of cycles for each experiment was counted throughout the test using an implemented software. The initial and final crack length estimates using the unloading compliance data can be compared to those of measured on the fracture surface (using digital imaging technique after specimen break open) to examine the accuracy of the unloading compliance approach.

4.2.4 Data Analysis

Similar to the creep crack growth rate calculations explained in Section 4.1.4, the crack extension rate per cycle, da/dN , was calculated using an incremental polynomial method for each of the FCG tests examined. Having measured the crack length for a given number of cycles throughout the test, the stress intensity factor range has been calculated for each increment of crack extension using Eqn (2.129).

da/dN is plotted against ΔK in log-log axes and a regression fit to the data made to determine the Paris law constants (see Eqn (2.130)) in the steady state FCG region. ASTM E647 recommends using the minimum of five data points of approximately equal spacing to establish the straight line fit to the FCG data [107].

4.3 Fracture Toughness

The fracture toughness tests in this work have been conducted following the ESIS P2-92 “procedure for determining the fracture behaviour of materials” [85]. The load and displacement data have been recorded during each loading and unloading. The load line displacement data have been recorded by external extensometry using LVDTs located at some distance away from the specimens and attached to extensometer legs which are connected to the shackles. A summary of the fracture toughness testing and analysis method for ductile materials with stable crack extension is given below.

4.3.1 Stable Crack Growth Parameters

In ductile materials, if the force *vs.* displacement behaviour of a cracked body is non-linear and stable crack growth is observed, the variation of J can be plotted against the crack extension Δa . This J *vs.* Δa variation is generally referred to as resistance curve (R-curve). There are two approaches for generating the resistance curve. In the first

method, a series of identical specimens are loaded to different load levels and then unloaded. Subsequent to stable crack growth appearance in each specimen, they are heat tinted to mark the crack extension and then broken open to measure the extent of crack propagation. The second approach is the unloading compliance method which is conducted on a single specimen. In this approach, the specimen is frequently unloaded during the test and the instantaneous crack length is approximated using the unloading compliance measurements.

Using any of the above approaches, sufficient number of data points must be achieved to describe J vs. Δa behaviour of the material adequately. The main difference between these two methods is that in the multiple specimen approach, the average fracture properties are attained whereas in the single specimen procedure some additional information on the material variability and inhomogeneity can be obtained.

The fracture parameters which are calculated based on the ESIS test method to quantify J value at the onset of crack initiation are:

- i. $J_{0.2/BL}$ which provides an engineering definition of crack initiation and measures the material resistance at 0.2 mm of stable crack extension. Note that 0.2 mm crack extension is known as a small measureable amount of crack propagation which is large enough to be detected by the test monitoring equipments and is small enough to represent initiation.
- ii. $J_{0.2}$ which provides a measure of material resistance at 0.2 mm total crack growth, including the crack tip blunting.
- iii. J_g which is the maximum fracture resistance value and can be measured from a test specimen.
- iv. dJ/da which gives the instantaneous slopes of the R-curve and is used to measure the resistance of material to crack growth.

4.3.2 Specimen Geometry

The specimen geometries recommended in ESIS P2-92 for fracture toughness tests are C(T) and single edge notch bend specimens. C(T) specimen geometry was selected for performing fracture toughness tests in this research. Standard dimensions based on the specimen width, W , similar to those of given in ASTM E1457 [71] and ASTM E1820 [84], are proposed in ESIS P2-92 standard. Note that according to ESIS P2-92, the initial crack length normalised by the specimen's width must be in the range of $0.45 < a_0/W < 0.65$ to obtain valid results from a fracture toughness test.

4.3.3 Pre-fatigue Cracking

In order to introduce a sufficiently sharp crack tip into the specimen, samples must be pre-fatigue cracked from the end of the machined notch. The fatigue pre-crack in fracture specimens must exceed the maximum of 1.5 mm or $0.05a_{notch}$, where a_{notch} is the crack length measured up to the end of the machined notch. The maximum pre-fatigue force must not exceed the minimum of $0.6F_y$ or the force corresponding to $K_{max}/E \leq 1.5 \times 10^{-4} \sqrt{m}$, where K is the stress intensity factor, E is the Young's modulus and F_y for a C(T) specimen is given by

$$F_y = \frac{B(W-a)^2}{(2W+a)} \sigma_{0.2} \quad (4.9)$$

where B is the specimen thickness and $\sigma_{0.2}$ is the yield strength of the material.

4.3.4 Side-Grooving

When the test specimen thickness is smaller than the component thickness, specimens are recommended to be side-grooved subsequent to pre-fatigue cracking. The advantage of having side-grooves in the specimen is to promote straight fronted ductile crack growth during the test. The total depth of the side-grooves in a specimen must be in the range of $0.10B < (B-B_n) < 0.25B$ where B_n is the net thickness between the side grooves. For most of the fracture specimens, the total side grooving of 20% is recommended.

4.3.5 Fracture Testing

When a single specimen approach is employed, a specimen is partially unloaded and then reloaded during the test at a number of specified intervals so that evenly spaced data points are obtained. A single specimen test with around 30 unloadings provides enough number of data points to generate the R-curve accurately with the data spacing requirements specified in the standard. According to the ESIS P2-92 standard, the unloading should not exceed 30% of the current force at each interval and preferably must be as small as practicable.

The linear unloading compliance at each stage provides an approximate measure of the instantaneous crack length at each interval. The loading/unloading rate should be as fast as possible to minimise time dependent effects, though must be slow enough to record sufficient number of data points for an accurate measurement of the compliance. It is recommended to hold the displacement constant prior to each unloading, until time dependent plasticity effects are stopped and thus force relaxation occurs in each

interval. It is also recommended to make at least three unloading compliance measurements at the load levels lower than the maximum pre-fatigue cracking force applied initially on the sample. Non-linear parts of the force vs. displacement record which may occur at low load levels due to possible crack closure effects must be excluded from compliance measurements.

After the final unloading, the load should be reduced to zero and the sample must be heat tinted to mark the final crack length. Subsequently the specimen is broken open to measure the final crack length using the 9 point average method or digital imaging technique.

4.3.6 Fracture Resistance Calculation

The fracture resistance at the k th interval can be calculated by

$$J_{0,k} = \frac{\eta U_k}{B_n(W - a_0)} \quad (4.10)$$

where B_n is the net thickness between the side grooves, a_0 is the initial crack length, U_k is the area under the force vs. displacement curve up to the line of constant displacement at the k th interval and $\eta = 2.2$ [67] for a C(T) specimen with the crack length in the range of $0.45 \leq a/W \leq 0.7$.

The equation above doesn't allow for crack extension during a test and thus is used to calculate the fracture resistance for the crack extensions $\Delta a < 0.10(W - a_0)$, where the errors in the J values are usually negligible. Though when $\Delta a \geq 0.10(W - a_0)$, a corrected solution of the fracture resistance can be approximated by,

$$J = J_0 \left(1 - \frac{(0.75\eta - 1)\Delta a}{(W - a_0)} \right) \quad (4.11)$$

4.3.7 Crack Length Estimates from the Unloading Compliance

Subsequent to single specimen fracture toughness test completion, the unloading compliance measured along the load line displacement must be determined at each interval. The instantaneous crack length normalised by the specimen's width can be calculated using the measured compliance at each interval by

$$a/W = \left(\frac{1.000196 - 4.06319\mu + 11.242\mu^2}{-106.043\mu^3 + 464.335\mu^4 - 650.677\mu^5} \right) \quad (4.12)$$

where

$$\mu = \frac{1}{\sqrt{B_e E_M C + 1}} \quad (4.13)$$

$$B_e = B - (B - B_n)^2 / B \quad (4.14)$$

and the effective Young's modulus E_M can be calculated from

$$E_M = \frac{1}{C_0 B_e} \left(\frac{W + a_0}{W - a_0} \right)^2 \left[\begin{aligned} &2.163 + 12.219(a_0 / W) - 20.065(a_0 / W)^2 \\ &- 0.9925(a_0 / W)^3 + 20.609(a_0 / W)^4 - 9.9314(a_0 / W)^5 \end{aligned} \right] \quad (4.15)$$

where C_0 is the average of the compliances determined in the elastic region. E_M calculated from the above equation is expected to be very close to the E value obtained from a tensile test.

To account for the changes in the specimen geometry, which can occur due to loading, the compliance should be corrected considering the rotation in the sample. The corrected compliance, C_c , can be calculated by

$$C_c = \frac{C}{[(h \sin \theta / r - \cos \theta)(D \sin \theta / r - \cos \theta)]} \quad (4.16)$$

where C is the measured compliance, h is half of the initial distance between the centres of the pin holes, r is the radius of rotation calculated as $0.5(W+a)$, D is one half of the initial distance between the displacement measurement points and θ is the angle of rotation given by

$$\theta = \sin^{-1} \left[\frac{0.5q + D}{\sqrt{D^2 + r^2}} \right] - \tan^{-1}(D / r) \quad (4.17)$$

In Eqn (4.17) q is the total measured load line displacement during the test.

4.3.8 Crack Growth Limit

The R-curve is generated by plotting J values against the corresponding crack extensions, Δa . In order to identify valid data points for fracture toughness analysis, the crack growth limits must be applied by

- i. Calculating the maximum allowable crack extension using

$$\Delta a_{max} = 0.10(W - a_0) \quad (4.18)$$

- ii. Specifying the slope of the blunting line (also known as construction line) using

$$\Delta a_B = 0.4d_n^* J / E \quad (4.19)$$

where d_n^* can be calculated by

$$d_n^* = \varepsilon_0^{n-1} (0.787 + 1.554n - 2.45n^2 + 16.952n^3 - 38.206n^4 + 33.13n^5) \quad (4.20)$$

In Eqn (4.20), n is the strain hardening exponent which can be determined from

$$\frac{\sigma_{0.2}}{\sigma_{UTS}} = \frac{1}{1 + (\sigma_{0.2} / E + 0.002)} \left(\frac{2.718 \ln(1 + (\sigma_{0.2} / E + 0.002))}{n} \right)^n \quad (4.21)$$

Also the reference strain ε_0 in Eqn (4.20) can be calculated by $\varepsilon_0 = \sigma_0 / E$ where the reference stress σ_0 is determined from

$$\sigma_0 = 10^t \sigma_{0.2} \quad (4.22)$$

where

$$t = \frac{n \log((\sigma_{0.2} + 0.002E) / \sigma_{0.2})}{n - 1} \quad (4.23)$$

Alternatively, the slope of the blunting line can be approximated by

$$J = 3.75 \sigma_{UTS} \Delta a_B \quad (4.24)$$

- iii. Plotting the blunting line on the R-curve graph
- iv. Constructing the crack growth exclusion line parallel to the blunting line at an offset corresponding to Δa_{max}
- v. Constructing an exclusion line parallel to the blunting line at the offset of 0.10 mm.
- vi. Considering the data points falling in between exclusion lines valid and the rest assumed invalid

4.3.9 Data Spacing Requirement

At least six data points must be employed to describe the crack growth fracture resistance behaviour. The data points should be evenly spaced.

4.3.10 Curve Fit

The equation of the line of best fit must be determined considering the data points lie between 0.1 mm and Δa_{max} exclusion lines on the J vs. Δa graph.

4.3.11 J Validity Limits

To determine material's fracture behaviour independent of the specimen size, an exclusion line must be constructed on the J vs. Δa graph at J_{max} value which can be calculated by

$$J_{max} = \text{Min}\{(W - a_0)(\sigma_{0.2} + \sigma_{UTS}) / 40, B(\sigma_{0.2} + \sigma_{UTS}) / 40\} \quad (4.25)$$

Subsequently, Ω parameter must be determined from

$$\Omega = \frac{W - a_0}{J_g} \frac{dJ}{d(\Delta a)} \Big|_{J_g} \quad (4.26)$$

where $\frac{dJ}{d(\Delta a)}$ is the differentiation of the equation of the line of best fit to the valid data points on the R-curve, as discussed in 4.3.10. In Eqn (4.26) $\frac{dJ}{d(\Delta a)}$ should be evaluated at J_g . Note that J_g , which is defined as the upper limit to J -controlled crack growth behaviour for the test specimen size, can be provisionally taken as the intersection of the best fit curve with either J_{max} or Δa_{max} exclusion lines. If $\Omega \geq 10$, the fracture resistance curve enclosed by J_{max} and Δa_{max} exclusion lines may be considered as material property independent of the specimen size.

4.3.12 J Fracture Parameters

The $J_{0.2/BL}$ fracture parameter can be calculated by constructing a line parallel to the blunting line at the offset of 0.2 mm. The intersection between this line and the best fit curve to the fracture resistance data points is regarded as $J_{0.2/BL}$. Note that at least one data point should exist between this offset line and the 0.1 mm exclusion line. $J_{0.2/BL}$ is only valid if,

$$J_{0.2/BL} < J_{max} \quad (4.27)$$

and

$$(dJ / da)_{BL} > 2(dJ / da)_{0.2/BL} \quad (4.28)$$

where $(dJ / da)_{BL}$ is the slope of the blunting line and $(dJ / da)_{0.2/BL}$ is the slope of the best fit curve at $J_{0.2/BL}$.

In addition, the $J_{0.2}$ fracture parameter can be evaluated by constructing a line corresponding to the total crack extension of 0.2 mm. The intersection between this constant crack growth line and the best fit curve to the valid fracture resistance data points on the J vs. Δa curve is defined as $J_{0.2}$. Note that at least one data point should exist between 0.2 mm and 0.4 mm crack extension and $J_{0.2}$ is only valid if it's smaller than J_{max} .

4.4 Residual Stress Measurements

Neutron diffraction (ND) is a non-destructive approach which is widely used to investigate the deformation of families of grains with different crystallographic orientations. A standard test method for determination of residual stresses/strains by neutron diffraction for polycrystalline materials is developed and presented in ISO/TTA

standard [110]. Further details on the ND methods and data analysis procedures are also described in [111]. Residual strain measurements can be performed using reactor sources with a continuous monochromatic beam of neutrons or spallation sources. The neutron wavelength in a reactor source is known and the diffraction angle is measured. However at spallation sources the diffraction angle is kept constant and the wavelength is determined by measuring the energy and the time it takes for a neutron to travel to the detector (*i.e.* time of flight (TOF)) [110]. By keeping the diffraction angle or the wavelength constant and quantifying the varying parameter, the lattice spacing can be calculated using the Bragg's law.

When a specimen is stressed, its lattice spacing is altered. The elastic residual strain in the material is determined from the change in crystal lattice spacing using Bragg's Law [110]

$$2d_{hkl}\sin\theta_{hkl} = n\lambda \quad (4.29)$$

where n is an integer, λ is the wavelength of the radiation, d_{hkl} is the lattice plane spacing and θ_{hkl} is the Bragg angle for a given crystallographic plane, denoted using Miller indices hkl [110].

At reactor sources, a continuous monochromatic beam of neutrons is produced by using a monochromator to select a given neutron wavelength from a polychromatic neutron beam. If a sample is placed in a monochromatic beam of neutrons, then its lattice spacing can be determined if the incident wavelength of the diffracting neutrons is known. Elastic strain will be apparent as a shift in the value of diffraction angle from an incident beam, $2\theta_{hkl}$, for a particular crystallographic plane. The residual strain in a specified direction is determined using

$$\varepsilon_{hkl} = \frac{d_{hkl} - d_{0,hkl}}{d_{0,hkl}} = (\theta_{0,hkl} - \theta_{hkl}) \cot\theta_{0,hkl} \quad (4.30)$$

where d_0 and θ_0 are the strain free values of the lattice spacing and Bragg angle, respectively. These values may be determined from measurements on strain free samples or by imposing force and moment equilibrium [110]. The standard uncertainty in the strain measurements, u_ε , can be calculated by

$$u_\varepsilon^2 = \frac{u_d^2}{d^2} + \frac{u_{d_0}^2}{d_0^2} \quad (4.31)$$

where u_d and u_{d_0} are the uncertainties in the d-spacing and strain free d-spacing measurements, respectively.

When strains are measured in three mutually orthogonal directions, the direct stresses, σ_1 , σ_2 and σ_3 can be determined from

$$\sigma_i = \frac{E_{hkl}}{1+\nu_{hkl}} \varepsilon_i + \frac{\nu_{hkl} E_{hkl}}{(1+\nu_{hkl})(1-2\nu_{hkl})} (\varepsilon_1 + \varepsilon_2 + \varepsilon_3) \quad (4.32)$$

where i is the direct stress index, E_{hkl} and ν_{hkl} are the elastic modulus and Poisson's ratio, respectively. When measurements are associated with a specific crystallographic plane, then plane specific values of the elastic modulus and Poisson's ratio must be used in Eqn (4.32). The properties employed in this work are based on the Kroner model for austenitic steels and the recommended plane for measurement is $\{311\}$ $[111]$. The standard uncertainty in the stress measurements, u_{σ_i} , can be calculated by

$$u_{\sigma_i}^2 = \left(\frac{E_{hkl}}{1+\nu_{hkl}} u_{\varepsilon_i} \right)^2 + \left(\frac{\nu_{hkl} E_{hkl}}{(1+\nu_{hkl})(1-2\nu_{hkl})} \right)^2 (u_{\varepsilon_1}^2 + u_{\varepsilon_2}^2 + u_{\varepsilon_3}^2) \quad (4.33)$$

where u_{ε_i} is the uncertainty in strain measurements along direction i .

In ND measurements, a suitable gauge volume size must be chosen to achieve acceptable resolution in the residual stress/strain measurements within a reasonable period of time. It is recommended in [110] to use the minimum gauge volume of 8 mm^3 in ND measurements to ensure sufficiently large number of grains are included in the measurements.

Chapter 5

Pre-Compression Effects on Mechanical and Uniaxial Creep Deformation

5.1 Introduction

The pre-straining process is generally known to have significant effects on the mechanical, creep damage and deformation behaviour of the engineering materials [97]. A review of the pre-straining effects on the fracture behaviour of 316 SS was given in Chapter 3-Section 3.1.1. As shown in Chapter 3, the introduction of 8% plastic pre-strain at room temperature, in particular, is known to have strong effects on the mechanical and creep properties of 316H steel [1, 89]. In this work, tensile tests at room temperature and 550 °C have been performed in addition to uniaxial creep rupture tests at 550 °C. The results are presented in this chapter to characterise the properties of the material employed (Header A). Note that for the case of PC material, all uniaxial samples were extracted from the pre-compressed blocks in such a way that the tensile axis was parallel to the pre-straining direction.

5.2 Tensile Response

Tensile tests have been performed on as-received (AR) (*i.e.* uncompressed) and pre-compressed (PC) materials at both room temperature and 550 °C. A comparison of the tensile response of the material with and without prior pre-compression is shown in Figure 5.1(a) and Figure 5.1(b) in terms of engineering and true stress-strain, respectively. The tensile properties of the pre-compressed and AR materials at both room temperature and 550 °C are determined and summarised in Table 5.1. The Young's modulus, E , ultimate tensile strength (UTS) and the tensile strain at failure shown in Table 5.1 have been taken from Figure 5.1(a) and are based on the engineering stress and strain response. The proof stress values tabulated in Table 5.1 have been extracted from Figure 5.1(b) and are based on the true strain values. Two tests were performed on each material at both room temperature and 550 °C and the results were found consistent. The tensile properties given in Table 5.1 are the average of the values obtained from two experiments. All tensile tests were conducted under

displacement control mode with a constant displacement rate of 2 mm/minute and the local displacement within the gauge length of the sample was measured using an external attachment to the specimen via extensometry for high temperature tests and a clip gauge for room temperature tests. The initial gauge lengths in room temperature and high temperature tensile samples were 52 mm and 36 mm, respectively, which led to a tensile strain rate of $6 \times 10^{-4} \text{ s}^{-1}$ in room temperature and $9 \times 10^{-4} \text{ s}^{-1}$ in high temperature tensile tests.

5.2.1 Room Temperature Tensile Tests

As seen in Figure 5.1, the yield strength of the material is considerably increased by pre-straining at both room temperature and 550 °C. A large discrepancy is observed in the room temperature tensile behaviour of AR and PC materials in the true stress range of 400-600 MPa. A comparison of the total true strain values at 500 MPa shows that the AR material experiences a strain value which is over twice that of PC material. However, it can be seen in both engineering and true stress-strain curves that the room temperature tensile curves lie close to each other for total strain values of over 30%.

The 0.2% and 0.5% room temperature proof stress values shown in Table 5.1 indicate that the hardening effects on PC material are exhibited at plastic true strains of over approximately 0.5%. This implies a combined isotropic/kinematic hardening behaviour in the materials examined.

Having mentioned that the room temperature tensile curves lie close to each other for strain values over 30%, the UTS and the engineering strain at failure using both axial displacement and the reduction of area (ROA) are found very similar in AR and PC materials. It is noted that the failure strain of PC material is slightly smaller than the one observed in AR material.

5.2.2 High Temperature Tensile Tests

The hardening effects in room temperature tensile tests are observed for true plastic strain values of over 0.5%, however Table 5.1 shows that a 50% increase in the 0.2% and 0.5% proof stress values of the PC material compared to the ones from AR material is observed at 550 °C. This means that the hardening effects at high temperature can be seen in a wider strain range, starting from the beginning of the tensile tests.

As seen in Table 5.1, the elastic Young's modulus of PC and AR materials in high temperature tensile tests are the same and the UTS in PC material is similar to the one observed in AR material. Further comparison between the failure strain of the PC and

AR materials at 550 °C shows that the PC material generally has a smaller failure strain, calculated using both axial displacement and ROA, compared to the AR material.

5.2.2.1 Dynamic Strain Aging (DSA)

Also observed in the tensile response of AR and PC materials at 550 °C in Figure 5.1 are the dynamic strain aging effects. Some DSA effects are observed in the PC material, though less evident than those of seen in the AR material, which may indicate that the PC process alters the distribution of the dislocations in the steel. Similar observation of the DSA effects is reported in [15] for AISI 316 material at elevated temperatures (400 °C and the tensile strain rate of 10^{-5} s^{-1}). DSA is also reported in [14] for 316 material with and without prior pre-strain tested under tension at 600 °C. The DSA phenomenon is often observed in the tensile response of the 316H SS when the test is performed at sufficiently high temperature with a large enough constant strain rate.

As seen in Figure 5.1, the DSA in the tensile tests at 550 °C start to occur when a certain amount of strain is attained, below which the material has no evidence of DSA effects. In the PC material, the DSA started to occur at the approximate total true strain of 1.9% which corresponds to 335 MPa true stress. However in AR material, DSA is observed at the true strain of around 0.7% and true stress of 200 MPa. Therefore, DSA is generally found to take place at a larger stress and strain in the PC material compared to the AR.

5.2.3 Ramberg-Osgood Material Model Constants

In order to investigate the variation of non-linearity in the tensile tests performed at 550 °C, power law fits have been made to the true plastic strain *vs.* true stress data ($\epsilon_p = A_p \sigma^N$) for the PC and AR materials in Figure 5.2(a) and Figure 5.2(b), respectively. As seen in these figures, power law constants have been worked out for three distinct regions of plastic strain and significant change in the power law exponent *N* is observed from small to intermediate and large strain regions. Although the non-linear behaviour of 316H steel cannot be described using a unique power law exponent, but *N* = 3 and 5 have been found to provide the best general fits to the tensile data for AR and PC materials, respectively.

The Ramberg-Osgood material model fits to the high temperature total true stress *vs.* total true strain data on the AR and PC materials are illustrated and compared in Figure 5.3. The constants obtained from these fits to the tensile data (see Eqn (2.7)) are given in Table 5.2. As seen in this table, the normalising stress and the Ramberg-Osgood

stress exponent, N , in PC material are larger than the ones in AR material. Note that the normalising stress, σ_{p0} , has been taken as $\sigma_{0.2}$ proof stress in Ramberg-Osgood material model calculations. Also seen in Table 5.2 is that the power law coefficient of the PC material is considerably smaller than that of found for the AR material.

5.3 Uniaxial Creep Rupture

In order to characterise the creep properties of Header A and also to investigate the influence of pre-compression on the uniaxial creep deformation of 316H stainless steel, creep rupture tests have been performed on four specimens made of PC material and 5 on AR material (see test matrices in Appendix A). Similar stresses have been applied in the tests performed on AR and PC materials to compare the results directly. Note that the uniaxial creep specimens made of pre-compressed material extracted from Header A, are marked as A2 and the ones made of AR material from Header A are designate with A1.

The uniaxial creep properties of the AR and PC materials extracted from this header are compared to those of previously achieved from Header B and Header C. All tests were performed at 550 °C. To compare the uniaxial creep behaviour of PC material to AR, the load up curves for all the uniaxial creep rupture tests performed are presented next and then the creep properties are described.

5.3.1 Load up Behaviour

The loading details for the uniaxial creep rupture tests conducted on the PC and AR samples are given in Table 5.3 and the test loading curves are presented and compared in Figure 5.4. As seen in Figure 5.4, predominant linear behaviour is observed in the load up response of the tests performed on PC specimens with the nominal stress range of 257-300 MPa whereas significant plastic strains of around 2.5-6% are observed in AR material under similar stress levels. The total and plastic strains observed during the loading process of each test are noted in Table 5.3. As seen in Table 5.3, negligible plasticity is quantified during load up in PC material. This can be related to the hardening effects which are introduced by pre-compression, as discussed in Section 5.2.2. Note that the stress range chosen for the uniaxial creep rupture tests are generally close to $\sigma_{0.2}$ obtained from the tensile tests on PC material at 550 °C, thus little plasticity is expected to happen during load up of PC specimens. The load up trends observed in these tests are consistent with those of previously seen in [1], for the materials extracted from different headers.

5.3.2 Uniaxial Creep Properties

The engineering stress and creep strain at failure, also known as creep ductility, are presented in Table 5.3 for all the uniaxial creep rupture tests performed on AR and PC specimens. The creep ductility for each test has been calculated by employing the axial displacement and reduction of area using Eqns (5.1) and (5.2), respectively.

$$\varepsilon_f^{eng} = \frac{\Delta_f - \Delta_{load}}{l_0} = \frac{\Delta_c}{l_0} \quad (5.1)$$

$$\varepsilon_f^{ROA} = \frac{A_0 - A_f}{A_0} - \varepsilon_{p,load}^{eng} \quad (5.2)$$

In Eqns (5.1) and (5.2) ε_f^{eng} is the creep ductility calculated using the axial displacement, Δ_c is the axial creep displacement during the test, which is calculated by subtracting the total axial displacement at failure Δ_f from the axial displacement at load up Δ_{load} , l_0 is the initial gauge length, ε_f^{ROA} is the creep ductility calculated based on ROA, A_0 is the initial cross sectional area at loading, A_f is the local cross sectional area at failure region and $\varepsilon_{p,load}^{eng}$ is the axial plastic strain at loading. Note that Eqn (5.2) provides an estimated value of the creep ductility since the cross sectional area after the plastic necking occurs during the loading process, $A_{p,load}$, cannot be measured and thus the axial plastic strain during load up is employed in this equation. As seen in Table 5.3, the creep ductility calculated using ROA is generally greater than the axial measurements. The advantage of ROA method is that the measurements are performed local to the failure region and thus more realistic creep ductility values may be obtained compared to the axial measurements in which the average of the total displacement in the entire gauge length is accounted for in calculations.

The true stress and true creep strain at failure, using both axial and ROA measurements, are also given in Table 5.3. The true stress is calculated using the nominal stress, σ_{nom} , and the engineering strain during load up, ε_{load}^{eng} , by

$$\sigma_{true} = \sigma_{nom} \left(1 + \varepsilon_{load}^{eng} \right) \quad (5.3)$$

where σ_{nom} is equal to the ratio of the applied load, P , to the specimen's original cross sectional area. The true creep strain is calculated using the axial engineering creep strain, ε^{eng} , by

$$\varepsilon^{true} = \ln\left(\frac{l}{l_0}\right) = \ln(1 + \varepsilon^{eng}) \quad (5.4)$$

The true strain using the area/diameter measurements can be calculated using

$$\varepsilon^{true} = \ln\left(\frac{A_0}{A}\right) = \ln\left(\frac{d_0}{d}\right)^2 = 2\ln\left(\frac{d_0}{d}\right) \quad (5.5)$$

where d_0 and d are the initial and final diameters, respectively. In order to calculate the true creep strain at failure using the local area/diameter measurements, the effect of plasticity during load up on the final diameter size must be taken into account. This can be provided by subtracting the true strain calculated based on ROA measurements from the axial true plastic strain during load up

$$\varepsilon_f^{true} = 2\ln\left(\frac{d_0}{d}\right) - \varepsilon_{p,load}^{true} \quad (5.6)$$

All the true stress and strain measurements, using axial displacement and ROA, for the tests performed on AR and PC materials are given in Table 5.3.

A comparison of the creep ductility for the AR and PC materials tested under similar stresses in Table 5.3 shows that the creep strain at failure in PC material calculated using both axial and ROA is generally smaller than the ones achieved in AR material. In other words the pre-compression process reduces the creep ductility.

Direct comparisons between the uniaxial creep behaviour of AR and PC materials, are shown in Figure 5.5 and Figure 5.6 for 300 and 257 MPa, respectively. As seen in Figure 5.5(a), and Figure 5.6(a), the creep ductility and time to rupture are significantly reduced in PC material compared to those observed in AR material, under both stress levels.

Detailed comparison of the percentage duration spent and creep strain accumulated in primary, secondary and tertiary creep regions are given in Table 5.4 for each test and examples are shown in Figure 5.5(b) and Figure 5.6(b) for two different stress levels. In Figure 5.5(b) and Figure 5.6(b), the time is normalised by the rupture time, t_r , and the creep strain is normalised by the creep strain at failure, ε_f^{eng} , to provide a clear comparison between the fraction of time and creep strain observed in each creep region. Note that the axial creep strain has been employed in these calculations to work out the percentage of strain and time in each creep region.

Comparing the percentage of elapsed time in each creep region shows that in PC material the largest proportion of time is generally observed in the secondary creep region at different stress levels. Under relatively high stresses (see Figure 5.5(b) for 300 MPa), the time elapsed in primary region is negligible and under low stresses (see

Figure 5.6(b) for 257 MPa) the time elapsed in tertiary region is small in the pre-compressed material compared to AR. Table 5.4 confirms that by decreasing the stress levels applied on the PC material, the percentage of the test time spent in tertiary creep region drops but the percentage of time spent in primary creep region increases. This can be explained by the fact that at lower stresses, it takes a longer time for the strain hardening to dominate the thermal recovery processes (see Chapter 2-Section 2.2.1.1), especially in the presence of high dislocation density (for the case of the PC material). Therefore a larger primary region is observed in the creep behaviour of the PC material at low stress levels. However, as the applied stress increases, the strain hardening effects become less significant and thus smaller primary regions are observed which result in larger percentages of tertiary creep.

Comparison of the time proportions spent in each creep region for the tests on AR material shows that decreasing the stress level leads to a considerable drop in the percentage of time elapsed in secondary creep region but the percentage of time spent in primary and tertiary regions increase, with the time spent in the tertiary region being dominant at lower stresses. Note that the stress ranges applied on the specimens are close to or just above the yield stress of the PC material at 550 °C but much higher than the yield stress of the AR material at this temperature. Therefore, large fractions of the tertiary creep observed in the AR material can be related to the significant plasticity effects under high stress levels applied (relative to yield).

Further shown in Table 5.4 is the percentage of creep strain measured in each region for different tests. It can be seen in this table that by decreasing the stress levels in the tests on PC material, the percentage ratio of creep strain to creep ductility in primary and secondary regions increase whilst a significant drop in the percentage of creep strain measured in the tertiary region is observed. It can be also seen in this table that by decreasing the applied stress on the AR material, the percentage of creep strain to creep ductility remains almost unchanged in primary region , however noticeable decrease and increase are observed in secondary and tertiary creep regions, respectively.

The variation of the creep ductility against true stress is examined and shown in Figure 5.7. Note that creep ductilities in Figure 5.7 are based on true strain definition. The creep strain at failure measured base on the axial displacement and ROA for the tests performed on specimens extracted from Header A are compared to the results available from other headers in Figure 5.7(a) and Figure 5.7(b), respectively. No particular trend can be observed in the creep ductility data for the AR material in Figure

5.7. Thus for the range of stresses examined, the creep strain at failure for AR material may be considered stress independent within the inherent scatter. It can be seen in Figure 5.7 that a large scatter is apparent when the results from Header A are compared to the ones available from Headers B and C. This can be related to the cast to cast variability effects and the inherent scatter in the creep data for 316H SS.

Figure 5.7(a) and Figure 5.7(b) show that some stress dependency can be seen in the creep ductility of the PC material for the range of stresses examined, especially when the data based on ROA are plotted against true stress (see Figure 5.7(b)). Also observed in Figure 5.7 is a negligible cast to cast variability in the creep ductility data on the PC material obtained from different headers. However, more number of tests on the PC material extracted from different casts and headers are needed to generalise this observation.

The minimum creep strain rate, time to rupture and average creep strain rate data are plotted against true stress in Figure 5.8, Figure 5.9 and Figure 5.10, respectively. In these figures, the lines of best fits to the experimental data points have been plotted to determine the creep properties of the AR and PC materials extracted from Header A. Also included in these figures are the corresponding data points available on other headers taken from [1]. The creep power law constants determined from the regression fits to the data points are summarised in Table 5.5. In Figure 5.8—Figure 5.10 the AR and PC data points are shown in diamond and square symbols, respectively, the data points from Header A are illustrated in solid symbols whereas those of obtained from other headers are shown in hollow symbols. Also included in these figures are the dashed lines and solid lines showing the trends of the AR and PC materials, respectively. The thinner lines represent the fits to the data points from Header B and C, whereas the thicker lines have been used to show the trend of the data points from Header A. Note that due to the limited number of data points available for the PC material extracted from Header C, indicative fits to the data points have been made in Figure 5.8—Figure 5.10 by assuming the same power law stress exponent as Header A. Similarly, the slope of the AR data trends from Header A have been assumed the same as those of obtained from Header B and C.

The minimum creep strain rate behaviour of the AR material is compared to the PC material in Figure 5.8 for the range of stresses examined. As seen in Figure 5.8, some cast to cast variability can be observed in the minimum creep strain rate behaviour of the AR material. Considering the same slope for the AR materials from different headers, the power law stress coefficient in Header A, given in Table 5.5, is found

larger than the one for Header B&C. It can be also seen in Figure 5.8 that unlike AR material, the cast to cast variability in the minimum creep strain rate behaviour of PC material is insignificant. Figure 5.8 shows that the minimum creep strain rate trend of the AR material from Header A is very similar to the PC material from the same header. This implies that the pre-compression process had negligible effects on the minimum creep strain rate behaviour of the material extracted from this header under the range of stress examined.

The rupture time behaviour of the AR and PC materials from different headers are shown in Figure 5.9 and the determined power law constants are summarised in Table 5.5. As seen in Figure 5.9, large cast to cast variability is observed in the rupture data on the AR material, whilst similar trends have been achieved from different headers for PC material. As seen in Figure 5.9, the time to rupture in PC material extracted from Header A is on average around three times smaller than the AR material, for a given value of true stress. Further seen in this figure is that the trend of the rupture time for PC material extracted from different headers are almost indistinguishable, whereas the AR material shows a noticeable cast to cast variability in the rupture life behaviour. More number of tests is needed to be performed on the PC material from different headers to confirm this observation.

The average creep strain rate behaviour of the PC and AR materials from different headers have been shown in Figure 5.10 and the power law constants are determined and given in Table 5.5. The axial creep displacement measurements are employed in all average creep strain rate calculations. As seen in Table 5.5, the stress exponent n_A found for the PC material extracted from Header A is significantly larger than that of obtained from the AR material in the same header. However, considering the possible uncertainties and the inherent scatter in the axial creep ductility measurements (which subsequently affect the average creep strain rate values), the slope of the regression fit to the AR data points from Header B&C was used to work out the mean fit to the PC material from different headers. Figure 5.10 shows that similar average creep strain rate trends are observed between the AR and PC materials extracted from Header A, when the slopes are considered the same ($n_A = 8.7$ and $A_A = 9.00 \times 10^{-27} \text{ MPa}^{-n_A} \text{ h}^{-1}$ for the PC material from Header A). The available data in Figure 5.10 indicate that the cast to cast variability is less significant for the PC material compared to the AR. However, more number of tests on the PC material is needed to study the cast to cast variability of the PC material in more details.

5.4 Comparison of the Creep Ductility Trends in PC and AR Materials

The plasticity effects are expected to have a significant influence on the creep deformation behaviour of Type 316H SS at elevated temperatures. As noted in Section 5.3, no particular trend was observed in the creep ductility data on the AR material from the tests performed under the stress range of 257–300 MPa (which is significantly higher than the yield stress of the AR material at 550 °C) possibly because of high stresses and thus strong plasticity effects. However, some stress dependency was apparent in the PC data under similar range of applied stress (which is close to or just above the yield stress of the PC material at 550 °C). The plasticity effects on the creep ductility trends of the AR and PC materials are investigated further for a wide stress range in this section. Furthermore, due to the lack of data on the AR material at stresses lower than $\sigma < \sigma_y$ at 550 °C, a new approach has been proposed to estimate the creep ductility trends of the AR material using the data available in the open literature at higher temperatures. The estimated trends have been employed to examine the plasticity effects on the creep ductility of the material over a wide stress range.

5.4.1 Effect of Plasticity on Creep

Type 316H stainless steel has a relatively low yield stress of around 170 MPa at 550 °C, thus high stress accelerated creep tests may introduce significant levels of plasticity into the material. It is known that plasticity has a significant influence on the creep failure behaviour of 316H stainless steel, especially the creep ductility [1, 2, 50, 89, 112]. The level of plastic strain in the material can be indicated by the ratio of the applied stress to the yield stress which is taken as the 0.2% proof stress, $\sigma/\sigma_{0.2}$, for a uniaxial specimen. For fractured geometries, the level of plasticity may be indicated by the normalised reference stress, $\sigma_{ref}/\sigma_{0.2}$ [50]. In order to take into account the influence of plasticity on the creep behaviour of the material, the creep ductility data for the PC and AR materials presented in Section 5.3 are correlated with the applied stress normalised by the yield stress at 550 °C ($\sigma_{0.2} = 259$ MPa for PC material and $\sigma_{0.2} \approx 170$ MPa for AR material) in Figure 5.11. To examine the creep behaviour of the PC material in a wider range of applied stresses, the available PC data are compared to a large data set for the AR material at 550 °C, provided by EDF Energy, in Figure 5.11. Note that the AR data were from a range of different casts and headers and may be treated as a single data set.

As seen in Figure 5.11, no particular trend within a wide scatter can be observed in the creep ductility data available on the AR material at 550 °C. However, a clear stress dependent trend can be inferred to the PC data available at this temperature. Figure 5.11 shows that the available AR data set is corresponding to the stresses larger than yield ($\sigma/\sigma_{0.2} > 1$). However, to examine the plasticity effects on the creep rupture behaviour of the AR and PC materials over a wider stress range, the creep ductility trends have been investigated further for lower stress levels where there is no data available in the literature. The AR creep ductility data available at higher temperatures have been employed in the next section to estimate the creep ductility trends of the AR material at 550 °C over a wide stress range.

5.4.2 The Variation of Creep Ductility over a Wide Stress Range

5.4.2.1 Existing Estimation Approaches

A significant number of relatively high stress (short term) uniaxial creep test results are available in the literature for 316H at 550 °C, however the long term data set available for this material is limited due to the cost and practicalities of long term testing [113, 114]. Simple data extrapolation from short term, high load creep tests to long term, lower load creep tests is generally not considered appropriate due to materials microstructural evolutions that can occur as a result of *e.g.* thermal aging effects [115], which may be indicated by changes in the hardness data in low stress crept materials [116, 117]. Therefore, prediction models are often employed to estimate the creep properties at long times (low stresses) where experimental data are generally unavailable.

An extrapolation method has been developed by Wilshire in [116] to estimate the creep life (time to rupture) of various steels. In Wilshire's approach the applied stress is normalised by the UTS and the activation energy for lattice diffusions is quantified. This enables the rupture properties at a range of temperatures to be superimposed and hence a master curve to be generated. The Wilshire's approach has been successfully applied to a range of steels including bainitic 1Cr-1Mo-0.25V rotor forgings [117], polycrystalline copper [118], martensitic 9-12% chromium grades [116, 119] and ferritic 1Cr-0.5Mo [115]. However, it has been shown by Evans [120] that the general Wilshire's model fails to predict the long term (100,000 hours or over) creep life of austenitic Type 316H stainless steel using shorter term data (5000 hours or shorter). Therefore a modified approach has been proposed in [120] to predict the long term rupture time in 316H over a range of stresses and temperatures. This model has been validated for the short term and long term rupture life data, up to around 200,000 hours

where the experimental data is available, for temperatures in the range of 600–850 °C. However, due to limited data availability at 550 °C no validation of the model at this temperature has been presented in [120]. Though these models can successfully predict the creep rupture time of the material for a range of stresses and time scales, they cannot be used to predict the creep ductility of the material. An independent study on the creep properties of 316H at 550 °C was conducted in [112] and constitutive equations were proposed. Due to the limited data available at the time of study, significant assumptions were made in [112] about the data trends beyond the stress range where data were available (data were only available for stresses close to or above yield which corresponds to relatively short term tests).

Relatively short and long term test data for 316H are provided by NIMS at 750, 700, 650 and 600 °C [114] and data for 550 °C are available and have been provided by EDF Energy (see *e.g.* [51]). However, at 550 °C the longest test duration available in literature at various temperatures is around 100,000 hours which corresponds to stresses above the yield stress of the material. In order to characterise the creep failure behaviour of 316H at 550 °C further longer term test data are required. The creep ductility data from higher temperatures are examined next and the trends observed are used to estimate the long term (relatively low stress) behaviour of creep ductility at 550 °C.

5.4.2.2 Influence of the Normalised Stress on Creep Ductility

The axial creep ductility has been plotted against stress normalised by the appropriate proof stress, (in log-log axes) as shown in Figure 5.12(a)–(e) at temperatures of 750, 700, 650, 600 and 550 °C. Note that the data for 600–750 °C taken from NIMS [114] and also the data provided by EDF Energy [51] for 550 °C, are from a number of different batches of material which were subjected to different heat treatments. However, within the range of data scatter they may be treated as one data set as no trends were observed between the different batches. As an indication of the influence of temperature on the 0.2% proof stress, the average value of $\sigma_{0.2}$ from different batches at a given temperature are summarised in Table 5.6, and an estimation of the data scatter is provided by calculating twice the standard deviation (2SD).

As seen in Figure 5.12(b) and Figure 5.12(c), creep ductility data are available for a wide stress range of around $0.1 < \sigma/\sigma_{0.2} < 2$ at 700 °C and 650 °C, respectively. Three regions may be observed in these figures. At high and very low normalised stress, the creep ductility may be assumed independent of stress. These two regions have been

identified by observation and are here denoted the upper shelf (US) and lower shelf (LS) creep ductility region, respectively. However, a clear power law trend can be observed between ϵ_f and $\sigma/\sigma_{0.2}$ at intermediate stress ranges. The average values of the creep ductility in the upper shelf and lower shelf regions at each temperature have been calculated, and shown in Table 5.6 and Figure 5.12. Note that due to significant data scatter there is some error associated with the fits made, as indicated by the two standard deviation (2SD) values quoted in Table 5.6. As shown in Table 5.6, the average lower shelf (low stress) creep ductility has been found to be approximately a factor of around 14 less than that at high stresses for the data available at 700 °C and 650 °C.

Similar trends can also be inferred at 750 °C in Figure 5.12(a) and 600 °C in Figure 5.12(d) though the upper shelf and lower shelf data, respectively, are not available at these temperatures. Assuming a constant slope for the stress dependent transition region at each temperature, a line of best fit has been plotted for data points in the intermediate stress levels to give some indication of the trends in this region. This line was generated using the average of the individual slopes calculated from a regression fit to the data in the transition region at each temperature.

Since only high stress data ($\sigma/\sigma_{0.2} \geq 1$) are available at 550 °C, which is the temperature of greatest interest for assessing the life times of industrial components, only the upper shelf creep ductility can be defined at this temperature. Therefore, in order to obtain an estimate of the long term, low stress creep ductility value at 550 °C, the trends at higher temperatures are next examined further and related to the data at 550 °C.

5.4.2.3 Creep Ductility Trends at 550 °C

A trend has been seen in the transition points between the upper shelf ductility and the stress dependent region in Figure 5.12(a)–(d) at various temperatures and similarly between the lower shelf ductility and the stress dependent region. These trend lines are shown as two dashed lines in Figure 5.13 together with the creep ductility data from Figure 5.12(c)–(e). These lines have been extrapolated to estimate the transition points at 600 °C and 550 °C, as shown in Figure 5.13. The lower shelf creep ductility at 550 °C has been estimated by assuming that the difference between lower shelf creep ductility at 650 °C and 550 °C is the same as the difference between the upper shelf ductilities at 650 °C and 550 °C. This is considered a reasonable assumption as a similar shift in both the upper and lower shelf creep ductilities at 700 °C and 650 °C are observed (see Table 5.6). A line to describe the stress dependent region of the creep

ductility data at 550 °C may subsequently be generated. The same approach has been employed to estimate the lower shelf ductility at 600 °C. The lower shelf creep ductility at 550 °C is estimated to be 0.9%, which is consistent with minimum creep ductility reported in [24]. Not shown here but given in [121] is that similar values of the rupture time ν_r and average creep strain rate n_A power law constants have been estimated for 316H at 550 °C in the stress ranges corresponding to the US and LS creep ductility regions. This confirms that the creep ductility of the material can be considered stress independent in the US and LS regions. More details of the creep ductility estimation approach for a wide stress and temperature range can be found in [121].

5.4.3 Comparison of the PC Creep Data to the Estimated AR Trend

The available creep data on the PC material are compared to the estimated creep ductility trends for the AR material at 550 °C in Figure 5.14. As seen in this figure, the data points from the PC material follow the estimated trend of the AR material in the intermediate stress dependent region when the creep ductility data are plotted against the normalised stress. In other words, although the creep ductility of the PC material shows a significant drop compared to AR material under the same level of applied stress, but normalising the applied stress by the yield strength allows taking the hardening effects into account and provides similar trends between AR and PC materials for a given value of normalised stress.

5.5 Discussion

It can be seen in Figure 5.10 that the average creep strain rate trends for the PC and AR materials from Header A are very similar. This implies that under the range of stresses examine, the time to rupture and the creep ductility of the PC material are reduced by similar factors which result in the average creep strain rate to remain almost unchanged.

It was shown in Figure 5.14 that the creep ductility data for the PC material fall close to the trends estimated for the AR material at 550 °C, when the applies stresses are normalised by the yield stress. As seen in Figure 5.14, the PC data were only available at intermediate stress range where the creep ductility trends are stress dependent. However, Figure 5.14 suggests that similar agreement between the creep ductility of the PC and AR materials may also be obtained at high stress (US) and low stress (LS) regions. In other words, Figure 5.14 indicates that the creep ductility trends of the AR material at 550 °C for a wide stress range may be estimated by performing tests on the PC material in much shorter timescales of on average around three times faster. Also

similar average creep strain rate trends observed for the PC and AR material in Figure 5.10 suggest that the short term and long term creep power law constants of the AR material (which have been estimated at 550 °C in [121]) may be approximated by performing tests on the PC material. More experiments need to be performed on the PC material at higher and lower stress levels (corresponding to upper and lower shelves) in future work to confirm these trends.

5.6 Summary

- The pre-compression process hardens the material and generates less evident DSA effects compared to the AR material in the high temperature tensile tests at 550 °C.
- A large fraction of primary creep region is observed when the applied stress on the PC material is relatively low at 550 °C. However, due to the increase in the dislocation density at higher stresses ($\sigma \gg \sigma_y$), the percentage ratio of the primary creep strain to the creep ductility decreases.
- Some differences are observed in the power law coefficients worked out for the time to rupture, minimum and average creep strain rate of the AR material from different casts whereas similar power law constants have been found in PC material taken from different headers. This suggests that cast to cast variability may be considered less significant for the PC material.
- Pre-compression process has a small effect on the minimum and average creep strain rate of the material, however the time to rupture in PC material is found on average three times shorter compared to the AR, for a given value of stress.
- The short term and long term uniaxial creep ductility and the power law creep constants of the AR material at 550 °C may be estimated over a wide stress range by performing tests on the PC material, in much shorter time scales.

5.7 Tables

Table 5.1: Tensile properties for as-received and 8% pre-compressed material at room temperature and 550 °C

Material Condition	Temperature	E (GPa)	$\sigma_{0.2\%}$ (MPa)	$\sigma_{0.5\%}$ (MPa)	UTS (MPa)	ϵ_f^{eng} Axial (%)	ϵ_f^{eng} ROA (%)
AR	Room	205	313	336	603	62.99	68.27
8% PC		217	256	347	604	60.38	65.30
AR	550 °C	140	177	199	432	46.70	58.45
8% PC		140	259	296	441	40.24	50.75

Table 5.2: Ramberg Osgood material model constants from a single fit to the tensile data at 550 °C

Material	AR	PC
N	3	5
α	6.42	3.27
ϵ_{p0}	1.21×10^{-3}	1.86×10^{-3}
σ_{p0} (MPa)	177	259
A_p (MPa ^{-N})	1.59×10^{-9}	5.11×10^{-15}

Table 5.3: Comparison of failure creep strain and load up strain measurements in AR and PC specimens extracted from Header A

Test ID	Material Condition	σ_{nom} (MPa)	σ_{true} (MPa)	ϵ_{load}^{eng} (%)	ϵ_{load}^{true} (%)	$\epsilon_{p,load}^{eng}$ (%)	ϵ_f^{eng} Axial (%)	ϵ_f^{true} Axial (%)	ϵ_f^{eng} ROA (%)	ϵ_f^{true} ROA (%)
A2-1	8% PC	300	301	0.24	0.24	0.03	3.41	3.35	23.04	26.20
A2-2	8% PC	280	281	0.28	0.28	0.07	2.93	2.89	11.33	12.04
A2-3	8% PC	270	271	0.25	0.25	0.05	3.07	3.02	8.62	9.02
A2-4	8% PC	257	258	0.19	0.19	0.01	1.20	1.19	4.25	4.34
A1-1	AR	335	356	6.20	6.02	5.96	12.14	11.46	17.56	20.79
A1-2	AR	300	313	4.35	4.26	4.07	13.39	12.56	21.35	25.34
A1-3	AR	290	304	4.75	4.64	4.60	10.97	10.41	22.73	27.27
A1-4*	AR	280	293	4.42	4.33	4.18	-	-	-	-
A1-5	AR	257	264	2.90	2.86	2.69	14.99	13.97	29.24	35.82

* The test was stopped prematurely prior to failure

Table 5.4: Relative duration and strain accumulation in the primary, secondary and tertiary creep regions as a percentage of the test failure time, t_r , and failure creep strain, ϵ_f^{eng}

Test ID	Material Condition	σ_{nom} (MPa)	ϵ_f^{eng} Axial (%)	ϵ_{Pri}^c (%)	ϵ_{Sec}^c (%)	ϵ_{Ter}^c (%)	t_r (h)	t_{Pri} (%)	t_{Sec} (%)	t_{Ter} (%)
A2-1	8% PC	300	3.41	3.18	26.11	70.70	760	3.78	70.42	25.80
A2-2	8% PC	280	2.93	8.88	42.06	49.07	2275	7.45	83.14	9.41
A2-3	8% PC	270	3.07	9.81	33.36	56.83	1632	8.81	69.13	22.06
A2-4	8% PC	257	1.20	30.45	44.44	25.11	2971	20.20	64.82	14.98
A1-1	AR	335	12.14	17.93	44.07	38.00	450	12.67	61.11	26.22
A1-2	AR	300	13.39	10.09	30.28	59.63	1791	11.06	51.09	37.86
A1-3	AR	290	10.97	19.58	17.79	65.70	4291	19.58	38.22	42.20
A1-4*	AR	280	-	-	-	-	1935	-	-	-
A1-5	AR	257	14.99	18.46	10.77	70.77	5658	19.65	33.56	46.79

* The test was stopped prematurely prior to final failure

Table 5.5: Creep strain rate and rupture power law constants based on the true stress

Material Condition	A (MPa ⁻ⁿ h ⁻¹)	n	σ_0 (MPa)	A_A (MPa ^{-n_A} h ⁻¹)	n_A	σ_{0_A} (MPa)	B_r (MPa ^{v_r} h)	v_r
AR-Header A	1.15×10^{-30}	10.2	840	1.11×10^{-26}	8.7	1052	2.50×10^{31}	11.3
AR-Header B and C	5.28×10^{-31}	10.2	930	4.49×10^{-27}	8.7	1068	5.27×10^{31}	11.3
8% PC-Header A	2.22×10^{-32}	10.9	802	1.09×10^{-39}	14.0	607	5.04×10^{22}	8.0
8% PC-Header C	2.50×10^{-32}	10.9	793	1.30×10^{-39}	14.0	599	6.50×10^{22}	8.0

Table 5.6: Proof stress and mean creep ductility values for range of temperatures

Temperature (°C)	$\sigma_{0.2}$ (MPa)	Upper shelf creep ductility (%)		Lower shelf creep ductility (%)	
		Average value	×/÷ 2SD	Average value	×/÷ 2SD
750	133	147.0*	-	10.5	1.3
700	136	77.3	1.5	5.5	1.7
650	140	64.2	1.4	4.5	1.8
600	150	46.0	1.5	3.2*	-
550	170	13.6	4.3	0.9*	-

*Estimated value

5.8 Figures

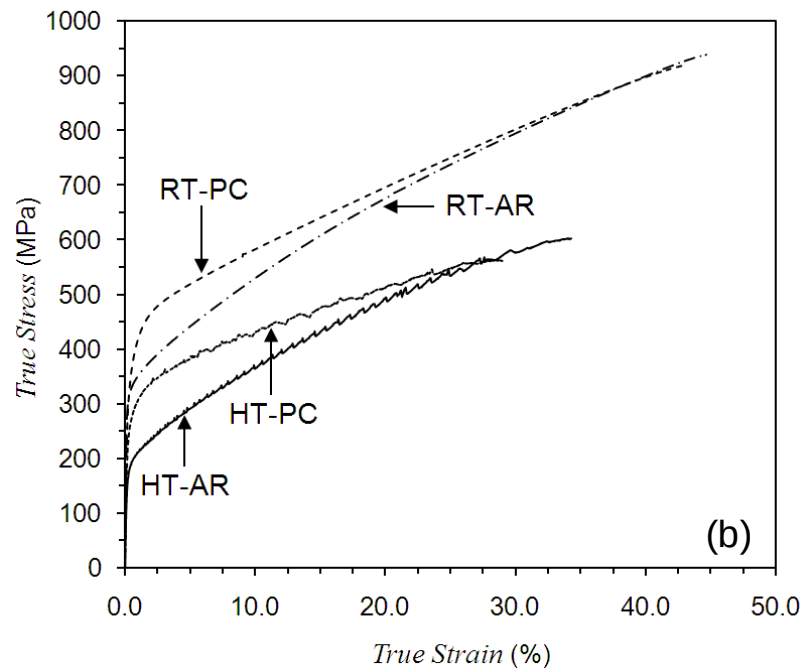
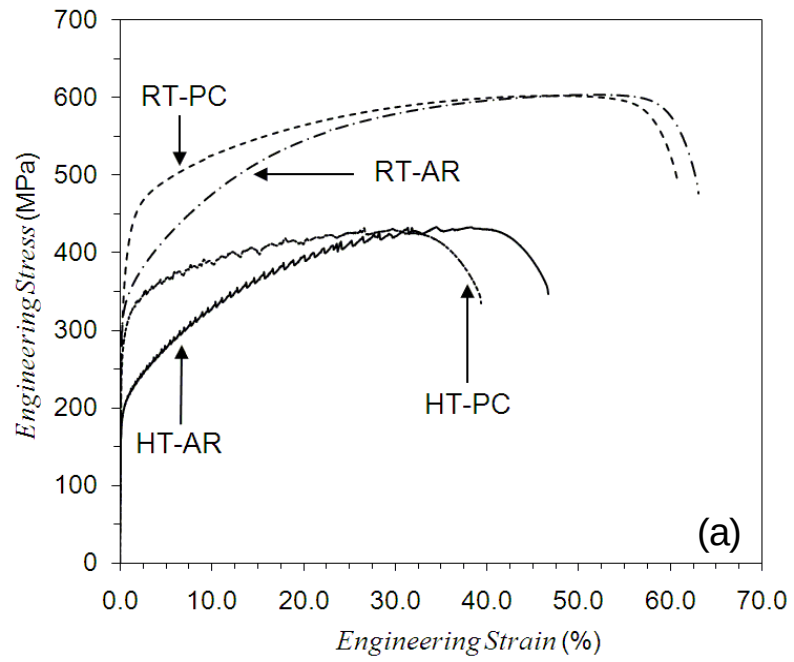


Figure 5.1: Room temperature (RT) and high temperature (HT) tensile responses of as-received (AR) and pre-compressed (PC) materials based on (a) engineering, and (b) true stress-strain

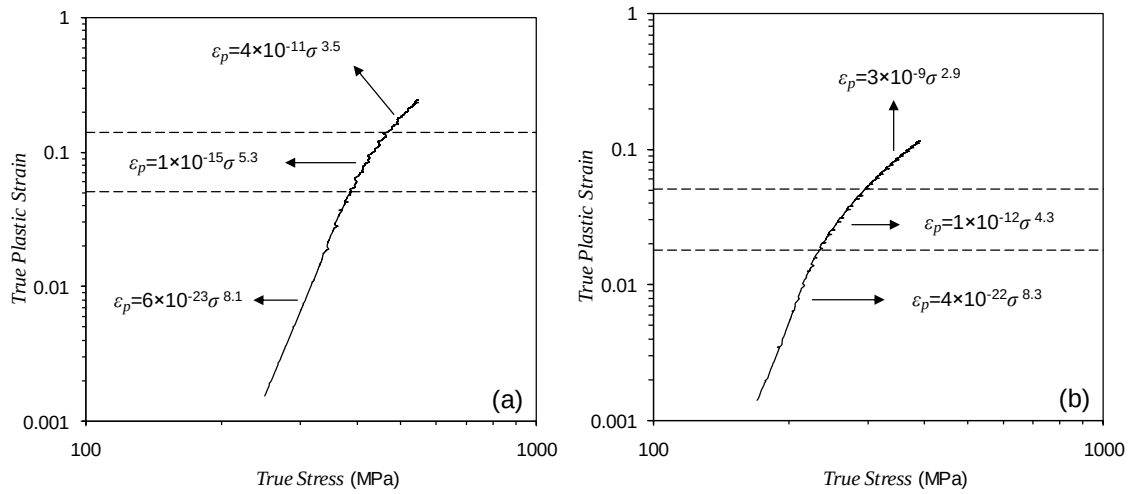


Figure 5.2: Power law fits to the plastic strain data for (a) PC (b) AR material

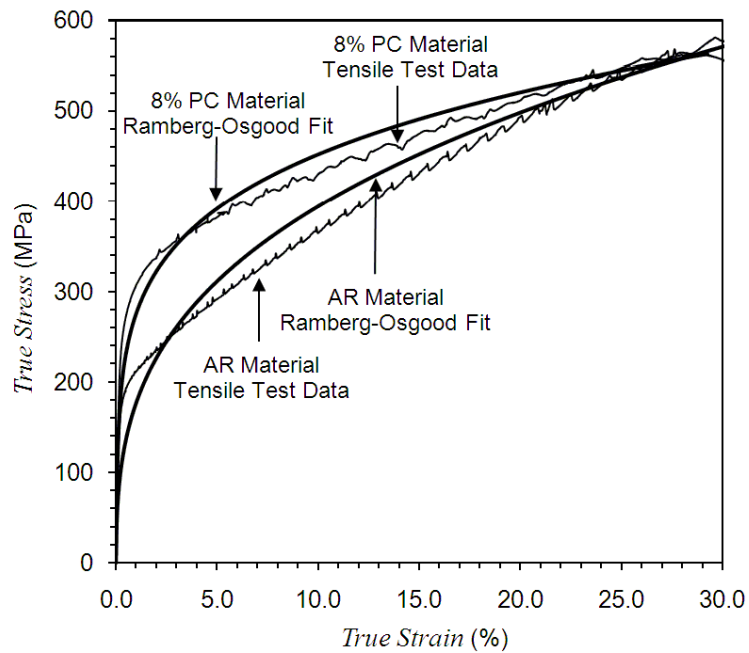


Figure 5.3: Ramberg-Osgood Fits to the high temperature tensile data on AR and PC materials

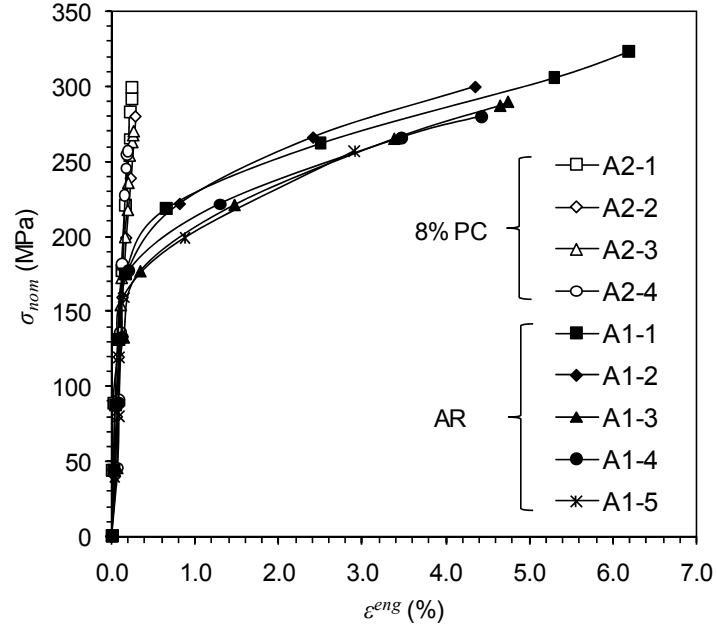


Figure 5.4: Comparison of load up curves in AR and PC materials

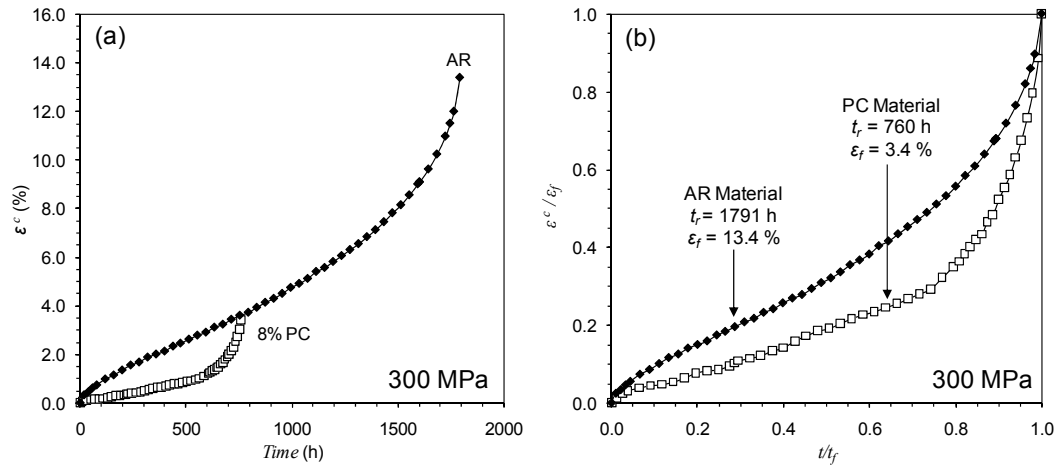


Figure 5.5: Comparison of uniaxial creep behaviour under 300 MPa applied stress shown in (a) actual form (b) normalised form

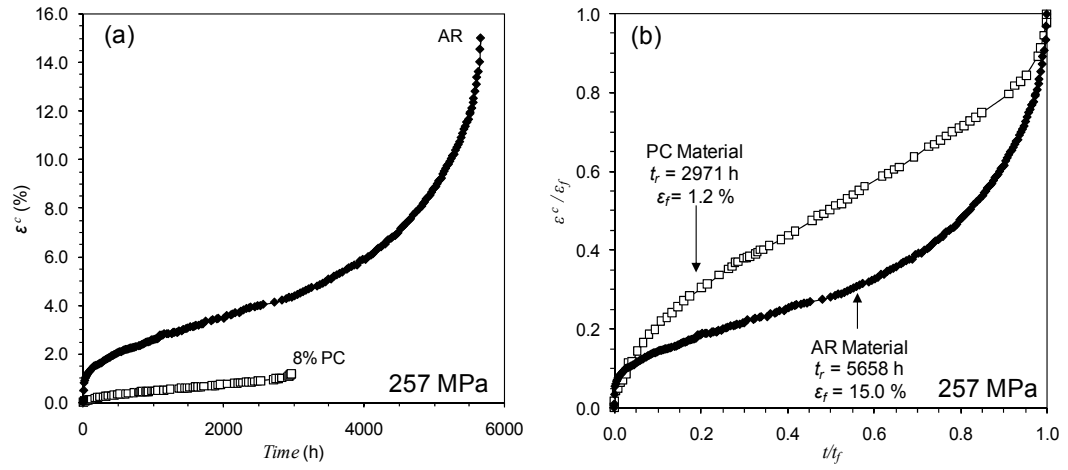


Figure 5.6: Comparison of uniaxial creep behaviour under 257 MPa applied stress shown in (a) creep strain vs. time form (b) normalised form

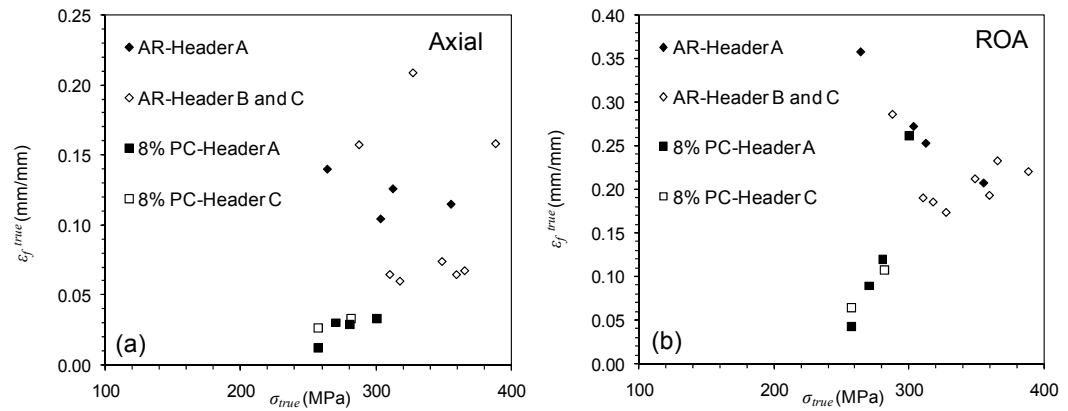


Figure 5.7: True creep strain at failure against true stress by (a) axial measurements (b) reduction of area calculations

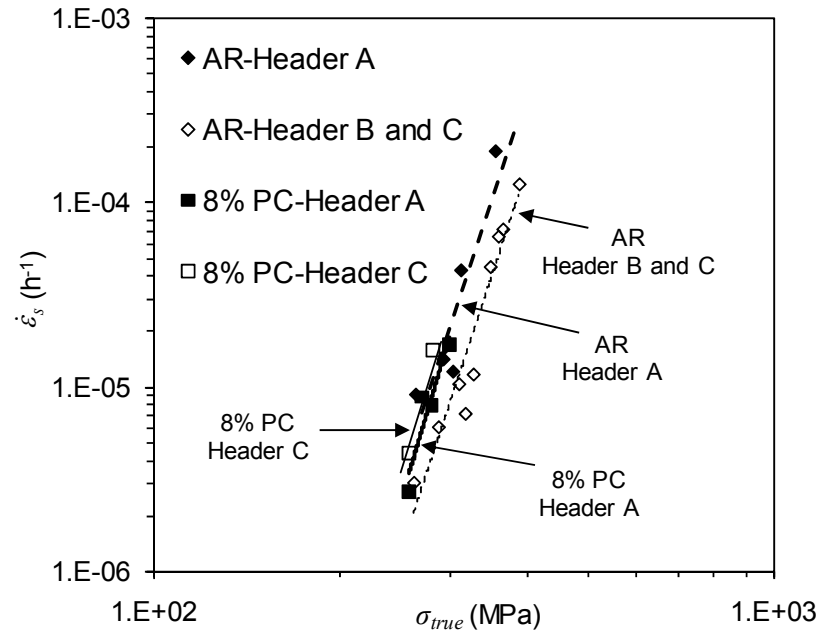


Figure 5.8: Comparison of the minimum creep strain rate behaviour of AR and PC materials in different headers

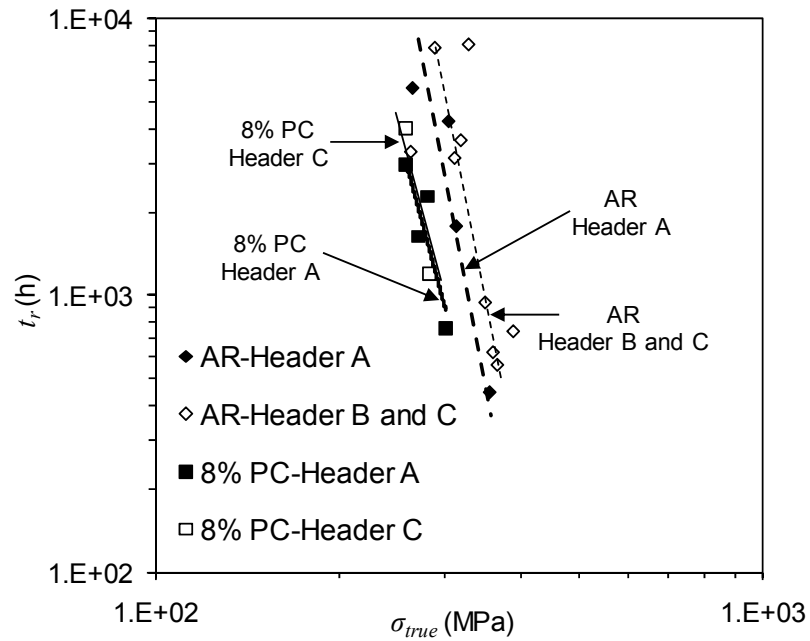


Figure 5.9: Comparison of the time to rupture between AR and PC materials extracted from different headers

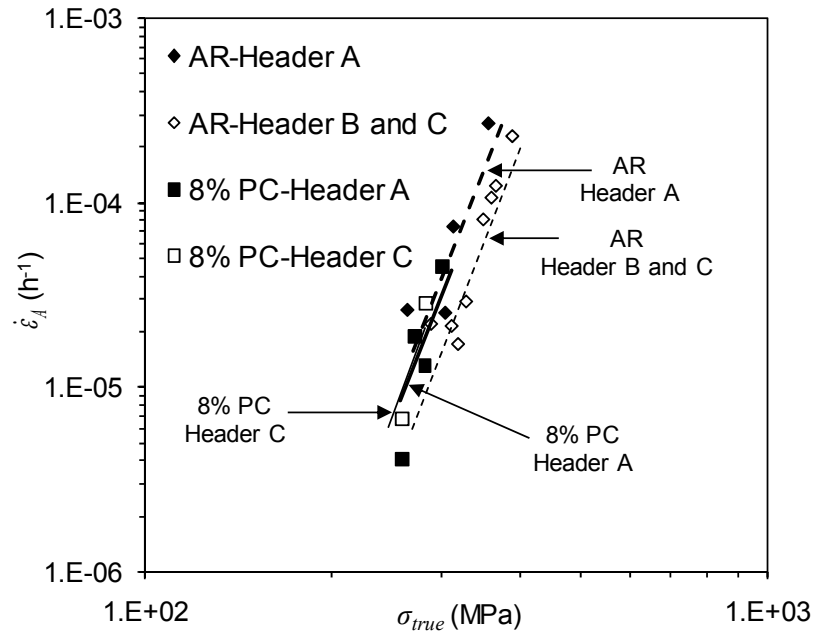


Figure 5.10: Comparison of the average creep strain rate behaviour of AR and PC materials in different headers

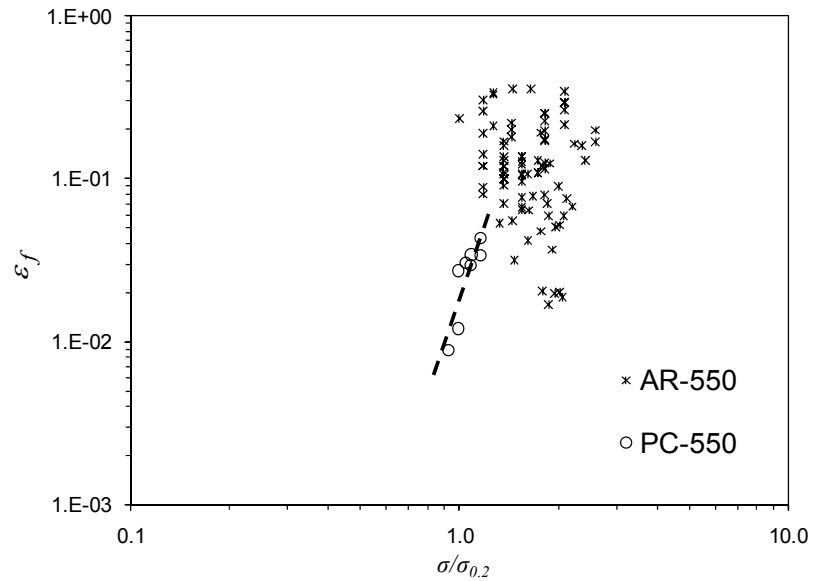


Figure 5.11: Comparison of the creep ductility of PC material to the AR data at 550 °C

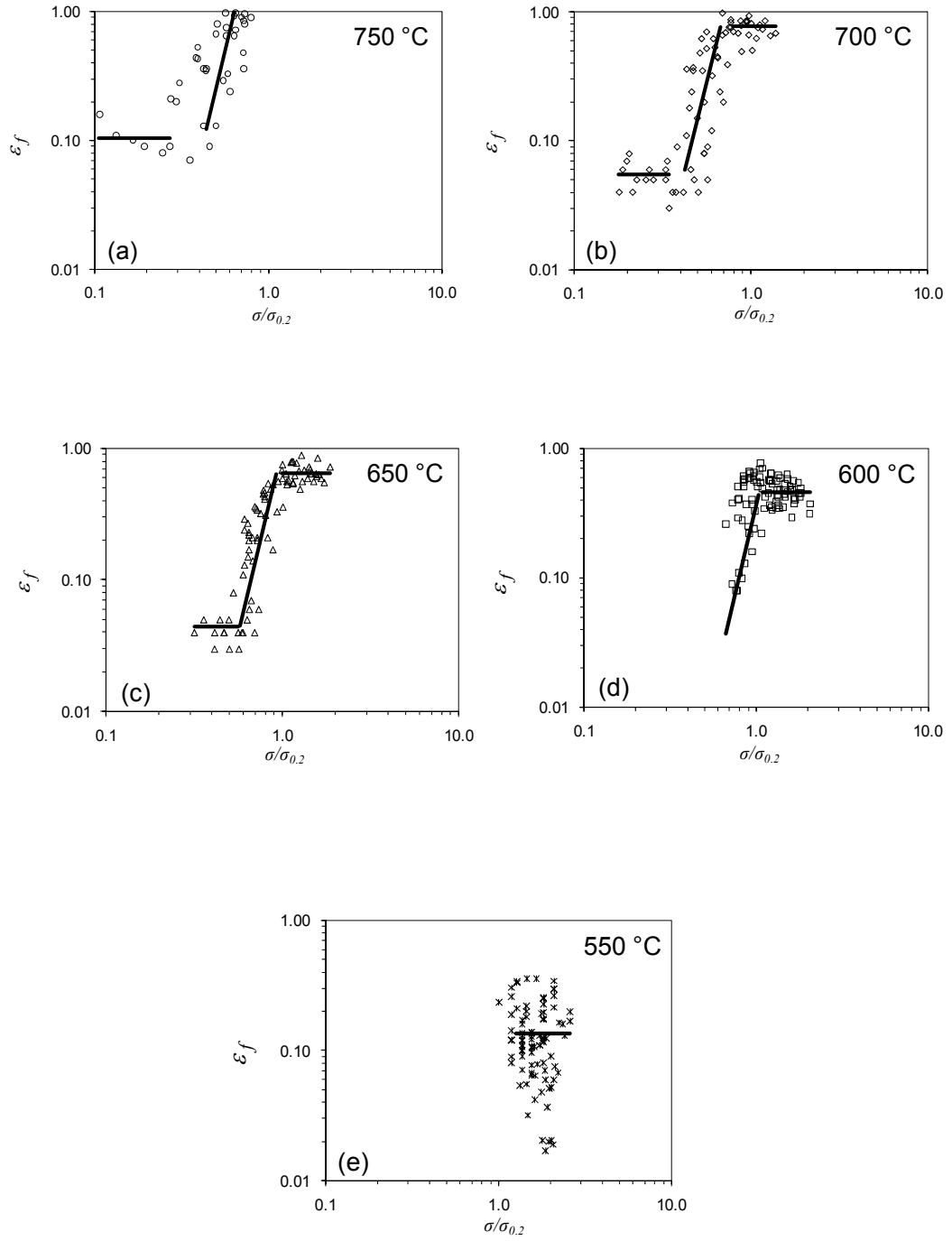


Figure 5.12: Variation of creep ductility against the normalised stress at (a) 750 (b) 700 (c) 650 (d) 600 and (e) 550 °C.

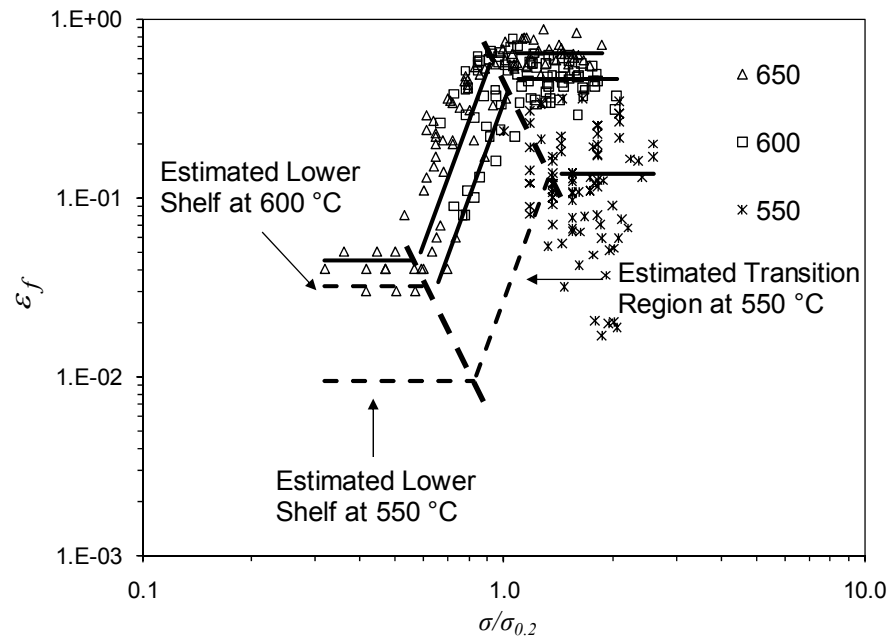


Figure 5.13: Estimation of the lower shelf creep ductility at 600°C and 550°C .

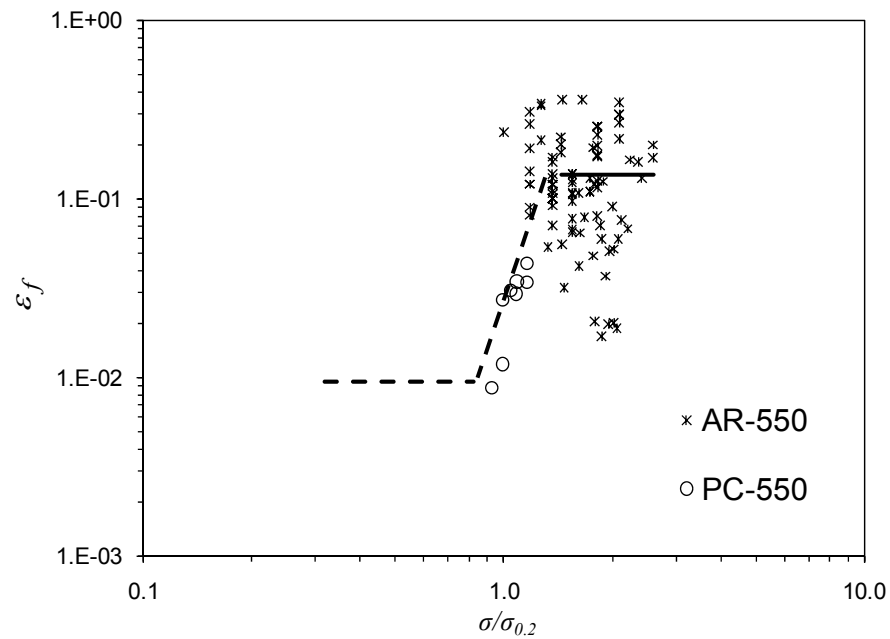


Figure 5.14: Comparison of the creep strain at failure in PC material to the estimated trends for AR material at 550°C

Chapter 6

Pre-Compression Effects on Creep Crack Initiation and Growth Behaviour

6.1 Introduction

In order to calibrate the creep crack growth (CCG) behaviour of the LCD specimens (which are made of PC material) and also to choose an appropriate load for these tests, three tests with different initial loading conditions were run on compact tension, C(T), samples nominally identical to LCD specimens with $a_0/W = 0.35$ (where a_0 is the initial crack length and W is the width of the specimen). The data from these calibration experiments were used to interrupt CCG tests on LCD samples at the approximate crack length of $a/W = 0.5$. In addition, a test was conducted on a standard C(T) specimen with the initial normalised crack length of $a_0/W = 0.5$ made of PC material extracted from the same header. The results from this test have been compared to those of obtained from the CCG calibration tests.

It has been shown in previous works that the pre-compression process influences the tensile, uniaxial creep rupture and CCG behaviour of the material [1, 89]. To maximise the information which can be obtained from the CCG tests performed in this work, the results from these tests are compared to the existing data on PC and AR materials from different cast and headers available in [89].

In order to analyse the CCG data, the elastic (K) and plastic/creep (J or C^*) fracture mechanics parameters need to be quantified for different crack lengths. The general solutions for these parameters are given in the literature for $a/W > 0.45$. However, the initial normalised crack length in LCD specimens is $a_0/W = 0.35$. Therefore numerical simulations have been performed to find out the accurate solutions of the fracture mechanics parameters, employed in the data analysis, for the range of normalised crack lengths $0.35 \leq a/W \leq 0.45$.

This chapter starts with a numerical study of the fracture mechanics parameters for short cracks followed by the comparison of the creep crack initiation (CCI) and CCG

behaviour of PC to AR material. Note that the PC data presented in this chapter are obtained from the specimens with the loading axis parallel to the pre-straining direction.

6.2 Evaluation of the Fracture Mechanics Parameters for $0.35 \leq a/W \leq 0.45$

In order to analyse the CCG test data, accurate solutions of the fracture mechanics parameters for a range of crack lengths are required. The elastic parameters which are used in the data analysis of the creep tests are the shape function, Y , and the elastic compliance, C_e . The general solutions for Y and C_e parameters are given in [60], however these solutions are only valid for the normalised crack lengths $a/W > 0.45$. Therefore FE simulations have been performed to determine the solutions of these parameters for $0.35 \leq a/W \leq 0.45$. The general solution of the η parameter, which is used in the experimental calculations of the fracture mechanics parameters J and C^* , are given in [67] for a range of cracked bodies. For the case of compact tension specimen, the solutions given in [67] are only valid for $0.45 \leq a/W \leq 0.7$. To calculate the plastic J and creep C^* fracture mechanics parameters accurately, FE simulations have been conducted to examine the numerical solutions of η for $0.35 \leq a/W \leq 0.45$ and the influence of N , which is the power law plastic hardening exponent, and stress state dependency have been investigated.

The numerical solutions of Y , C_e and η parameters are determined for a C(T) specimen with $a/W \leq 0.45$. Two dimensional (2D) symmetrical C(T) model has been designed in ABAQUS [122] with a focused mesh at the crack tip region which provides accurate solutions of the contour integrals in the numerical simulations. An example of the focused mesh in an elastic simulation, before and after deformation, is given in Figure 6.1. In order to provide a sharp crack in elastic simulations, the crack tip has been defined using a single node. However in plastic simulations, multiple nodes have been used for the crack tip to describe the blunting effects. In all simulations, the load point displacement (LPD which is also referred to as LLD) is compared to CMOD along the load line (see Figure 6.2). The samples are loaded under tension using a concentrated load applied on a node located at the centre of the pin position. The local deformation of the node, at which the load is applied, has been minimised by assigning elastic properties to the elements in the pin region. Simulations have been run under both PS and PE conditions to study the stress state effects on the numerical solutions. The range of a/W examined is from 0.35 to 0.45 in increments of 0.025 and small strain assumption has been employed in all simulations.

6.2.1 Shape Function, Y

It has been claimed in [123] that the general solution of the shape function given by Eqn (2.66), which is also known as Tada solution, has 0.5% accuracy for $a/W > 0.2$. However numerical simulations have been performed to evaluate the accurate solutions of the shape function for $0.35 \leq a/W \leq 0.45$. FE simulations using elastic material properties are conducted and the elastic fracture mechanics parameter, K , is calculated for different crack lengths using 40 contours around the crack tip, in the focused mesh area. The numerical values of the shape function are calculated using Eqn (2.64).

The numerical solutions of the shape function for $0.35 \leq a/W \leq 0.45$ are shown in Figure 6.3 and compared to the empirical solutions given by Eqn (2.66). The percentage of error in the shape function values calculated by Eqn (2.66) is given in Table 6.1. This table shows that the shape function values obtained using Eqn (2.66) encounter percentage error of around 2% for short cracks. Though relatively accurate values, with 0.2% error, are obtained for $a/W = 0.45$. Thus, it can be concluded that Eqn (2.66) provides accurate solutions for $a/W > 0.45$, however the numerical solution of the shape function for $0.35 \leq a/W \leq 0.45$ can be obtained by a regression fit to the FE data and defined as

$$Y(a/W) = 5.03 + 44.83(a/W) - 135.73(a/W)^2 + 161.74(a/W)^3 \quad (6.1)$$

Further seen in Figure 6.3 is a comparison between the obtained solutions of the shape function from FE simulations and the numerical calculations available in [124, 125] for a standard C(T) specimen with $LPD = 37.5$ mm. As seen in this figure relatively good agreement is found between FE trends and the numerical results available in the literature.

6.2.2 Elastic Compliance, C_e

The elastic displacement along the load line in a cracked body is defined as

$$\Delta_e = C_e(a)P \quad (6.2)$$

where P is the applied load and $C_e(a)$ is the elastic compliance at crack length a . The empirical solution of the elastic compliance for a standard compact tension specimen with the normalised crack length in the range of $0.2 \leq a/W \leq 0.95$ is given in non-dimensional form as [126]

$$C_e(BE') = \left(\frac{1+a/W}{1-a/W} \right)^2 \left[\begin{aligned} &2.1630 + 12.219(a/W) - 20.065(a/W)^2 \\ &- 0.9925(a/W)^3 + 20.609(a/W)^4 - 9.9314(a/W)^5 \end{aligned} \right] \quad (6.3)$$

where B is the specimen thickness, E' is the effective Young's modulus (which is equal to E for plane stress and $E/(1-\nu^2)$ for plane strain conditions). It has been claimed in [126] that Eqn (6.3) provides solutions with 0.5% accuracy in the range of crack lengths specified.

In order to find out the accurate solutions of the elastic compliance for the normalised crack lengths in the range of $0.35 \leq a/W \leq 0.45$, numerical simulations have been performed on C(T) specimens with different crack lengths and elastic properties (*i.e.* $E = 140$ GPa, $\nu = 0.3$) are assigned to the elements. The variation of displacement along the load line at different crack lengths showed that the elastic displacement at the load point is identical to what is observed at the CMOD along the load line. This means that under a constant load, the elastic compliance calculated based on the LLD is the same as CMOD (*i.e.* $C_e^{LLD}(a) = C_e^{CMOD}(a)$) for the range of crack lengths examined. Therefore the numerical elastic compliances obtained from CMOD are presented here and the results are assumed to be the same as LLD. The equality of the LLD and CMOD elastic compliances is observed under both plane stress and plane strain conditions.

The non-dimensional elastic compliance solutions obtained from numerical simulations under PS and PE conditions are shown in Figure 6.4(a) and compared to the values given by Eqn (6.3) for a wide range of crack lengths. This graph shows that the FE solutions of the elastic compliance for the normalised crack lengths in the range of $0.35 \leq a/W \leq 0.70$ are identical under PS and PE conditions, when the non-dimensional values are compared. A relatively good agreement between the values given by Eqn (6.3) and the FE solutions are seen in this figure for $a/W > 0.45$, however some discrepancy is observed for shorter cracks. This means that Eqn (6.3) provides valid solutions of the elastic compliance only for $a/W > 0.45$. The percentage of error between the numerical and analytical solutions is shown in Table 6.2. As seen in this table large percentage of error is encountered when the elastic compliance for a short crack is calculated using Eqn (6.3). A more detailed comparison between the numerical and analytical solutions of the elastic compliance for shorter cracks in the range of $0.35 \leq a/W \leq 0.45$ is shown in Figure 6.4(b). This figure shows that there are some considerable differences (4–6% error) between the values of the elastic compliance given by Eqn (6.3) and the numerical solutions, for the range of crack lengths examined. The numerical solution of the elastic compliance for $0.35 \leq a/W \leq 0.45$

obtained using a regression fit to the FE data is shown in Figure 6.4(b) and can be expressed as

$$C_e(BE') = \left[-16.29 + 174.73(a/W) - 427.38(a/W)^2 + 572.31(a/W)^3 \right] \quad (6.4)$$

6.2.3 Eta Factor, η

The general solution of the η parameter for compact tension specimens with the normalised crack lengths in the range of $0.45 \leq a/W \leq 0.7$ and $N \geq 3$ is given in [67] as $\eta^{LLD} = \eta^{CMOD} = 2.2 \pm 0.1$ under both plane stress and plane strain conditions. In order to determine the numerical solutions of the η parameter for shorter cracks in C(T) specimens, non-linear FE simulations have been performed under plane stress and plane strain conditions for a range of N values (*i.e.* $N = 3, 5, 6, 10$ and 20) at a few different crack lengths in the range of $0.35 \leq a/W \leq 0.45$. For consistency, the properties used in [67] (*i.e.* $E/\sigma_{p0} = 318$, $\alpha = 1$, $\nu = 0.3$, $N = 3, 5, 6, 10, 20$) have been employed. 2D symmetry C(T) model has been designed in ABAQUS and plastic analyses are performed using 4-noded quadrilateral CPS4 and CPE4H elements. Knowing that the J -integral is path independent, numerical solutions of the J parameters are evaluated by defining 40 contours (rings of elements) around the crack tip. The non-linear component of the J parameter, J_p , is calculated by subtracting the total J from the elastic J_e (calculated by Eqn (2.81)) and the numerical values of η parameter at each crack length are calculated by rearranging Eqn (2.83) and writing it as

$$\eta = \frac{J_p B_n (W - a)}{P \Delta_p H} \quad (6.5)$$

where P is the applied load, Δ_p is the plastic displacement and $H = N/(N+1)$ for a C(T) specimen. Note that the numerical solutions of the shape function given in Eqn (6.1) for $0.35 \leq a/W \leq 0.45$ have been employed to calculate the stress intensity factor and subsequently J_e , for the range of crack lengths examined. The plastic displacement in Eqn (6.5) is calculated by subtracting the total load line displacement, Δ_{Total} , from the elastic displacement, Δ_e , as

$$\Delta_p = \Delta_{Total} - \Delta_e \quad (6.6)$$

where the elastic displacement itself is given by Eqn (6.2).

Simulations have been performed for $a/W = 0.350, 0.375, 0.400, 0.425$ and 0.450 using the properties mentioned above and the solutions of η parameter are quantified

at each crack length. The general solution of the η parameter for $0.35 \leq a/W \leq 0.45$ is presented here and more detailed comparisons are given in Appendix B.

The numerical solutions of the η factor at different crack lengths in the range of $0.35 \leq a/W \leq 0.45$ for $3 \leq N \leq 20$ evaluated for plane stress conditions are shown in Figure 6.5. It can be seen in this figure that using both CMOD and LLD, the numerical values of the η parameter fall within the previously found range of 2.2 ± 0.1 . The numerical values of the η factor evaluated using LLD and CMOD under plane strain conditions by employing different values of N are presented in Figure 6.6. It can be observed in this figure that η solution has a strong dependency on the N value under plane strain conditions. Further seen in this figure is that the numerical solutions found under plane strain conditions for large values of N (*i.e.* $N = 10$ and 20) are almost identical using both LLD and CMOD. However some discrepancy between η^{LLD} and η^{CMOD} is observed for small values of N (*i.e.* $N \leq 6$). A linear regression fit to all the plane strain data points in the range of $0.35 \leq a/W \leq 0.45$, evaluated for $3 \leq N \leq 20$ using LLD and CMOD describes the general solution of the η factor under PE conditions as

$$\eta^{LLD} = \eta^{CMOD} = 0.6082 + 3.6279(a/W) \quad (6.7)$$

It can be seen in Figure 6.6 that at the crack lengths close to $a/W = 0.45$, the values calculated using Eqn (6.7) fall close to the 2.2 ± 0.1 which is the general solution of the η factor for $a/W > 0.45$. Therefore it can be assumed that Eqn (6.7) gives the general solution of η for $0.350 \leq a/W \leq 0.425$ under plane strain conditions and the general solution for larger cracks is 2.2 ± 0.1 .

Figure 6.5 and Figure 6.6 show that the numerical solution of the η factor for short cracks generally depends on the stress state, LLD/CMOD and N . The PE solutions presented in Figure 6.6 suggest choosing the values of η factor based on the N value which is to be applied in the CCG data analyses for the material of interest. Having considered the dependency of the η factor on N , the line of best fit to the data points obtained under plane strain conditions using both LLD and CMOD is plotted for $N = 10$ in Figure 6.7. The power law creep stress exponent of 316H stainless steel is close to 10 (see Table 5.5), therefore the equation of the trend line shown in Figure 6.7 is used to calculate accurate solutions of the η parameter using

$$\eta^{LLD} = \eta^{CMOD} = 0.0037 + 5.1295(a/W) \quad (6.8)$$

for $0.35 \leq a/W \leq 0.45$ in the CCG data analyses on 316H. It can be seen in Figure 6.7 that the PE values obtained using the regression line are close to previously found range of 2.2 ± 0.1 at the crack lengths close to $a/W = 0.45$. Therefore, Eqn (6.8) has

been assumed to give accurate solutions for the normalised crack lengths in the range of $0.350 \leq a/W \leq 0.425$ and the general solution given in [67] can be used for larger crack lengths. The solutions of the η factor under plane stress and plane strain conditions for a wide range on a/W and N are summarised in Table 6.3.

6.3 Creep Crack Growth

Three CCG tests were performed on C(T) specimens with the initial crack length of $a_0/W = 0.35$ (8PC-A1, 8PC-A2 and 8PC-A3). The results from these tests were analysed and used to calibrate the CCG behaviour of the LCD specimens. Furthermore, an additional CCG test was conducted on a standard C(T) specimen with $a_0/W = 0.5$ (8PC-A4). The CCG test results on these specimens, which are all made of PC material extracted from Header A, are presented in this section and compared to the existing data on C(T) specimens with the initial normalised crack length of around $a_0/W \approx 0.5$, made of AR (AR-E1 and AR-C20) and PC (8PC-D1 and 8PC-D2) materials extracted from other cast and headers.

The specimen geometries and material condition for each sample are detailed in Table 6.4. As seen in this table, the long term tests on AR material (*i.e.* AR-C20) have been performed on C(T) specimens of width $W = 38$ mm, however the rest of the tests are conducted on standard specimens with $W = 50$ mm. Shown in [33] is that the CCG trends are similar in the tests performed on C(T) specimens made of AR material with the width of 50 mm and smaller. The loading condition, time to failure, t_f , amount of crack extension, Δa , transition time, t_T , and the initiation time corresponding to 0.2 mm and 0.5 mm crack extension are given in Table 6.5 for the tests considered here. Note that t_f in this table indicates the CCG test duration. All creep crack growth tests were performed at 550 °C.

6.3.1 Load up Curve

The load up behaviour of the CCG calibration tests on the C(T) specimens with $a_0/W = 0.35$, 8PC-A1–3, are presented in Figure 6.8. In this figure the load and displacement are presented in the normalised form. It can be observed in this figure that a linear load vs. displacement response is generally seen in the load up behaviour of the PC specimens. The comparison of the load up behaviour of the tests performed on PC material under different initial load levels show that some small deviation from linearity is observed in the loading behaviour of specimen 8PC-A1 at high loads, however 8PC-A2 and 8PC-A3 are showing a fully linear response which implies that performing local creep damage tests under this load level will ensure relatively small

non-linearity in the loading behaviour. It has been shown [89] that a linear elastic response is observed in the loading behaviour of long term (low load) tests on AR material, whereas some deviation from linearity and thus plasticity on loading is evident in short term (high load) tests on AR material.

A measure of the plastic zone size in each test is estimated for PS and PE conditions using Eqn (2.70) and given in Table 6.6. As seen in this table, the predicted plastic zone sizes in CCG tests on PC material are relatively small under both plane stress and plane strain conditions, apart from 8PC-A1 and 8PC-D1 which have higher $K(a_0)$ values (stress intensity factor at loading) and thus bigger plastic zone size compared to other PC specimens. Due to smaller yield strength of the AR material at 550°C, the estimated plastic zone size is noticeably larger in the CCG tests on AR material compared to PC. The short term (high load) test on AR material, which has the highest $K(a_0)$ among the tests compared in Table 6.6, is predicted to have the largest plastic zone size ahead of the crack tip at initial loading. The estimated plastic zone size in this test is around 5 times larger than those of predicted for 8PC-A2 and 8PC-A3 specimens.

The initial normalised reference stress has been calculated for each test under both plane stress and plane strain conditions using the load values given in Table 6.5 and the yield stress values in Table 5.1. As shown in Table 6.6, the normalised reference stress values calculated for the CCG tests on PC material are much smaller than unity under both plane stress and plane strain conditions, which implies that there is limited plasticity at initial loading in these specimens. It can be seen in Table 6.6 that the smallest values of the normalised reference stress are found in CCG tests performed on 8PC-A2 and 8PC-A3 specimen loaded under $K(a_0) = 25.5 \text{ MPa}\sqrt{\text{m}}$.

Table 6.6 shows that the normalised reference stress in the short term test on AR material, AR-E1, is greater than unity. However values close to and smaller than unity are calculated for the long term test on AR material, under plane stress and plane strain conditions, respectively. Further shown in Table 6.6 are the calculated normalised reference stress values for 8PC-A1–3 (which have $a_0/W = 0.35$) at $a/W = 0.5$. It can be seen in this table that under $K(a_0) = 25.5 \text{ MPa}\sqrt{\text{m}}$, the normalised reference stress values at both initial loading and $a/W = 0.5$ are smaller than unity. This means that relatively little plasticity at loading and unloading (test interruption) is expected to be observed in LCD specimens if those tests are loaded under the initial stress intensity factor of $K(a_0) = 25.5 \text{ MPa}\sqrt{\text{m}}$. Therefore, creep damage effects can be

investigated on the subsequent fatigue and fracture behaviour of the material in the absence of any significant plasticity.

It can be also seen in Table 6.5 that the CCG tests on PC material under $K(a_0) = 25.5 \text{ MPa}\sqrt{\text{m}}$ took around 1000 hours to complete. This test duration provides reasonably sufficient time to interrupt a CCG test at a PD rate corresponding to the approximate normalised crack length of $a/W \approx 0.5$. Therefore, $K(a_0) = 25.5 \text{ MPa}\sqrt{\text{m}}$ is selected as the initial stress intensity factor for LCD tests. 8PC-A2 and 8PC-A3 are used as the main CCG calibration test data to calculate the PD rate at which the normalised crack length reaches the approximate value of $a/W \approx 0.5$ in LCD specimens.

6.3.2 Load Line Displacement and Crack Growth Behaviour

The variation of the load line displacement, Δ , normalised by the specimen width, W , has been plotted against the time normalised by the test duration, t_f , in Figure 6.9 for all the CCG tests performed on PC and AR materials. The displacements after load up (during CCG tests) are presented in this figure and initial elastic-plastic load up displacements have been excluded from these measurements. It can be seen in Figure 6.9 that the load line displacement behaviour of 8PC-A2 and 8PC-A3 tests, which have the same dimensions and were performed under the same initial $K(a_0)$, are effectively coincident. Further seen in this figure is that the LLD of the AR and PC specimens with the normalised crack lengths of around $a_0/W \approx 0.5$ are generally smaller than the ones with $a_0/W = 0.35$, except towards the end of the test where some of the PC specimens with $a_0/W = 0.5$ show rapid acceleration in the LLD response.

In order to calibrate PD data accurately, the majority of the tests were stopped prior to final fracture of the specimen. However the large displacements at the end of the tests on 8PC-A1 and 8PC-D2 are attributed to the fact that relatively large crack extensions were obtained at the time these tests were stopped. Due to the rapid acceleration of the CCG and LLD rates exhibited in the tests performed on PC material, particularly towards the end of the tests, significant care was required to stop these tests before fast fracture occurs.

The variation of the crack length, a , normalised by the initial crack length, a_0 , is plotted against the time normalised by the test duration in Figure 6.10. As shown in this figure, similar crack growth behaviour is seen in 8PC-A2 and 8PC-A3 specimens which had the same dimensions and loading conditions. Further seen in Figure 6.10 is a rapid acceleration in the creep crack growth rates towards the end of the tests in the specimens made of PC material compared to those of measured on AR specimens.

6.3.3 Correlation with K

When the creep zone is smaller than the elastic field ahead of the crack tip, the crack growth behaviour can be characterised by the stress intensity factor, K . A material with the stress intensity factor as the controlling crack growth parameter is referred to as creep brittle. The creep crack growth rates have been correlated with the elastic stress intensity factor K for the specimens made of PC and AR materials in Figure 6.11 to investigate whether a unique relationship between the crack growth rate and K parameter can be observed. As seen in this figure, the test data from AR and PC materials show too much variability and no definite trend can be inferred. This means that the AR and PC 316H may not be considered as creep brittle materials and their CCG behaviour cannot be described by the K parameter. Furthermore, it can be seen in Figure 6.3 that the change in the shape function, Y , against the normalised crack length is relatively small for shallow cracks. Therefore small changes in the stress intensity factor are observed in the majority of the test data in Figure 6.11. This is more evident for the first few points in each of the data series, where the crack lengths are yet too small to make a noticeable change in the shape function and subsequently the stress intensity factor, K .

6.3.4 Correlation with C^*

At long times, where a steady state creep deformation and damage has developed at a crack tip, the CCG rate, da/dt , may be described by the C^* crack tip fracture mechanics parameter. The C^* validity criteria proposed by ASTM E1457, R5 and EDF Energy, which were explained in Chapter 4-Section 4.1.7, are applied to the experimental data and the CCG rate of the valid data points are correlated with the C^* parameter. A brief summary of the comments made on each criterion is described next.

6.3.4.1 ASTM E1457 Validity Criteria (see Chapter 4-Section 4.1.7.1)

- i. $\dot{\Delta}^c / \dot{\Delta}^T \geq 0.5$: Relatively small level of plasticity is expected to exist ahead of the crack tip and also in the LLD behaviour during CCG tests on PC material compared to AR. Therefore, the plastic displacement rate has been assumed negligible for the PC material and is not taken into account in the calculation of the creep load line displacement rate (see Eqns (2.103)–(2.105)). Also according to the ASTM E1457 standard [71], the analytical equation to calculate plastic load line displacement rate (see Eqn (2.105)) often leads to negative values of creep displacement rate. Knowing that the plastic term of the

displacement rate is generally small, this term can be neglected in creep displacement rate calculations.

- ii. $t > t_T$: In all CCG tests on PC material the transition time is calculated based on the average creep stress exponent, n_A , and the total load line displacement rates have been employed in C^* calculations. The transition times evaluated under PE conditions are given in Table 6.7.
- iii. $t > t_{0.2}$: It has been shown in [24], that 0.2 mm initiation time is usually longer than the transition time calculated using Eqn (4.2) in CCG tests on the AR 316H stainless steel material. This means that excluding the data points prior to the initiation time, may automatically exclude the data points corresponding to the times smaller than the transition time. It has also been mentioned in ASTM E1457 [71] and shown in [127] that the transition time is generally smaller than the initiation time ($t_T < t_{0.2}$) in creep-ductile materials.
- iv. $\Delta^{LLD} < 0.05W$: This criterion is often satisfied for the CCG test data.

6.3.4.2 R5 Additional Validity Criterion (see Chapter 4-Section 4.1.7.2)

$\lambda < 0.5$: The experimental C^* values employed in Eqn (4.3) are calculated based on the total load line displacement rates.

6.3.4.3 EDF Energy Proposed Criteria (see Chapter 4-Section 4.1.7.3)

- i. Digital imaging on the fracture surface: All the creep crack growth measurements presented in this work have been carried out using digital imaging technique.
- ii. Employment of the total displacement rate in C^* calculations: All of the experimental C^* values presented in this work are calculated using the total displacement rate.
- iii. Data smoothing: In order to overcome the noise problems in the raw displacement and PD data, a data reduction technique has been employed to smooth the LLD and crack extension curves.
- iv. t_{red}^{eng} criterion: It has been shown in [24] that t_{red} and t_{red}^{eng} are generally smaller than $t_{0.2}$ in the CCG tests performed on 316H stainless steel. This suggests that

$$\frac{\Delta_c(t_{red})}{\Delta_e(t_{red})} \text{ and } \frac{\Delta_c(t_{red}^{eng})}{\Delta_e(t_{red}^{eng})} \text{ ratio can be calculated assuming that no crack extension}$$

occurs prior to redistribution time ($\Delta a \approx 0$ for $t < t_{red}$ and $t < t_{red}^{eng}$).

Considering a constant load applied during CCG tests and no crack extension

before redistribution time, the elastic and creep load line displacements at the redistribution time can be defined as

$$\Delta_e(t_{red}) = \Delta_e(t_{red}^{eng}) = C_e(a_0)P \quad (6.9)$$

$$\Delta_c(t_{red}) = \Delta_c(t_{red}^{eng}) = \Delta_t \quad (6.10)$$

Therefore, the elastic displacement is assumed to be the same as that of initial loading and the creep displacement is taken as the total load line displacement Δ during creep, considering no additional elastic and plastic contribution in the LLD subsequent to the loading process and before the time that redistribution occurs.

In order to work out the elastic displacement at initial loading accurately, the numerical solutions of the elastic compliance (Eqn (6.3) for $a/W > 0.45$ and Eqn (6.4) for $0.35 \leq a/W \leq 0.45$) were employed in calculations. The numerical solutions of the elastic load line displacement for different tests are summarised in Table 6.8.

6.3.4.4 Summary of C^* Validity Criteria

The C^* validity criteria according to the ASTM E1457 and R5 are summarised in Table 6.7 and the experimental redistribution times proposed by EDF Energy are given in Table 6.8. As seen in Table 6.7, the λ criterion is satisfied for all the test data examined. It can be observed in Figure 6.9 and Table 6.7 that Δ^{LLD} / W is less than 0.05 in all CCG tests examined, therefore the ASTM E1457 Δ^{LLD} criterion is satisfied for all tests. As shown in Table 6.8 the experimental redistribution time calculated using Eqn (4.6) generally gives larger values compared to those of obtained from Eqn (4.8). This indicates that the redistribution times calculated using Eqn (4.6) may be assumed over predicted as noted in [24]. Therefore the engineering redistribution time, calculated by Eqn (4.8), may be considered a more appropriate criterion to ensure that the data points within the transient creep region are excluded in da/dt vs. C^* graphs. The comparison of the engineering redistribution time to the initiation and transition time is plotted in Figure 6.12 for both PS and PE conditions. Figure 6.12 shows that in the CCG tests on PC material, t_{red}^{eng} is generally equal to or smaller than t_T , however in the tests on AR material the transition time is found smaller.

It has been explained in [24] that Eqn (4.2) can significantly over estimate the transition time in materials with large primary creep region *i.e.* AR 316H steel. Therefore the engineering redistribution time criterion has been proposed to be used instead. It was shown in the Chapter 5 that the primary creep behaviour of the PC

material is small for high stresses and relatively large under low stresses. Since the CCG tests on PC material have been performed within a range of initial applied loads, which may correspond to small or large primary creep, the transition time can be considered overestimated or accurate. In order to satisfy the validity criteria conservatively, t_{red}^{eng} has been selected as the more appropriate parameter to specify the transient creep region and thus the data points corresponding to the times shorter than t_{red}^{eng} are considered invalid.

Also shown in Figure 6.12 is the comparison of the t_T and t_{red}^{eng} to the initiation time, $t_{0.2}$. This figure shows that the initiation time is equal to or larger than t_T and t_{red}^{eng} in most of the tests examined, therefore exclusion of the data points prior to $t_{0.2}$ will automatically satisfy t_T and t_{red}^{eng} criteria. Note that in the long term test on AR material t_T and t_{red}^{eng} are both larger than the initiation time. Also in 8PC-A2 and 8PC-A3, the transition time is found much larger than the engineering redistribution and initiation times. This can be due to relatively low level of stress in these tests which may correspond to a large primary creep and thus overestimated t_T values calculated using Eqn (4.2). So the over predicted values of t_T in these tests have not been accounted for in the C^* validity analyses and the engineering redistribution times have been considered more realistic to determine the transient creep behaviour.

The ratio of creep to total load line displacement rate is illustrated in Figure 6.13. As seen in this figure $\dot{\Delta}^c / \dot{\Delta}^T$ is larger than 0.5 for the majority of the test duration in most of the tests examined. Seen in Figure 6.13 is that the CCG calibration test data on 8PC-A2 and 8PC-A3 are found to satisfy this criterion for almost the first 50% and 75% of their test durations, respectively. However significant elasticity effects (and thus drop in $\dot{\Delta}^c / \dot{\Delta}^T$ ratio) is evident towards the end of these two tests. As mentioned earlier, the plastic load line displacement rate has been considered negligible and not been taken into account for the calculation of creep load line displacement rates.

In addition to $\dot{\Delta}^c / \dot{\Delta}^T$, all other validity criteria from ASTM E1457, R5 and EDF are applied in Figure 6.13. In this figure, the valid data points are shown in black and the invalid points are shaded in grey. Also observed in Figure 6.13 are small or negative values of the $\dot{\Delta}^c / \dot{\Delta}^T$ ratio during a large proportion of the test duration in the long term test on AR material. It has been explained in [24] that the elastic displacement rate formula in Eqn (2.104) gives overestimated values for long term tests on AR 316H steel and thus leads to small or negative values of $\dot{\Delta}^c / \dot{\Delta}^T$. This can be related to possible discontinuous growth in the cracking behaviour of the long term tests which makes the experimental elastic displacement rate being significantly smaller than those of

calculated using Eqn (2.104). Because of the overestimating nature of the analytical elastic displacement rate, the $\dot{\Delta}^c / \dot{\Delta}^T$ criterion is not applied for the long term test on AR material (AR-C20).

Figure 6.13, Table 6.7 and Table 6.8 show that $t_{0.2}$ and $\dot{\Delta}^c / \dot{\Delta}^T$ are the main validity criteria for the CCG tests performed on specimens made of PC material. Therefore, all other criteria proposed by ASTM, R5 and EDF are automatically satisfied by applying the $t_{0.2}$ and $\dot{\Delta}^c / \dot{\Delta}^T$ criteria on PC material. For the CCG tests on AR material, $\dot{\Delta}^c / \dot{\Delta}^T$, $t_{0.2}$ and t_{red}^{eng} are found to be the key criteria to specify valid data points. Though for long term tests on AR material the $\dot{\Delta}^c / \dot{\Delta}^T$ criterion has been found unsuitable due to the large fluctuations observed in Figure 6.13.

6.3.5 NSW CCG Predictions

The validity criteria have been applied to the CCG calibration test data on samples 8PC-A1–3 and the valid CCG rate data points are correlated with the C^* parameter in Figure 6.14. Note that appropriate values of the η parameter (see Section 6.2.3) have been employed in C^* calculations for these tests. It can be seen in this figure that some “tail” regions in the early stage CCG trends are apparent even though the validity criteria are attained. The CCG data from the tests performed on PC material are presented and compared to the CCG trends predicted by NSW models in Figure 6.15 and Figure 6.16. Included in these figures are the data obtained from the CCG tests on 8PC-A1–4 specimens, extracted from Header A, and also those of available in [2, 89] from other cast and headers. Note that the data previously available in the literature have been re-analysed by applying the additional C^* validity criteria.

The average value of the uniaxial creep ductilities presented in Chapter 5 for the PC material extracted from Header A, measured based on both axial deformation and ROA, are given in Table 6.9. As seen in this table, the strain values calculated using engineering and true strain definitions are fairly close, therefore only the mean values of the creep ductilities calculated based on the true strain definition have been employed in NSW predictions. It can be seen in Table 6.9 that ε_f^{true} - ROA is approximately 5 times larger than ε_f^{true} - Axial. This means that the NSW CCG rate predictions using ε_f^{true} - ROA are around 5 times lower, for a given value of C^* .

The available experimental CCG data on PC material are compared to the plane stress and plane strain NSW lines predicted by employing the ε_f^{true} - ROA in Figure 6.15 and Figure 6.16. Similar predictions using ε_f^{true} - Axial, which results in CCG rates of around 5 times higher than the ones shown in Figure 6.15 and Figure 6.16, are

presented in Appendix C. In these figures, the valid and invalid data points are shown in black and grey, respectively. In order to examine the n dependency of the CCG predictions using NSW models, the average and minimum power law creep stress exponents are employed and the results are shown in Figure 6.15 and Figure 6.16, respectively. The comparison of Figure 6.15 and Figure 6.16 shows that the CCG rates predicted using the minimum n value of 10.9 are slightly smaller than the ones estimated using the average creep stress exponent $n_A = 14.0$ in NSW and NSW-MOD models, however as seen in Eqn (2.118) NSWA is insensitive to the change in n value.

Apart from the early regions of the tests where some tails can be observed, the majority of the valid data points fall between plane stress and plane strain prediction lines using NSW and NSWA models in Figure 6.15(a) and Figure 6.15(c), respectively. Similar observations are made in Figure 6.16(a) and Figure 6.16(c) where the minimum creep stress exponent is employed in NSW models. Figure 6.15(b) shows that NSW-MOD plane stress line provides an excellent prediction of the experimental trend for the valid data points (excluding tails) when $n_A = 14.0$ is employed in calculations. However when the minimum power law exponent $n = 10.9$ is used, the valid data points fall in between the plane stress and plane strain NSW-MOD predictions lines as seen in Figure 6.16(b).

Comparison of the predictions made using different NSW models in Figure 6.15 and Figure 6.16 shows that generally good predictions are made when the ROA creep ductility is employed in calculations. It is also seen that NSW-MOD model provides less conservative estimations of the CCG rate, compared to NSW and NSWA, especially under plane strain conditions.

6.3.6 Comparison with the Available Short/Long Term CCG Data on AR Material

The valid data points from the CCG tests on PC material are compared to the short term (large C^*) and long term (small C^*) data on AR material in Figure 6.17. This figure shows that the PC data from different tests fall close to each other and the CCG behaviour of the samples made of PC material with different initial crack lengths of $a_0/W = 0.35$ and $a_0/W \approx 0.5$ is found similar. Figure 6.17 shows that a power law relationship between da/dt and C^* can be inferred for the valid data points (excluding tails) obtained from the tests performed on PC material. Also included in this figure are the mean, upper bound (UB) and lower bound (LB) fits to the PC, short term AR and long term AR data which are shown in black, blue and pink lines, respectively. The mean fits to the data are shown in solid lines whilst the UB/LB trends have been

illustrated by dashed lines. The power law exponent, ϕ , and the coefficient, D , values determined (see Eqn (2.93)) from these fits as given in Table 6.10. Also included in Table 6.10 is the ratio of the upper bound to the mean fit regression lines, for a given value of C^* . These ratios indicate the level of scatter in the data. It can be seen in Table 6.10 that the slope of the short term AR and PC creep crack growth data may be considered equal ($\phi = 0.9$ which is a fraction close to unity as expected [74]). Table 6.10 and Figure 6.17 show that the CCG rate in PC material is on average around an order of magnitude higher than the mean line fitted to the short term AR data, for a given value of C^* . Also seen in Figure 6.17 is that the steady state CCG test data (excluding tail regions) on PC material follow the trend of the UB line fitted to the long term AR data, though the C^* values in PC data are larger than the ones in long term AR. This suggests that, for this material and conditions, the behaviour of long term CCG tests on AR material may be simulated in much shorter time scales through material pre-compression. This is thought to be due to:

- i. High specimen constraint effects: Widespread plasticity is known to reduce the specimen constraint resulting in a behaviour more closely to plane stress rather than plane strain conditions and thus leading to lower CCG rates [50, 128]. It has been observed that the pre-compression process hardens the material and thus the levels of crack tip plasticity and the extent of non-linearity exhibited during loading are reduced. The level of crack tip plasticity is also limited in the long term tests on AR material due to the low load level applied. Table 6.6 shows that the normalised reference stress values during load up, for the tests on PC material and also the long term test on AR material (AR-C20) are close to or smaller than unity, which indicates limited plasticity at the crack tip in these samples.
- ii. Similar creep ductility: As discussed in Chapter 5, the creep ductility is reduced by the pre-compression process, which leads to higher CCG rates. It was also shown in Figure 5.13 that the creep ductility of the AR material reduces in long term (low stress) tests due to *e.g.* thermal aging effects. Therefore, similar creep ductilities can be attained in PC material and long term tests on AR material which may provide similar CCG trends when da/dt is plotted against C^* .

6.3.7 Comparison of the Cracking Modes in PC and AR Materials

Due to the similar trends observed in the CCG tests conducted on PC material and long term tests on AR material, the cracking modes for these tests are investigated in this section. The cracking mode of the AR material in short term and long term CCG

tests are taken from [24] and shown in Figure 6.18. The optical microscopy pictures have been captured at the mid thickness of C(T) specimens, subsequent to test completion. Figure 6.18 shows that although the CCG is generally an intergranular process, but the cracking mode can change with the test duration. The cracking behaviour in a short term (high load) test on AR material, shown in Figure 6.18(a), is ductile intergranular where a bifurcation along the planes of the maximum shear stress (which are at angles of around $\pm 45^\circ$ with respect to the pre-cracking direction) happens at the crack tip in the early stage of creep crack initiation and grain deformation is observed duration the creep crack growth process. The cracking mode of a long term (low load) test on AR material shown in Figure 6.18(b) can be described as brittle intergranular, where the creep crack initiation and growth happen with small deviation from the symmetry plane (the pre-cracking axis) and little or no evidence of grain deformation is observed [24].

The cracking mode of a C(T) specimen made of PC material (8PC-A2) is shown in Figure 6.19 and found similar to that of observed in the long term test on AR sample. The microscopic pictures show that the creep crack initiation and growth take place along the symmetry plane of the sample and small deviations are seen along the crack path. Therefore similar to the long term AR specimen, the cracking mode of the PC material can be described as brittle intergranular. The comparison of the cracking modes in PC material and long term test on AR material show that more significant intergranular creep damage exists ahead of the final crack length in PC material compared to AR. In the upper image of Figure 6.19, taken from the mid thickness of the sample, the creep crack growth seems discontinuous at the early stages of crack growth. However the lower pictures in Figure 6.19, taken from a side groove plane, exhibit continuous crack extension. This is due to the 3-dimensional nature of the CCG behaviour in polycrystalline materials. Further seen in Figure 6.19 is that the significant creep damage in PC material doesn't only exist ahead of the final crack tip but micro cracks along the grain boundaries are also observed in the vicinity of the main crack path, in a direction normal to the maximum principal stress.

The final crack length measured by breaking open the other half of the 8PC-A2 specimen is also shown in Figure 6.19. It can be seen in this figure that breaking open a C(T) specimen made of PC material may link up the micro cracks at the end of the main crack path and thus the actual crack length existing in the specimen may be slightly smaller than the one measured from the fracture surface. This small difference may imply some crack discontinuities in the creep damage zone around the final crack tip.

6.4 Creep Crack Initiation

The $t_{0.2}$ and $t_{0.5}$ initiation time normalised by the test duration for the CCG tests performed on AR and PC materials are given in Table 6.5. Comparing the ratio of initiation time (corresponding to 0.2 and 0.5 mm crack extension) to the test duration for the CCG calibration tests with $a_0/W = 0.35$ on 8PC-A1–3 specimens in Table 6.5, it can be seen that the 8PC-A1 test with a higher applied load (and thus larger normalised reference stress) had a larger percentage of $t_{0.2}/t_f$ compared to the lower load tests on 8PC-A2 and 8PC-A3. This may be due to local plasticity and blunting effects at the crack tip in the high load test on 8PC-A1 which means that it takes a longer time for the crack to initiate from a pre-existing blunt notch. The comparison of the ratio of CCI to the test duration for the CCG tests with $a_0/W \approx 0.5$ on AR (with pre-fatigue starter crack) and PC (with EDM starter crack) materials in Table 6.5 confirms that applying a higher load on a C(T) sample increases the ratio of initiation time to the test duration, regardless of the pre-cracking type and material's pre-straining condition.

6.4.1 Correlation with C^* and Comparison to AR Data

It was shown in Section 6.3 that the CCG behaviour of the C(T) specimens made of PC material may be described as creep ductile. Therefore, the initiation time measured in the tests performed on PC material are correlated with the C^* parameter and the results are compared to the existing short term and long term initiation data on AR material in Figure 6.20. It can be seen in this figure that the initiation time in AR material is inversely related to C^* , when the graph is plotted in log scale (*i.e.* longer test duration means smaller C^* and leads to a larger t_i).

Figure 6.20(a) shows that for a given value of C^* , the initiation time corresponding to 0.2 mm crack extension in relatively low load CCG tests on PC material (8PC-A2, 8PC-A3, 8PC-A4 and 8PC-D2) is on average around an order of magnitude shorter than those of obtained from the AR material. Though $t_{0.2}$ in higher load CCG tests on PC material (8PC-A1 and 8PC-D1) is found very similar to the initiation time in AR material, for a given value of C^* .

Similar trends are observed for $t_{0.5}$ in Figure 6.20(b), however as seen in this figure the difference between the initiation times in the low load CCG tests on C(T) specimens with $a_0/W \approx 0.5$ extracted from PC material (8PC-A4 and 8PC-D2) and the high load tests on AR material decreases by considering the CCI to occur at a larger crack extension. Also seen in this graph is that the $t_{0.5}$ in high load tests on PC material

(8PC-A1 and 8PC-D1) is again similar to the AR data, but other tests on PC specimens with $a_0/W = 0.35$ initiate faster than AR material by around an order of magnitude.

Also included in Figure 6.20(a) and Figure 6.20(b) are the average CCI trends calculated by Eqn (2.119) for 0.2 mm and 0.5 mm crack extensions, respectively, using the steady state CCG constants found for the PC material (see Table 6.10). It can be observed in Figure 6.20 that the experimental CCI line provides an overall fit to the initiation data for the PC material. Also seen in this figure is that should the mean CCI fit to the PC data be extrapolated to lower C^* values, then it would provide a good fit to the existing long term CCI data on AR material.

6.4.2 NSW CCI Predictions

The available initiation data from the tests performed on PC material are compared to the NSW prediction lines in Figure 6.21 and Figure 6.22 for 0.2 and 0.5 mm crack extension, respectively. In these figures, the ROA creep ductility is employed in NSW calculations. Other predictions using the axial creep ductility are shown in Appendix C. As shown previously in Section 6.3.5, the change in n has small effects on NSW and NSW-MOD creep crack growth predictions and doesn't have any influence on NSW model. Therefore, the CCI predictions made using the average power law creep stress exponent ($n_A = 14.0$) are only presented here and other calculations using minimum power law stress exponent ($n = 10.9$) are given in Appendix C.

As seen in Figure 6.21, the initiation time $t_{0.2}$ in higher loads CCG tests on PC material (8PC-A1 and 8PC-D1) generally fall in between the upper bound and lower bound PS predictions (closer to the upper bound PS) made using NSW models. However the lower load tests on PC material fall in between upper bound and lower bound PE predictions lines. These predictions for the CCI are consistent with the CCG behaviour. Applying a higher load on the specimen increases the level of plasticity at the crack tip which may result in the loss of specimen constraints if widespread plasticity occurs. Therefore the material behaviour can be described as PS. However under lower loads, there is a higher level of constraint in the specimen and PE condition is expected to be dominant. Figure 6.21 shows that acceptable predictions for $t_{0.2}$ can be made by all NSW models, but the lines based on the steady state CCG rates under plane strain conditions are over conservative in NSW and NSW models.

Similar trends are observed in the CCI initiation prediction for $t_{0.5}$ in Figure 6.22. As seen in this figure, all NSW models provide relatively good prediction of the CCI time for the tests performed on PC material. The comparison of the experimental data to the

NSW lines shows that for $\Delta a = 0.5$ mm, the creep crack initiation data points for the low load tests on specimens with $a_0/W = 0.35$ (8PC-A2 and 8PC-A3) stay close to the PE prediction lines, but the higher load tests on samples with $a_0/W \approx 0.5$ (8PC-A1 and 8PC-D1) fall closer to the PS prediction lines. In other words, the CCI time corresponding to larger crack extensions may be described by PS prediction lines even in some of the low load tests examined.

6.5 Discussion

In this work, due to the limited number of uniaxial creep rupture tests on PC material, the average of the available creep strains at failure presented in Chapter 5 was applied in NSW models for CCI and CCG predictions. However in order to provide accurate predictions of the CCI and CCG behaviour using NSW models for the tests performed on PC material under different stress levels, an appropriate creep ductility value from the upper shelf/transition/lower shelf creep ductility trend shown in Chapter 5-Section 5.4.3 must be employed in NSW model calculations. Knowing that the NSW predictions are based on a constant value of creep ductility, FE models are needed to define a variable creep ductility failure criterion to simulate creep crack growth based on the local stress levels existing ahead of the crack tip. These simulations have been performed and the results are presented in Chapter 7.

All CCG test results shown in this chapter were from C(T) specimens made of PC material with the loading axis parallel to the pre-compression direction. However, the specimen orientation, with respect to the pre-compression axis, can significantly influence the tensile, creep deformation, crack initiation and growth behaviour of 316H stainless steel at 550 °C. These effects have been examined and discussed in [129, 130].

6.6 Summary

- The CCG rate in PC material is around an order of magnitude higher than short term tests on AR material, for a given value of C^* , and the data follow the trend of upper bound long term tests on AR material.
- NSW-MOD model provides good predictions of the CCG behaviour of the PC material when ROA creep ductility is employed in calculations.
- Acceptable but more conservative predictions are also made by NSW and NSWA models.
- The ratio of CCI time to the test duration time increases by increasing the applied load in the tests performed on AR and PC materials.

- For a given value of C^* , the initiation time in low load tests on PC material is around an order of magnitude shorter than the AR data. However, high load tests have shown initiation times similar to AR material.
- The initiation time data for the long term tests on the AR material fall close to the experimental CCI trend obtained from the tests performed on PC material.
- The CCI time in the high load tests conducted on PC material can be predicted using plane stress NSW lines, however lower load tests are well described by PE initiation time predictions made by NSW models.

6.7 Tables

Table 6.1: % Error between numerical values of the shape function and Tada solutions

a/W	$Y(a/W)$ FE	$Y(a/W)$ Eqn (2.66)	% Error
0.350	11.03	10.80	2.0
0.375	11.30	11.13	1.4
0.400	11.58	11.51	0.6
0.425	11.99	11.94	0.5
0.450	12.45	12.43	0.2

Table 6.2: % Error between numerical and analytical solutions of the elastic compliance

a/W	C_eBE' FE	C_eBE' Eqn (6.3)	% Error
0.350	17.06	18.10	6.1
0.375	19.32	20.35	5.3
0.400	21.86	22.88	4.7
0.425	24.70	25.74	4.2
0.450	27.93	28.98	3.8

Table 6.3: Numerical solution of the η factor for a wide range of a/W and N under plane stress and plane strain conditions

Range of a/W	$\eta^{LLD} = \eta^{CMOD}$	PS/PE conditions	Range of N
$0.425 < a/W \leq 0.7$	2.2 ± 0.1	PS and PE	$3 \leq N$
$0.35 \leq a/W \leq 0.425$	2.2 ± 0.1	PS	$3 \leq N$
	$0.6082 + 3.6279(a/W)$	PE	$3 \leq N \leq 20$
	$0.0037 + 5.1295(a/W)$		$N = 10$

Table 6.4: Specimen geometries

Test ID	Material Condition	W (mm)	B (mm)	B_n (mm)	a_0/W
8PC-A1	8% PC	50	25	17.5	0.35
8PC-A2	8% PC	50	25	17.5	0.35
8PC-A3	8% PC	50	25	17.5	0.35
8PC-A4	8% PC	50	25	17.5	0.50
8PC-D1	8% PC	50	25	19.8	0.54
8PC-D2	8% PC	50	25	20.0	0.50
AR-E1	AR	50	25	20.2	0.53
AR-C20	AR	38	19	15.1	0.54

Table 6.5: Creep crack growth loading details and test results

Test ID	P (kN)	$K(a_0)$ (MPa $\sqrt{m}$)	t_f (h)	Δa (mm)	t_T / t_f (%)	$t_{0.2} / t_f$ (%)	$t_{0.5} / t_f$ (%)
8PC-A1	21.9	30.6	606	15.6	4.2	33.9	57.9
8PC-A2	18.3	25.5	1303	8.5	5.0	0.7	2.9
8PC-A3	18.3	25.5	966	9.8	6.5	1.0	2.7
8PC-A4	12.1	25.0	194	9.4	3.9	4.8	54.5
8PC-D1	15.4	34.7	1205	5.0	1.0	37.3	45.9
8PC-D2	12.9	25.0	357	11.9	5.5	5.0	33.6
AR-E1	20.0	42.3	1221	1.4	15.1	27.9	61.4
AR-C20	7.8	26.0	17850	4.5	24.7	6.6	23.8

Table 6.6: The plastic zone size estimations and normalised reference stress calculations for the CCG tests

Test ID	a_0/W	$\sigma_{ref} / \sigma_{0.2}$ Load up PS	$\sigma_{ref} / \sigma_{0.2}$ Load up PE	$\sigma_{ref} / \sigma_{0.2}$ $a/W=0.5$ PS	$\sigma_{ref} / \sigma_{0.2}$ $a/W=0.5$ PE	r_p –PS (mm)	r_p –PE (mm)
8PC-A1	0.35	0.59	0.40	1.09	0.75	2.97	0.99
8PC-A2	0.35	0.49	0.33	0.91	0.63	2.06	0.69
8PC-A3	0.35	0.49	0.33	0.91	0.63	2.06	0.69
8PC-A4	0.50	0.60	0.41	-	-	1.98	0.66
8PC-D1	0.54	0.84	0.58	-	-	3.80	1.27
8PC-D2	0.50	0.56	0.39	-	-	1.98	0.66
AR-E1	0.53	1.51	1.05	-	-	9.85	3.28
AR-C20	0.54	1.09	0.76	-	-	3.72	1.24

Table 6.7: ASTM E1457 and R5 C^* validity criteria

Test ID	t_f (h)	$t_{0.2}$ (h)	Time for $\lambda < 0.5$ PE (h)	Time for $\Delta^{LLD} < 0.05W$ (h)	Time for $\dot{\Delta}^c / \dot{\Delta}^T \geq 0.5$ PE (h)	t_T PE (h)
8PC-A1	606	205	606	606	603	23
8PC-A2	1303	9	1303	1303	662	61
8PC-A3	966	10	966	966	746	61
8PC-A4	194	9	194	194	167	7
8PC-D1	1205	449	1205	1205	1205	11
8PC-D2	357	18	357	357	342	18
AR-E1	1221	340	1221	1221	1221	168
AR-C20	17850	1182	17850	17850	16507 < t < 17541	4017

Table 6.8: EDF Energy C^* validity criteria

Test ID	Numerical Δ_e^{LLD} (mm)		Experimental redistribution times (h)			
			t_{red}		t_{red}^{eng}	
	PS	PE	PS	PE	PS	PE
8PC-A1	0.107	0.097	25	21	4	4
8PC-A2	0.089	0.081	28	21	6	6
8PC-A3	0.089	0.081	19	15	7	6
8PC-A4	0.128	0.116	10	9	5	5
8PC-D1	0.206	0.187	25	17	2	2
8PC-D2	0.137	0.124	86	73	27	24
AR-E1	0.246	0.224	793	748	381	317
AR-C20	0.134	0.122	15040	14288	9580	8610

Table 6.9: The uniaxial creep properties employed in NSW predictions

Material	ε_f^{eng} Axial (%)	ε_f^{true} Axial (%)	ε_f^{eng} ROA (%)	ε_f^{true} ROA (%)	$\dot{\varepsilon}_0 = \dot{\varepsilon}_{0_A}$ (h ⁻¹)	σ_0 (MPa)	σ_{0_A} (MPa)	n	n_A
8PC-Header A	2.12	2.09	9.45	10.32	1	802	607	10.9	14.0

Table 6.10: The CCG power law constants obtained from the mean fits to the data

Material	ϕ	D	UB/LB factor
PC	0.91	94	3.01
AR (Short term)	0.89	11	2.17
AR (Long term)	0.73	9	1.83

6.8 Figures

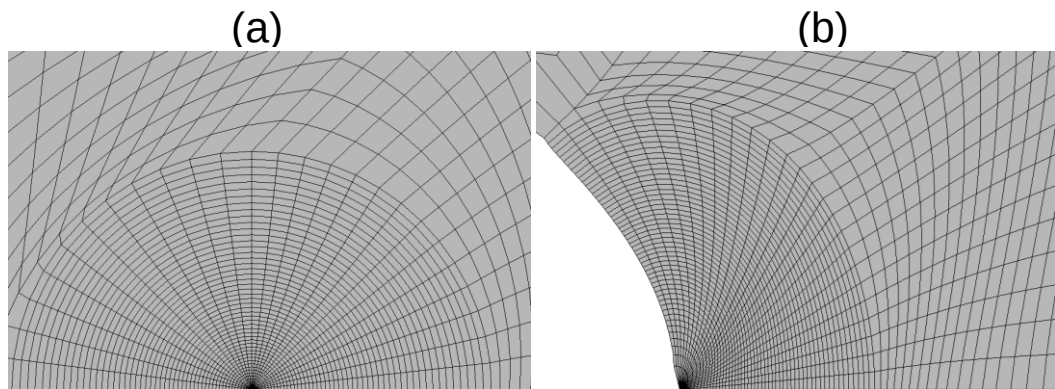


Figure 6.1: An example of a single node crack tip focused mesh (a) before and (b) after, deformation

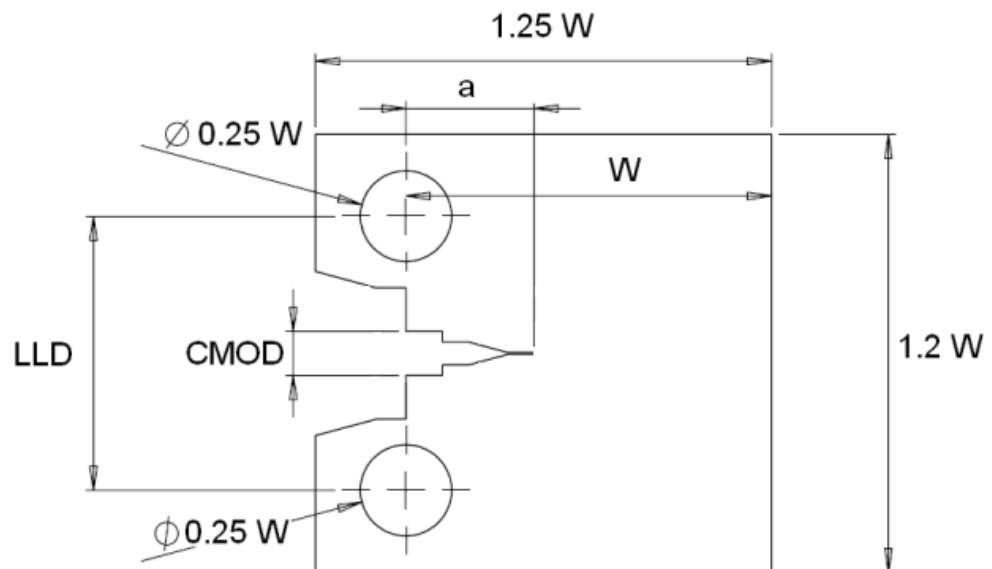


Figure 6.2: Illustration of the standard compact tension specimen dimensions for CCG tests [71], where the width is usually taken as $W = 50$ mm

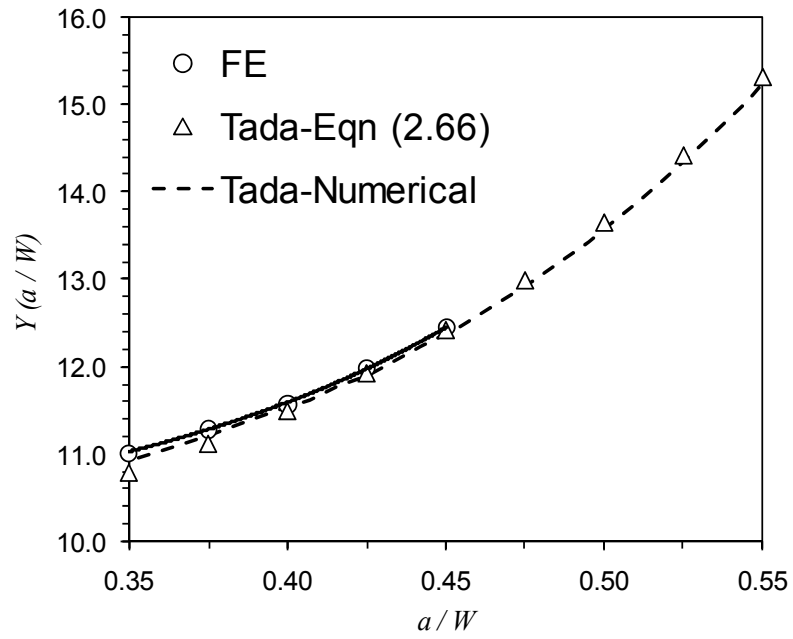


Figure 6.3: Comparison of FE solutions of the shape function to the empirical Tada solutions (given by Eqn (2.66)) and Tada-Numerical solutions given in [60] for $0.35 \leq a/W \leq 0.45$

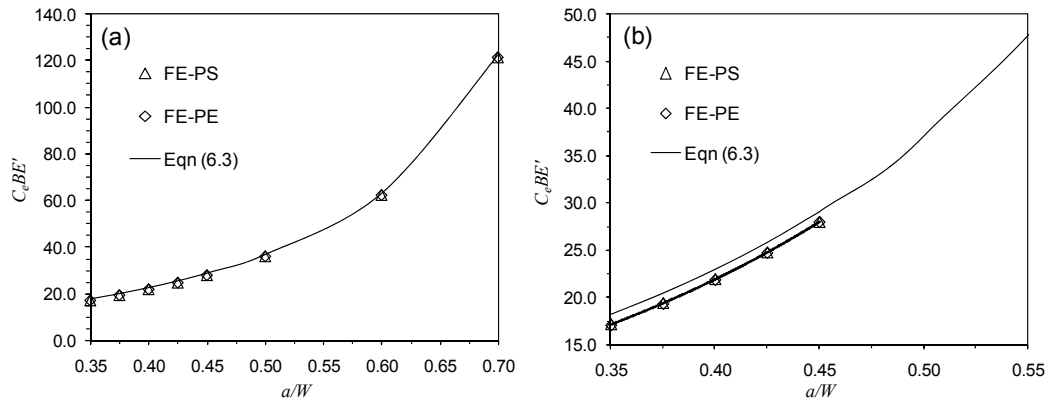


Figure 6.4: Comparison of the numerical and analytical solutions of the elastic compliance in the non-dimensional form for (a) $0.35 \leq a/W \leq 0.70$ (b) $0.35 \leq a/W \leq 0.45$, using CMOD under PS and PE conditions

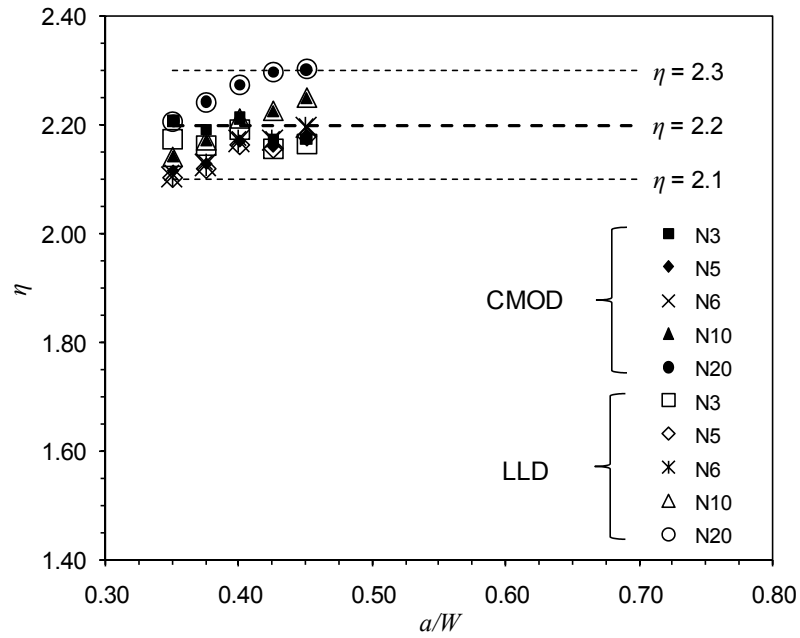


Figure 6.5: Numerical solutions of the η factor at crack lengths in the range of $0.35 \leq a/W < 0.45$ for different values of N under plane stress conditions

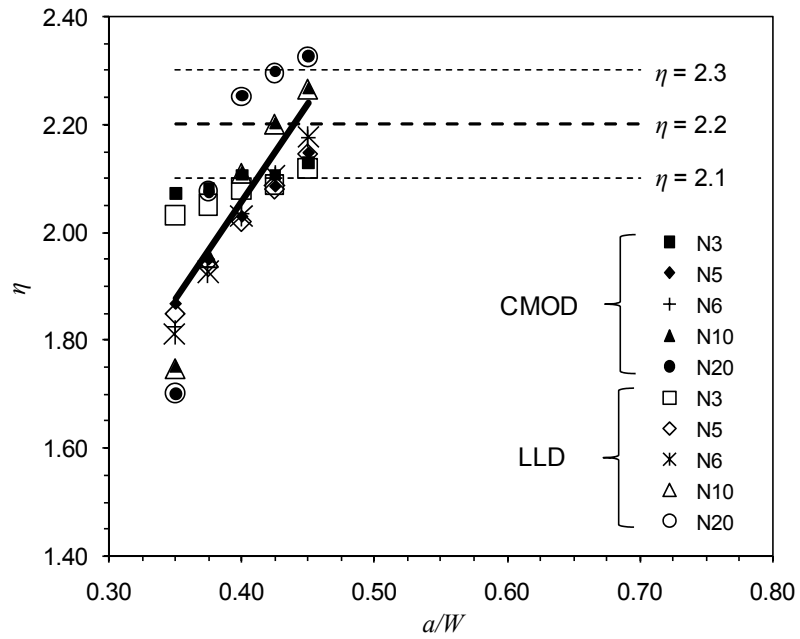


Figure 6.6: Numerical solutions of the η factor at a crack length in the range of $0.35 \leq a/W < 0.45$ for different values of N under plane strain conditions

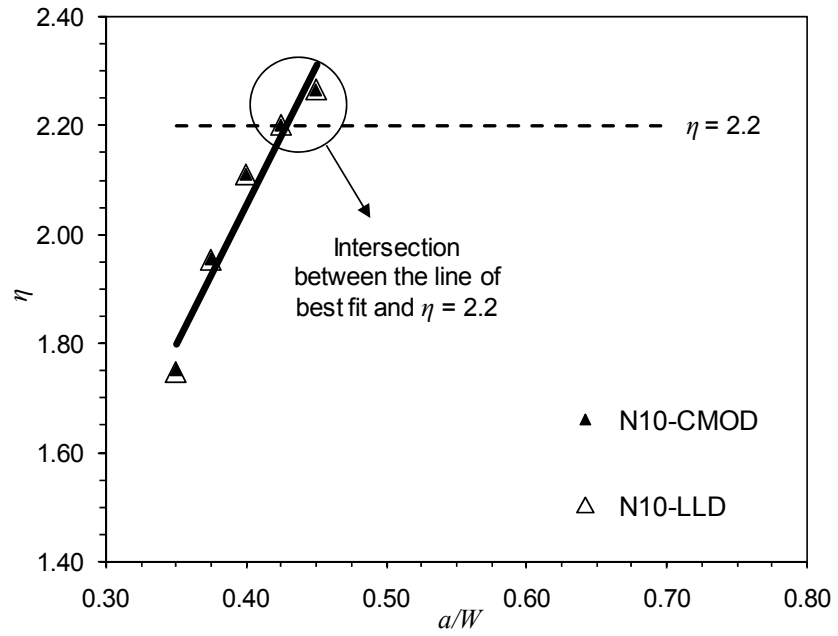


Figure 6.7: The line of best fit to the numerical solutions of the η factor at the crack lengths in the range of $0.35 \leq a/W \leq 0.45$ for $N = 10$ under plane strain conditions

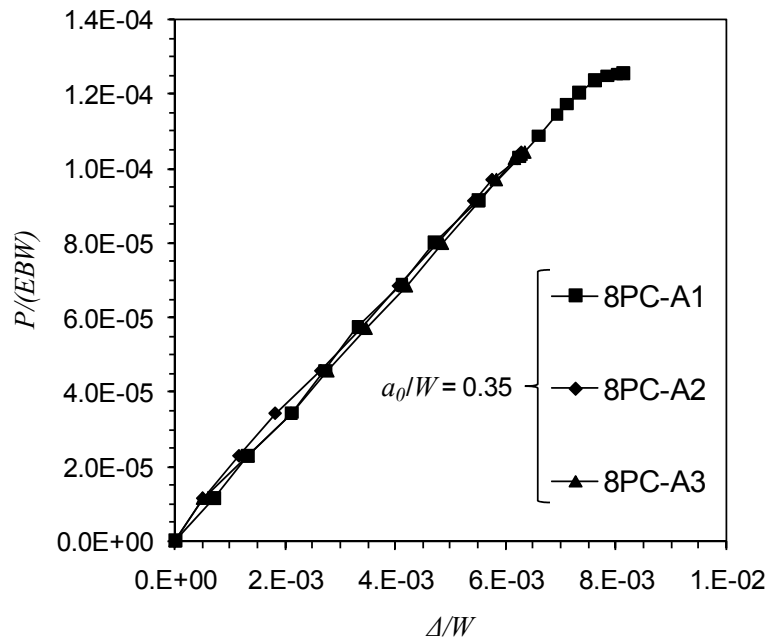


Figure 6.8: Comparison of the load up curves in CCG calibration tests

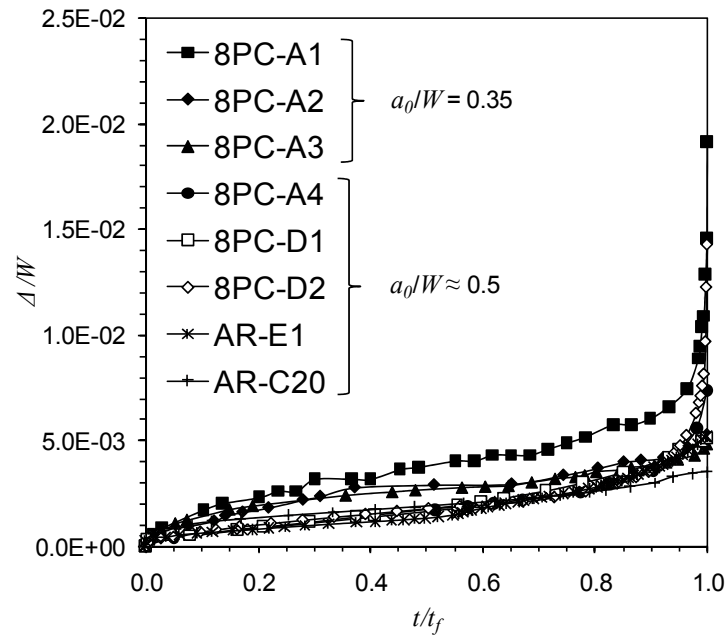


Figure 6.9: Variation of normalised load line displacement against the time normalised by the test duration

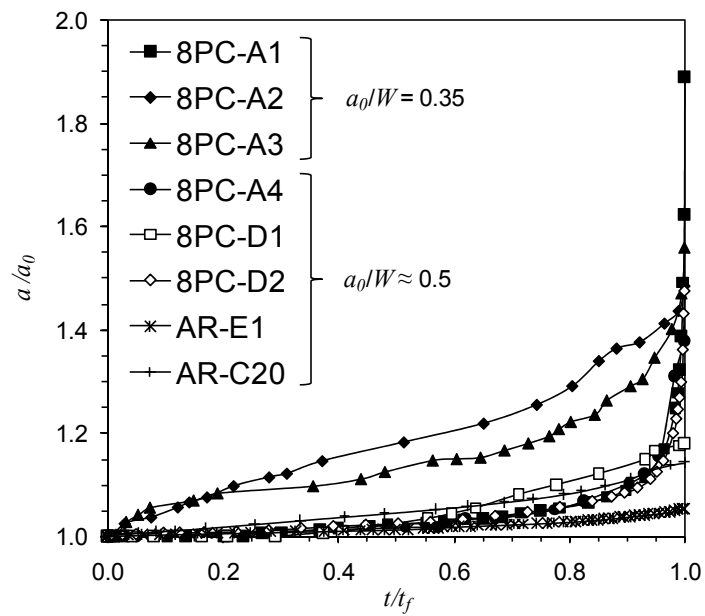


Figure 6.10: Variation of the normalised crack length against the time normalised by the test duration

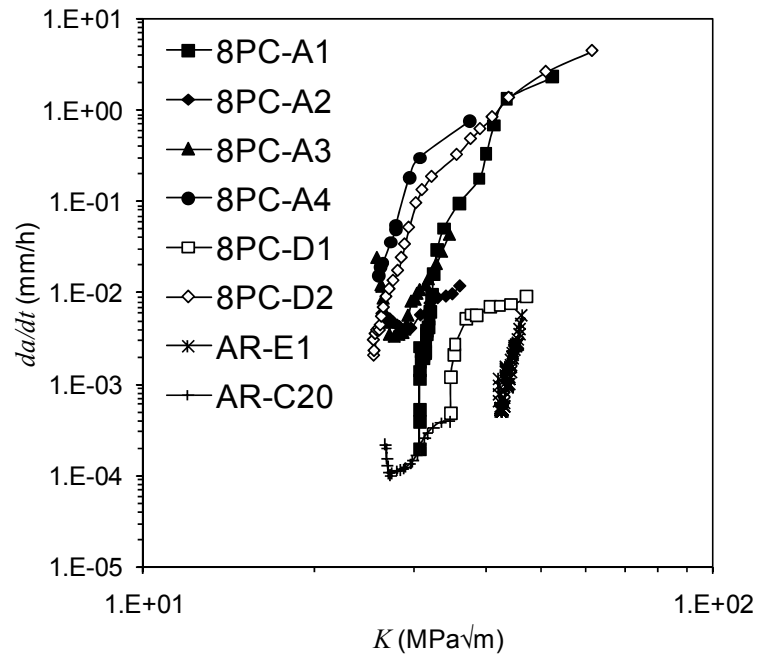


Figure 6.11: Creep crack growth rate plotted against stress intensity factor for pre-compressed and as-received materials

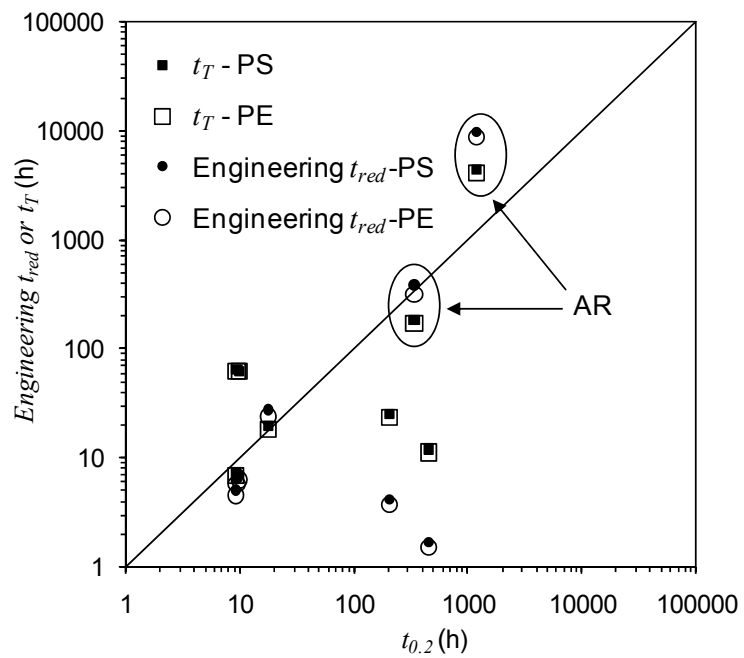


Figure 6.12: Comparison of the transition and engineering redistribution time to 0.2mm initiation time

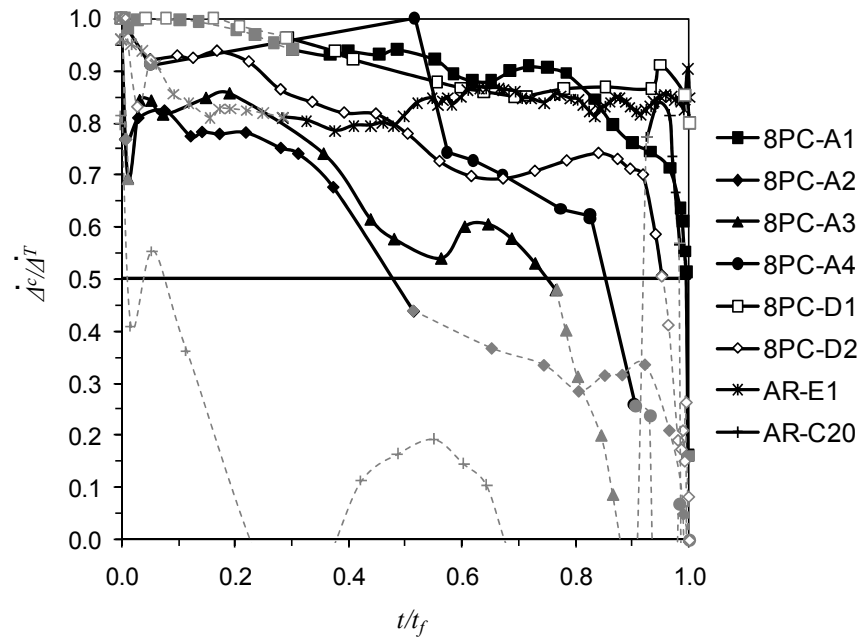


Figure 6.13: Illustration of the C^* validity criteria.

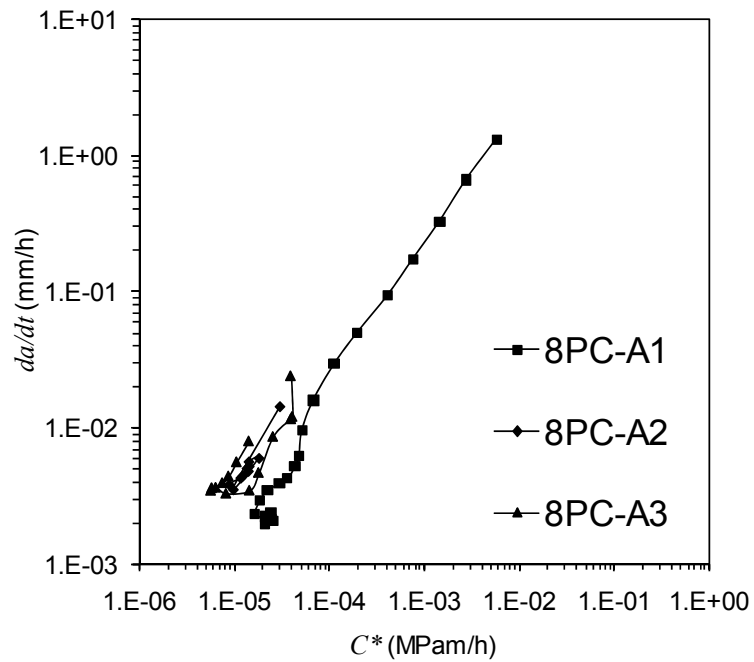


Figure 6.14: Comparison of the CCG test data for the PC specimens with $a_0/W = 0.35$

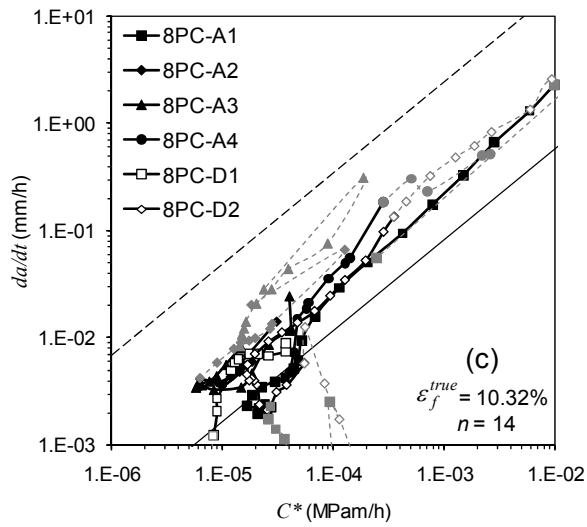
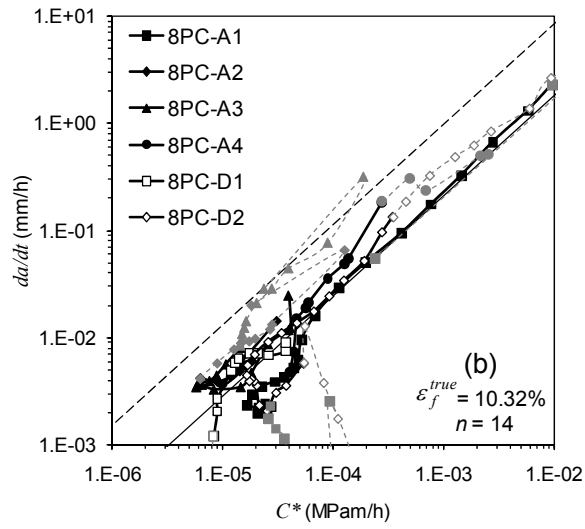
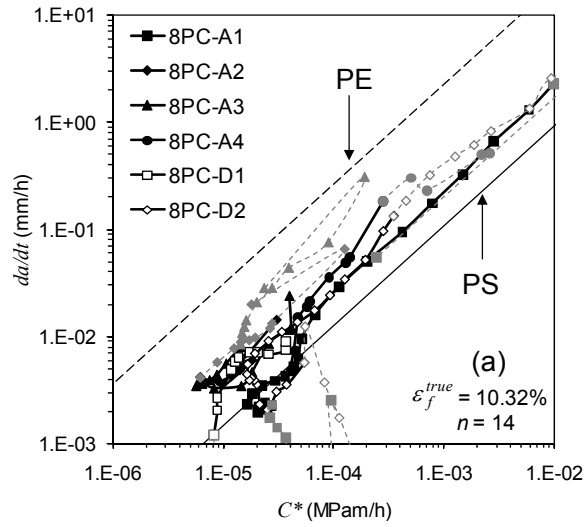


Figure 6.15: CCG predictions using the average creep stress exponent and ε_f^{true} - ROA by (a) NSW (b) NSW-MOD (c) NSWA models

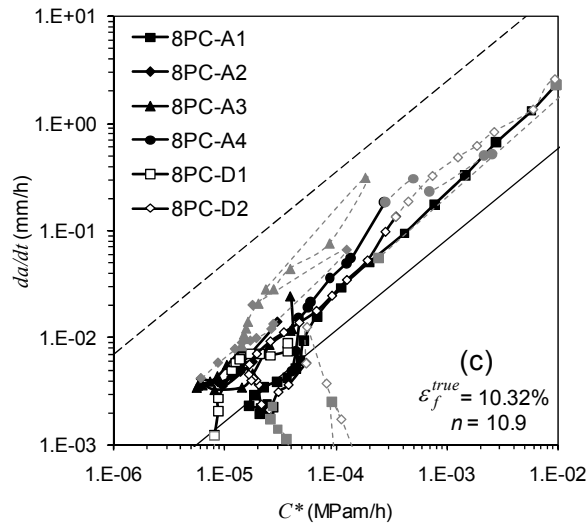
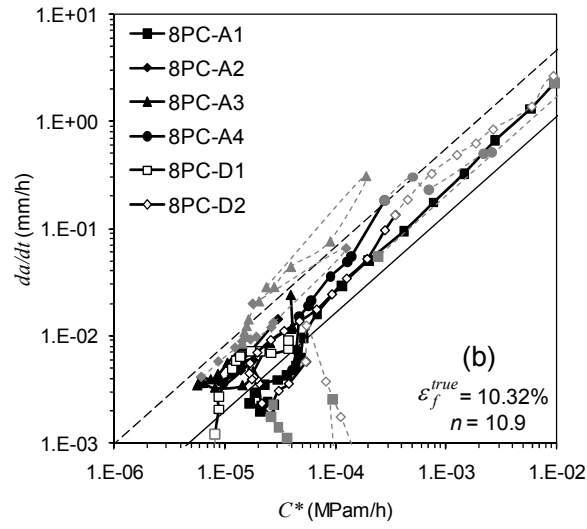
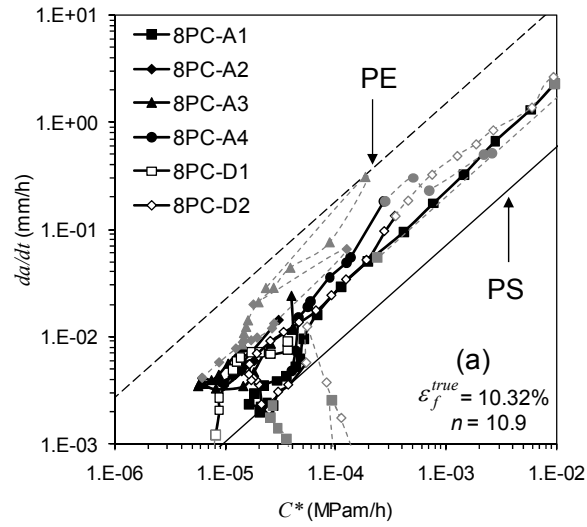


Figure 6.16: CCG predictions using the minimum creep stress exponent and ϵ_f^{true} -ROA by (a) NSW (b) NSW-MOD (c) NSWA models

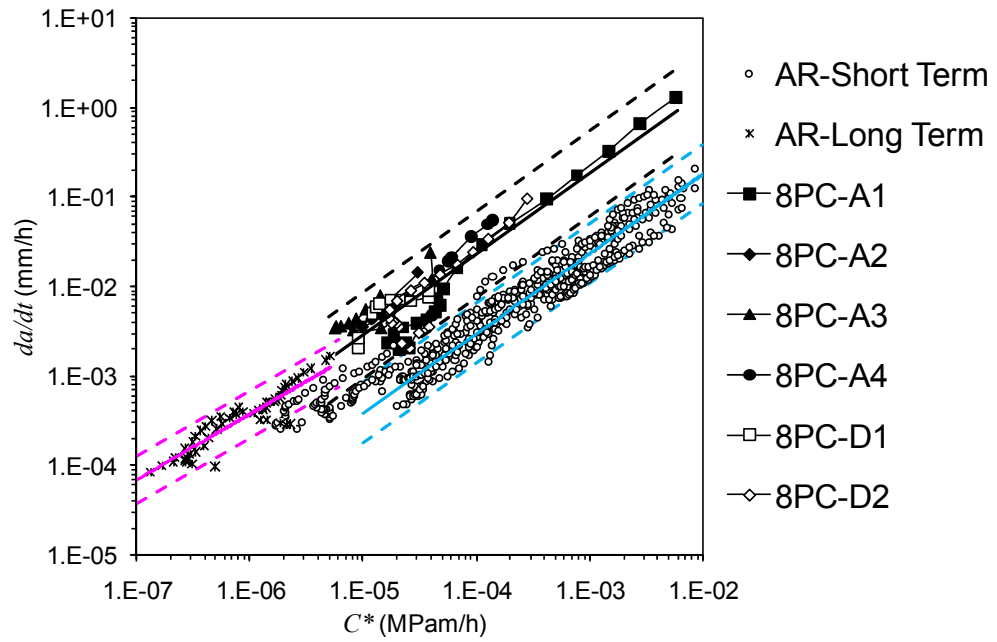


Figure 6.17: Comparison of PC data to short term and long term data on AR material

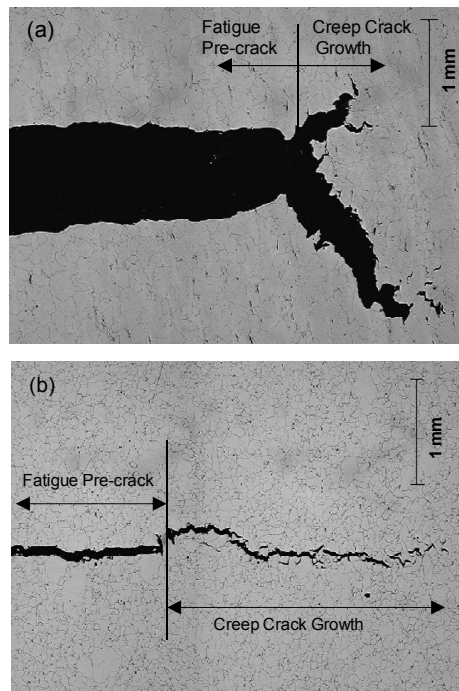


Figure 6.18: The cracking mode in (a) short term and (b) long term, CCG tests on C(T) specimen extracted from AR 316H steel (taken from [24])

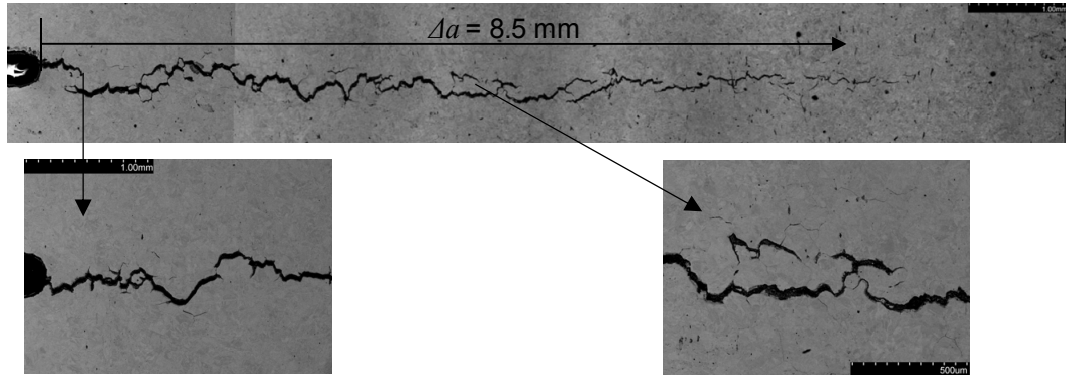


Figure 6.19: The cracking mode in PC material (specimen 8PC-A2) at mid thickness (upper image) and side groove (shown by arrows) planes

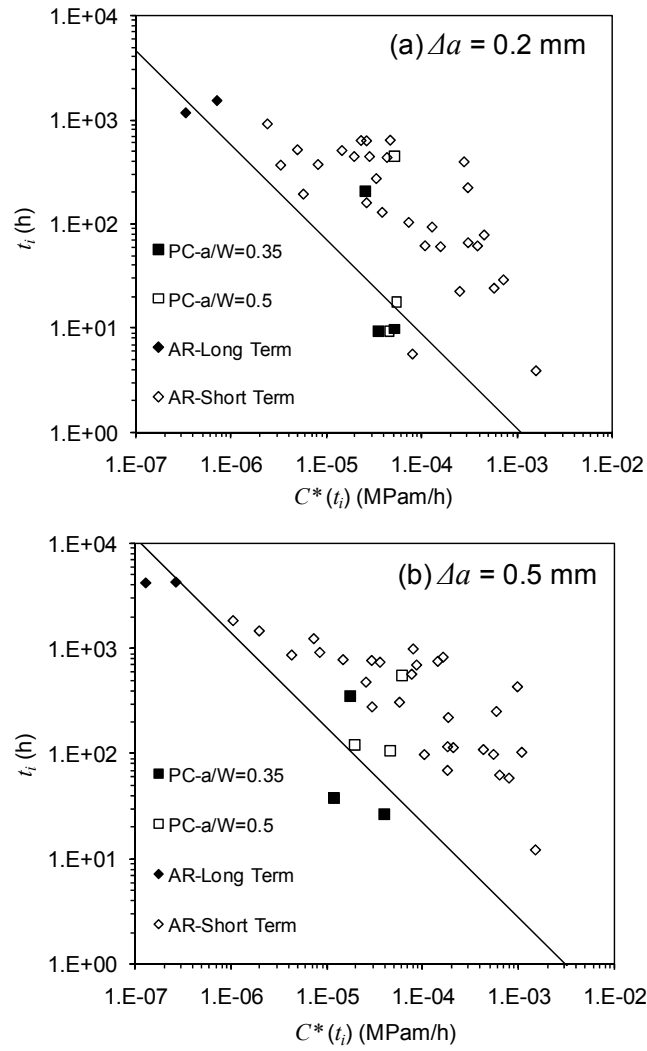


Figure 6.20: Comparison of initiation time corresponding to (a) 0.2 mm and (b) 0.5 mm, crack extension for AR and PC materials

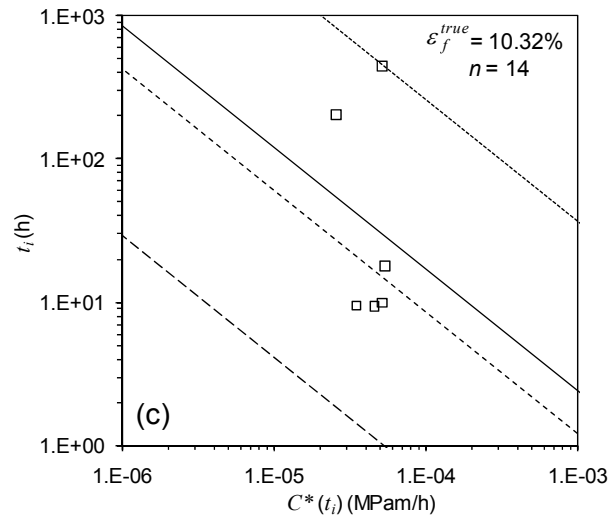
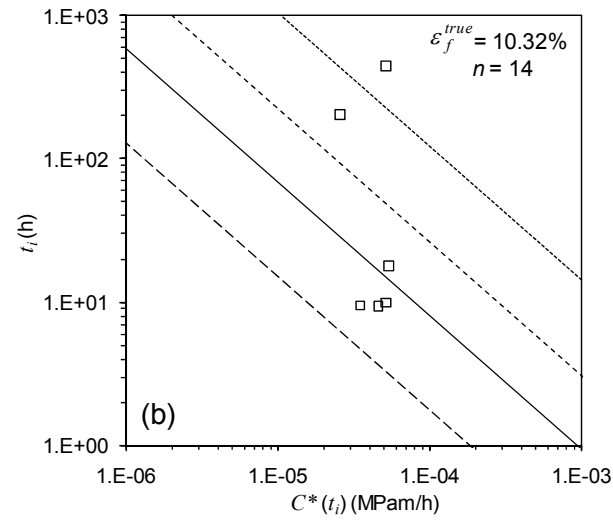
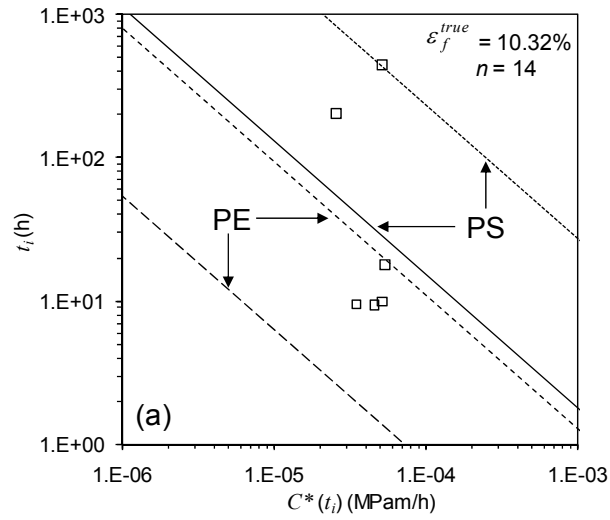


Figure 6.21: Prediction of $t_{0.2}$ trends using the average creep properties and ROA by (a) NSW (b) NSW-MOD (c) NSW-A models

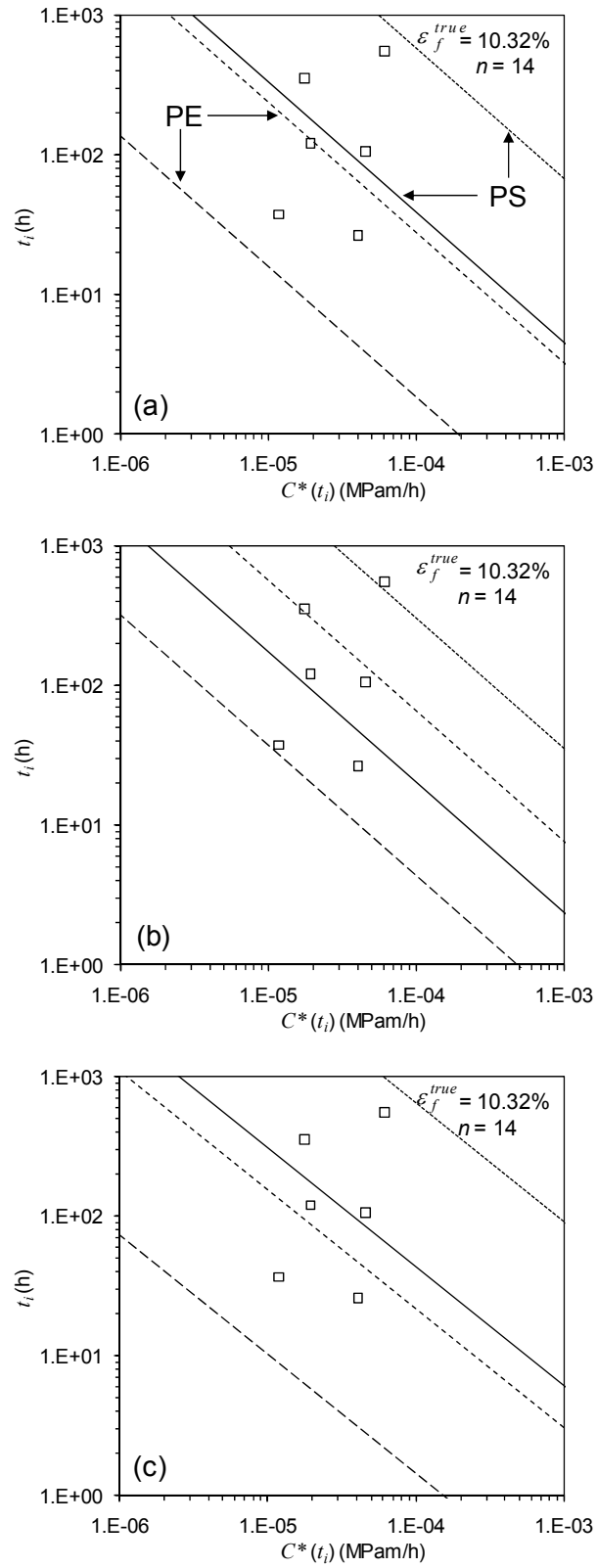


Figure 6.22: Prediction of $t_{0.5}$ trends using the average creep properties and ROA by (a) NSW (b) NSW-MOD (c) NSWa models

Chapter 7

Creep Crack Growth Modelling

7.1 Introduction

Predicting CCG is fundamentally important in high temperature components life time assessments. Finite element (FE) analysis provides a tool to enable CCG predictions to be performed and the sensitivity of key parameters to be studied. FE analysis is therefore a highly beneficial aid for understanding CCG behaviour. In this chapter CCG predictions are obtained for AR and PC materials. The influence of plasticity on the CCG behaviour of the material has been considered and taken into account in these simulations by employing a stress dependent creep ductility model for 316H SS at 550 °C presented in Chapter 5. The creep ductility trends estimated in Chapter 5 have therefore been incorporated into FE analyses to predict CCG and the results are compared to experimental short and long term data available on the AR and PC materials.

7.2 Existing CCG Modelling Approaches

Creep damage modelling in cracked geometries has been previously performed by many researches to study the rupture life of components made of engineering materials under creep deformation conditions (*e.g.* [131-136]). There are two key approaches in the literature to simulate CCG behaviour of materials using FE simulations. The first method is to simulate crack growth by element removal (*e.g.* [131, 135]) or reducing an element's load carrying capacity (*e.g.* [49, 137]) when the damage parameter, ω , attains a critical value. In the second approach when the damage reaches a critical value at two integration points in an element located ahead of the crack tip, the crack tip node releases and virtual crack propagation is modelled [48, 76, 138]. In this method, which is often referred to as node release model, the crack path has to be specified prior to CCG simulations. The defined crack path in node release model is usually along the symmetry line at the mid height of the specimen geometry [48].

An important disadvantage of the node release model compared to the first approach described above is that the crack tip angle effects are ignored by enforcing the CCG to occur along the symmetry plane at zero crack tip angle, $\theta = 0^\circ$. Another

drawback of the node release model is that once a crack tip node is released, an artificially large stress concentration is applied on the following node which may lead to unrealistic predictions made by this model. In addition, unlike the load drop technique which has the potential capability of modelling crack growth discontinuity due to the freedom of the crack path, the node release model can only simulate continuous CCG along a pre-defined crack path.

The failure criteria in previous studies using both load bearing capacity drop and node release techniques (*e.g.* [137, 138]) are based on a constant creep ductility, ε_f . However, the recent study on the creep ductility trends for 316H SS in Chapter 5 shows that employing constant creep ductility based on the high stress uniaxial creep data may only provide appropriate predictions of short term (high load) CCG trends. To provide accurate predictions of the CCG behaviour of a material, especially in intermediate and long term tests, it is essential to employ appropriate creep ductility trends, such as those of suggested in Chapter 5, in FE simulations.

7.3 Deformation, Damage and Fracture Models

The tensile and creep deformation behaviour of 8% PC and AR material at 550 °C has previously been compared in Chapter 5 and their CCG behaviour compared in Chapter 6. As discussed in Chapter 5, DSA effects are apparent in the tensile response of the AR and PC material at 550 °C. However to avoid numerical difficulties, smooth tensile curves, shown in Figure 7.1, have been implemented in FE simulations.

Creep deformation is considered to be composed of three regimes— primary, secondary and tertiary creep. The use of an average creep rate obtained directly from creep rupture data has been proposed to account for all three stages of creep. The values of average creep strain rate power law exponent n_A and coefficient A_A may vary over a wide stress range as discussed in [121]. The estimated creep properties made for AR 316H stainless steel at 550 °C shown in [121] indicate that for $\sigma/\sigma_{0.2} \geq 1.035$, $A_A = 1.24 \times 10^{-23}$ and $n_A = 7.5$, and for $\sigma/\sigma_{0.2} < 1.035$, $A_A = 6.56 \times 10^{-12}$ and $n_A = 2.3$. As shown and discussed in Chapter 5, the pre-compression process may be considered to have negligible effects on the average creep strain rate of the material. Therefore, the values of A_A and n_A for the PC material employed in FE simulations have been assumed the same as those of found for the AR material in [121]. The normalised stress level at which the switch over between short term and long term A_A and n_A occurs has also been considered the same in PC and AR materials.

The estimated creep ductility trends over a wide stress range presented in Chapter 5 for the AR material at 550 °C are implemented in FE simulations. Note that the upper shelf (US) creep ductility values specified in Table 5.6 are the average of the experimental data available within the scatter. The same stress dependent creep ductility model has also been applied in FE simulations for the PC material.

The creep ductility exhaustion approach (see Eqn (2.41)) has been used to model creep damage accumulation by employing the Cocks and Ashby [43] model to work out the multiaxial strain factor (see Eqn (2.53)). The details of the ductility exhaustion approach are given in Chapter 2-Section 2.4.1. In order to compare da/dt vs. C^* trends obtained from FE simulations to those of predicted from the constant ductility CCG models, different versions of the NSW and NSW-MOD models, which are detailed in Chapter 2-Section 2.8, are examined in this chapter.

7.4 Finite Element Model

7.4.1 Mesh Design and Simulation Matrix

Finite element analyses were conducted on a two dimensional (2D) FE model of a compact tension, C(T), specimen of width, $W = 50$ mm, thickness $B = 25$ mm and initial normalised crack length $a_0/W = 0.5$ using ABAQUS v6.11 as shown in Figure 7.2. Half of the specimen has been modelled and symmetry conditions employed. In the region of the crack path, regular square elements of size 20 μm , have been used as shown in the magnified region bounded by a box in Figure 7.2. Note that this mesh has been designed for CCG simulations only, however to study CCI a focused mesh similar to those employed in [139] will be needed. The mesh size effects have previously been examined by other researchers (see *e.g.* [137, 138]). A small geometry change analysis has been performed employing four noded continuum elements for plane stress analyses (CPS4) and “hybrid” elements (CPE4H) for plane strain analyses. Previous work has demonstrated that plane strain analyses are consistent with full three dimensional (3D) analyses [137]. Three loading cases have been considered using both AR and PC material properties, as described in Table 7.1, under both plane stress (PS) and plane strain (PE) conditions. These applied loads are corresponding to the initial stress intensity factors in the CCG tests performed on the PC material. Twelve simulations have therefore been run, using the stress dependent ductility model (upper shelf/transition/lower shelf criteria). In addition, simulations have also been run using the constant upper shelf and lower shelf (LS) creep ductility as the failure criterion for each of the cases examined for the AR material. For comparison purposes, the

constant creep ductility simulations have also been repeated on the intermediate load case for the PC material (PC2). Note that the normalised reference stress, $\sigma_{ref}/\sigma_{0.2}$, shown in Table 7.1 can be used as an indicator of the level of plasticity at the crack tip on loading. As seen in this table, the PC material exhibits smaller extent of plasticity compared to the AR, when identical loads are applied on both materials.

7.4.2 Creep Damage and Crack Growth Simulation

Following the approach in [49, 137], crack growth is simulated by reducing an elements load carrying capacity when the damage parameter, ω , attains a critical value. To avoid numerical difficulty, the stress (at a gauss point) in a damaged element is limited to a small value by switching its material response to elastic-perfectly plastic behaviour with a yield stress of 1 MPa. This is done using a user defined field (USDFLD) subroutine in ABAQUS where the creep damage parameter is also evaluated [139]. The USDFLD subroutine was also used to switch the values of A_A and n_A depending on the normalised stress value as described in Section 7.3. When the damage at the centroid of the element attains $\omega = 1.0$ then that element is considered fully damaged and the crack may advance. The analyses were run until terminated by the program when numerical difficulties were encountered. Therefore various crack extensions have been achieved in different simulations.

7.5 Creep Crack Growth Predictions

7.5.1 Specimen Loading Behaviour

In order to examine the load up response of the AR and PC materials, elastic-plastic simulations have been performed under PS and PE conditions and the results are shown in Figure 7.3. Note that in this figure, predictions of the load up behaviour of the AR and PC materials under the highest examined load level ($P = 19.5$ kN) have been illustrated and the load line displacement (LLD) predictions at initial loading under $P = 14.5$ kN and $P = 10.0$ kN are shown in dashed and dotted lines, respectively. It can be seen in this figure that for a given load, the load line displacements predicted by PS simulations are larger than the PE ones. Figure 7.3 shows that the load vs. LLD response for the PC material is almost linear under the load levels examined, whereas some non-linearity is observed in the AR material, especially at higher load levels. An extent of non-linear displacement of around two times larger is predicted by PS simulations in the AR material, compared to the PC, at the highest load level examined ($P = 19.5$ kN).

The prediction of the plastic zone size, r_p , in the AR and PC material has been examined under plane stress and plane strain conditions and the results for the highest applied load ($P = 19.5$ kN) are shown in Figure 7.4(a)–(d). It can be observed in this figure that the plastic zone size predicted for the AR material is around twice of the PC material under plane stress conditions. Further seen in Figure 7.4 is that the predicted plastic zone size for each material is approximately three times larger under PS conditions compared to the PE, as expected.

7.5.2 Load Line Displacement and Crack Growth Behaviour

The predicted LLD and crack length variations against time have been shown for the AR material in Figure 7.5 and Figure 7.6, respectively. Similar predictions have also been made for the PC material as shown in Figure 7.7 and Figure 7.8. The results presented in these figures are from the CCG simulations performed using the stress dependent ductility criteria. Note that the total load line displacements, Δ , measured during the CCG simulations (after applying the full load) are displayed in Figure 7.5 and Figure 7.7.

It can be observed in Figure 7.5 that for a given time, the largest LLD for the AR material is predicted for the case where the highest load is applied and the smallest LLD is found in the simulations with the lowest applied load level. Further seen in this figure is that the simulations performed under PS conditions terminated after a longer period compared to those of conducted under PE conditions. As seen in Figure 7.6, larger crack propagations are generally predicted under PE conditions compared to the PS, at the end of CCG simulations. Note that the jumps in the LLD and crack growth predictions presented for the AR material in Figure 7.5 and Figure 7.6 are due to relatively large frequencies requested for the field outputs in each of these simulations to optimise the output file size and also data reductions have been applied on the raw data post-simulation. Furthermore, not shown for brevity but observed in FE simulation results is that the damage formation may occur in the elements ahead of the crack tip (which may replicate the micro crack formation) during CCG simulations. The micro cracks ahead of the crack tip have been considered in the analyses by taking their coordinate as the crack tip position. As a result, the jumps in the crack length predictions for the AR1 case performed under PS and PE conditions in Figure 7.6 may be associated with the formation of these micro cracks ahead of the crack tip.

The variation of the LLD during CCG simulations presented for the PC material in Figure 7.7 shows that larger displacements along the load line are obtained from the PS simulations compared to the PE for a given applied load and the LLD for a given

time increases with increasing the applied load. Shown in Figure 7.8 is that under the applied loads of $P = 19.5$ kN and $P = 14.5$ kN, the predicted crack lengths from PE simulations are found larger than those from the PS at the end of the CCG simulations. However for the lowest load level examine, $P = 10$ kN, the final crack length in PE simulations is found smaller than that of obtained from the PS.

The LLD and crack extension predicted by 2D FE simulations under plane stress and plane strain conditions using the stress dependent ductility criteria have been compared to the existing experimental data on the PC material where the data were available from the tests performed at a similar initial stress intensity factor. Note that no experimental data was available on the AR material for the specimen dimensions and loading conditions examined in FE simulations. Therefore, direct comparison between the LLD and crack growth data from the FE to the test data couldn't be made for the AR material. Also note that the designed mesh employed in this work (see Figure 7.2) is made to study the CCG behaviour of the material only. Thus to exclude the initiation effects in direct comparisons between the FE predictions and the experimental data for PC material, the CCG time, LLD and crack extensions beyond 0.5 mm crack growth are presented in Figure 7.9 and Figure 7.10. The parameters t_g , Δ_g and Δa_g shown in Figure 7.9 and Figure 7.10 are subtracted by their corresponding values at 0.5 mm crack extension (beyond which the mesh effects on the creep crack growth predictions are considered negligible) and the subscript "g" has been used to indicate that the analysed data are for the CCG beyond 0.5 mm.

The LLD variations (subsequent to CCI) predicted by the FE simulations under the initial stress intensity factor of around $K(a_0) = 34$ MPa $\sqrt{m}$ are compared to the experimental data available from a CCG test performed on 8PC-D1 specimen under a similar $K(a_0)$, the results of which are previously discussed in Chapter 6. This comparison, presented in Figure 7.9, shows that the experimental data fall in between the PS and PE predicted trends and the test data from 8PC-D1 are found closer to the PE predictions for the majority of the CCG test duration. The comparison of the crack extension subsequent to CCI in Figure 7.10 shows that the experimental crack propagation for the given load and geometry can be successfully predicted by the FE simulations performed under PE conditions.

Similar comparisons have been made between the Δ_g and Δa_g obtained from the experiments performed on specimens 8PC-A4 and 8PC-D2 under $K(a_0) = 25$ MPa $\sqrt{m}$ to those of predicted by FE simulations at the same initial stress intensity factor. As seen in Figure 7.11 and Figure 7.12, similar trends have been predicted by FE

simulations compared to the LLD and crack extension data (subsequent to CCI) available from these two tests. However, some differences can be observed in the CCG times at which the exponential acceleration in the $\dot{\Delta}_g$ and $\dot{\Delta}a_g$ start to occur. This can be related to the mesh size and crack tip plasticity effects in FE simulations. It has been shown in [138] that the predicted crack growth rates will increase with decreasing the mesh size. By employing a finer mesh in FE simulations, the local stress and strain levels will increase in the vicinity of the crack tip and therefore the time required for the damage to reach a critical value decreases. Furthermore, the formation of plasticity at the crack tip will decrease the stress levels and subsequently reduces the creep strain rates.

7.5.3 C^* Validity Criteria

The FE data from CCG simulations have been analysed according to ASTM E1457 standard [71] and the C^* validity criteria, as detailed in Chapter 4-Section 4.1.7, have been applied to distinguish the valid data points which can be correlated with the C^* fracture mechanics parameter. The C^* validity criteria for the data obtained from simulations using the stress dependent ductility model have been illustrated in Figure 7.13 and Figure 7.14 for the AR and PC materials, respectively. The valid data points in these two figures have been shown in black symbols whereas the invalid points are shaded in grey. Note that similar to the data analyses in Chapter 6, the plastic term of the displacement rate has been considered small and thus neglected in creep displacement rate calculations.

As seen in Figure 7.13, values of $\dot{\Delta}^c / \dot{\Delta}^T$ range from 0.9–1.0 for the FE data from the simulations performed on the AR material under PS conditions. This indicates that elasticity has small contribution in the LLD rate for the results obtained from PS simulations at different load levels. Unlike PS results on the AR material, the data obtained from FE simulations under PE conditions exhibit significant fluctuations above and below $\dot{\Delta}^c / \dot{\Delta}^T = 0.5$ which leads to a few valid data points being remained for AR1 and AR2 simulations. Also seen in Figure 7.13 is that no valid data points are found for the simulation performed on the AR material under PE conditions at a relatively high load, AR3. Note that the fluctuations observed in the creep to total LLD rate behaviour under PE conditions in Figure 7.13 may be related to the lack of enough data points due to the large field output frequencies requested in simulations.

A summary of the C^* validity criteria for the FE data on the PC material is shown in Figure 7.14. It can be observed in this figure that relatively large values of $\dot{\Delta}^c / \dot{\Delta}^T$ ranging from 0.65–0.90 are found in the numerical tests performed under PS

conditions. Comparison of the $\dot{\Delta}^c / \dot{\Delta}^T$ for the PS simulations on the AR and PC materials shows that although relatively large values of creep to total LLD rate of greater than 0.5 are observed in both materials, but the contribution of elasticity on the LLD rate is more significant in the PC material compared to the AR.

Figure 7.14 shows that rapid drop in $\dot{\Delta}^c / \dot{\Delta}^T$ trends is generally seen under PE conditions for the majority of test duration in all the cases examined on the PC material. Thus only a few data points from the PE predictions are found valid for correlating the CCG rate with C^* parameter. Also seen in Figure 7.14 is that the $\dot{\Delta}^c / \dot{\Delta}^T$ falls and remains below 0.5 for the majority of the test duration in PC3 simulation performed under PE conditions. Therefore no data points are found valid in this high load CCG simulation.

7.5.4 Comparison of FE Predictions to NSW Models

In order to interpret the CCG trends obtained from FE simulations, it is important to provide an approximate measure of the extent of plasticity at the crack tip. As explained previously in Chapter 2-Section 2.6.3.3, the normalised reference stress $\sigma_{ref}/\sigma_{0.2}$ indicates the significance of crack tip plasticity. Therefore, the variation of the normalised reference stress during CCG simulations is plotted against C^* and shown in Figure 7.15 and Figure 7.16 for the FE data on the AR and PC material, respectively. Note that only the valid data points are shown in these two figures. In Figure 7.15 and Figure 7.16, the CCG behaviour at the data points corresponding to the normalised reference stress values of much larger than unity ($\sigma_{ref}/\sigma_{0.2} \gg 1$) is considered to be substantially influenced by plasticity. The widespread plasticity decreases the specimen constraint and subsequently leads to a drop in the CCG rates.

The FE predictions are compared to the constant creep ductility NSW models' predictions in Figure 7.17 for the AR material under plane stress conditions. Details of the NSW models are given in Chapter 2-Section 2.8. The characteristic distance r_c employed in NSW predictions has been taken as 0.05 mm in this study. The FE CCG data from the simulations performed under different load levels using the stress dependent creep ductility model have been analysed by employing $n_A = 7.5$ in all calculations.

As expected the US FE data fall close to or upon the NSW-US model's prediction. Note that although the C^* validity criteria are satisfied for all the FE data points presented, some tails may still remain in $d\Delta/dt$ vs. C^* trends as may also be observed in the experimental data. The LS FE and NSW-LS predictions generally follow the

same trend, however the AR1-LS and AR2-LS data fall above the NSW prediction which may be attributed to the transient creep and also crack tip angle effects. The initial data for AR3-LS are coincident with the NSW-LS model prediction. As seen in Figure 7.15(a), large values of the normalised reference stress ($\sigma_{ref}/\sigma_{0.2}$) are observed in this simulation especially in the higher C^* region. This indicates a significant plasticity effect at the crack tip which leads to a drop in the CCG rates in the latter half of the AR3-LS simulation data. The AR2 and AR3 data from the stress dependent model fall exactly on the NSW-US model because the normalised reference stress for these valid points are relatively low as seen in Figure 7.15(a). The da/dt vs. C^* data for AR1 simulation start at the NSW-LS prediction, but then fall towards and below the NSW-US prediction line. As seen in Figure 7.6 the crack extension achieved in the simulation for AR1 under PS conditions was significantly higher than that of AR2 and AR3. Therefore, even though the applied loads in specimens AR2 and AR3 were higher than that of AR1-PS, the $\sigma_{ref}/\sigma_{0.2}$ values shown in Figure 7.15(a) are higher in AR1-PS simulation and significantly greater than unity for the majority of the data points. This indicates a significant plasticity effect which leads to a reduction in the CCG rate as the crack grows.

In order to account for the crack tip angle effects in the CCG behaviour, the plane stress NSW-MOD prediction line has also been included in Figure 7.17. A study of the angular distribution of triaxiality in [108] has shown that the maximum value of the angular function in NSW-MOD equation (see Eqn (2.117)) would be artificially large for $n < 5$. Thus the NSW-MOD predictions are considered valid for materials with the power law creep exponent of $n \geq 5$. As a result, the predictions using NSW-MOD model have only been made for the US ductility and creep strain rate properties (*i.e.* $n_A = 7.5$). As seen in Figure 7.17, NSW-MOD-US line provides a trend similar to the NSW-US, however the CCG rates predicted by this model are slightly higher. It can be observed in Figure 7.17, that the predictions provided by NSW-MOD-US line are coincident with the FE predictions from AR1-US and AR2-US. This implies that considering the crack tip angle effects for these simulations will provide a better prediction of the CCG trends. Further seen in Figure 7.17 is that the CCG trend predicted by AR3-US can be better described by NSW-US than NSW-MOD-US.

7.5.5 Comparison of FE Predictions to Experimental Data

The plane stress FE predictions, shown in Figure 7.17, are compared to plane strain predictions in Figure 7.18. Also included in this figure are the mean (solid lines) and upper bound (UB) and lower bound (LB) (shown in dashed lines) fits to the short and

long term AR experimental CCG data which were previously shown and discussed in Chapter 6. Comparing Figure 7.17 and Figure 7.18 it can be seen that the mean fit to the short term AR experimental data falls very close to the NSW-PS-US line and the UB fit to the short term AR data is coincident with NSW-MOD-PS-US prediction line. It was observed in Figure 7.17 that the plane stress FE data at high C^* values fall below the NSW-PS-US prediction, however Figure 7.18 shows that FE predictions made under PS conditions fall upon the LB fit to the short term AR experimental data.

All FE predictions in Figure 7.18 fall between the UB fit to the long term data and the LB fit to the short term data. Also there is a general trend of increasing CCG rates as C^* decreases, at relatively low C^* values, as also seen in the experimental data, except for the AR-PS-US data which follow the short term AR material trends extrapolated to low C^* values in Figure 7.18, as expected. This implies that the current modelling approach can successfully simulate the transition behaviour between the short term and long term CCG trends observed in the experimental data. As previously shown in Figure 7.13, a limited number of valid data points are achieved from the PE simulations when the C^* validity criteria are applied, however some trends from the valid PE data can be observed in Figure 7.18. The AR1-PE-US and AR2-PE data, which are at relatively small C^* values, fall on the UB line fitted to the long term AR data. The AR3-PE-US, AR2-PE-US and AR1-PE data fall close to the LB long term trend line, which is also close to the UB short term trend line. Note that the CCG rate of the AR2-PE data is higher than that of AR1-PE, for a given C^* , due to the high values of $\sigma_{ref}/\sigma_{0.2}$ in the valid data available for AR1-PE as seen in Figure 7.15(b).

In Figure 7.19 the PC predictions are compared to the AR and PC trend lines fitted to the experimental data (as previously shown in Chapter 6). All FE predictions shown fall within the UB and LB fits to the experimental data available on the PC material. The CCG rate predictions from the stress dependent creep and ductility model decrease as C^* increases, and tends from the upper bound fit to the PC data (if extrapolated) to the LB fit to the PC data. Figure 7.19 shows that the PS data for the three loads considered (PC1-PS, PC2-PS and PC3-PS) are approximately coincident for C^* ranging between 1×10^{-5} to 1×10^{-4} MPamh⁻¹, with the small differences corresponding to the relative $\sigma_{ref}/\sigma_{0.2}$ values of the FE data at these C^* values as seen in Figure 7.16. For this range of C^* , the PE CCG predictions from PC2-PE are effectively the same as the PC2-PS predictions, indicating a balance between the increase in CCG rate due to higher triaxiality in PE and a decrease in CCG due to the relatively large $\sigma_{ref}/\sigma_{0.2}$ values (see Figure 7.16) for these data. The data for PC1-PE however have relatively low $\sigma_{ref}/\sigma_{0.2}$ values (see Figure 7.16), and hence the CCG rate falls along the UB fit to the PC data

(if extrapolated). As seen in Figure 7.16, the data for PC2-PE-US have relatively low $\sigma_{ref}/\sigma_{0.2}$ values and the CCG rate predictions fall along the mean fit to the experimental PC data in Figure 7.19. Also seen in this figure is that the lower shelf constant creep ductility predictions from PC2-PS-LS data fall close to the low load (and $\sigma_{ref}/\sigma_{0.2}$) predictions from the stress dependent model for C^* values ranging between 1×10^{-6} to 1×10^{-5} MPamh⁻¹. Finally, it can be observed in Figure 7.19 that the predictions from PC2-PS-US fall along the LB fit to the experimental PC data (if extrapolated), as expected.

A comparison of the AR and PC data, from the stress dependent model, at different loads is shown in Figure 7.20(a) and Figure 7.20(b) for plane stress and plane strain conditions, respectively. As seen in Figure 7.20(a), the plane stress CCG rate of the PC material is on average around an order of magnitude faster than the AR at different loads for a given C^* . This is consistent with the experimental observations shown in Chapter 6. Figure 7.20(b) shows that this factor between the CCG rate of AR and PC materials may be reduced for plane strain conditions, however there are insufficient data available at present to deduce such trends.

The influence of plasticity alone on the CCG rate may be seen in Figure 7.21 by comparing AR2-PS-US and PC2-PS-US in which the FE predictions are made at a given load ($P = 14.5$ kN) using the same constant US ductility criteria for both AR and PC materials but smaller extents of plasticity are expected in the PC material due to hardening effects. As seen in this figure a marginally higher CCG rate for a given C^* is observed for the PC material which has lower $\sigma_{ref}/\sigma_{0.2}$ values (see Figure 7.15(a) and Figure 7.16). Similar trends are also seen between AR2-PS-LS and PC2-PS-LS. This figure shows that for each of the materials examined, a higher CCG rate at a given C^* is predicted using LS ductility compared to that of obtained from the US due to smaller creep strain at failure, ϵ_f , at the lower shelf. Furthermore, comparing the data from AR2-PE-US and PC2-PE-US in Figure 7.21 it can be seen that if the trends are extrapolated, the CCG rate is predicted to be higher for the PC material with a factor similar to that of predicted by the stress dependent ductility model under plane stress conditions (Figure 7.20(a)). Though, not enough valid data points are available from AR2-PE-US to deduce a trend for this numerical test.

7.6 Summary

- A finite element model has been developed to predict the CCG behaviour of the AR and PC materials, using a novel stress dependent creep strain rate and ductility approach.

- FE predictions are validated against analytical models and the experimental data.
- Good agreement has been found between the CCG rate vs. C^* predictions and the experimental data.
- The new stress dependent model has enabled the transition from high CCG rates to lower CCG rates at low and high C^* values, respectively, to be predicted (at $C^* \approx 1 \times 10^{-5} \text{ MPamh}^{-1}$).
- FE results show that the long term CCG behaviour of the AR material can be predicted numerically by employing the stress dependent creep ductility model.
- By comparing the FE results from the stress dependent and constant creep ductility models' predictions, the influence of crack tip plasticity (indicated by the normalised reference stress variation against C^*) and creep ductility failure criterion on the CCG behaviour of the material has been determined.
- The accelerated CCG rate in PC compared to AR material observed experimentally has been successfully predicted by the FE model.
- This model can be used to predict the CCG behaviour of the AR and PC materials over a wide stress range.

7.7 Tables

Table 7.1: Summary of C(T) specimens modelled and loading details

ID	P (kN)	$K(a_0)$ (MPa $\sqrt{m}$)	$\sigma_{ref}/\sigma_{0.2}$ PS	$\sigma_{ref}/\sigma_{0.2}$ PE
AR1	10.0	17.3	0.53	0.37
AR2	14.5	25.0	0.77	0.53
AR3	19.5	33.6	1.03	0.71
PC1	10.0	17.3	0.35	0.24
PC2	14.5	25.0	0.50	0.35
PC3	19.5	33.6	0.67	0.46

7.8 Figures

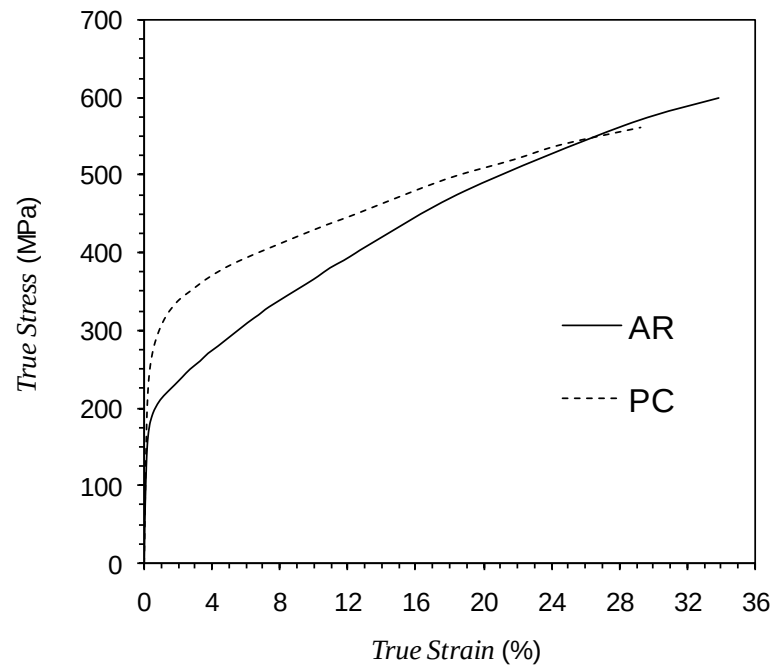


Figure 7.1: Smooth tensile curves of the AR and PC materials at 550 °C

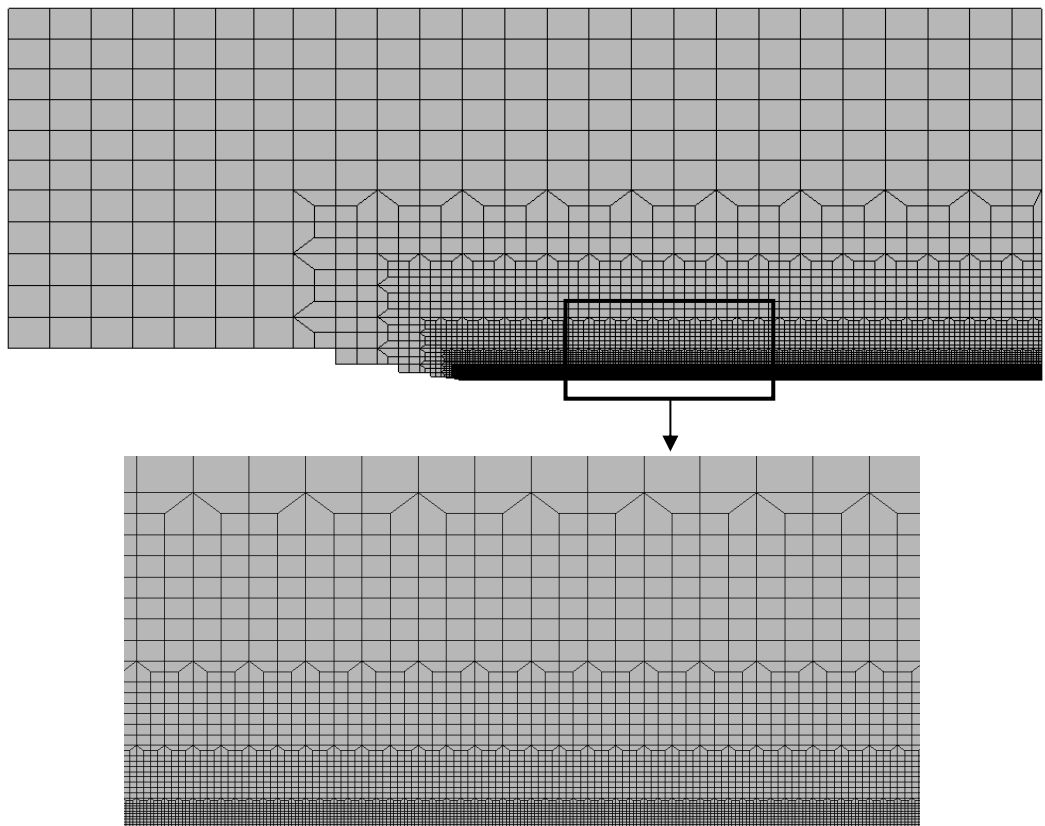


Figure 7.2: Illustration of the designed mesh for CCG simulations

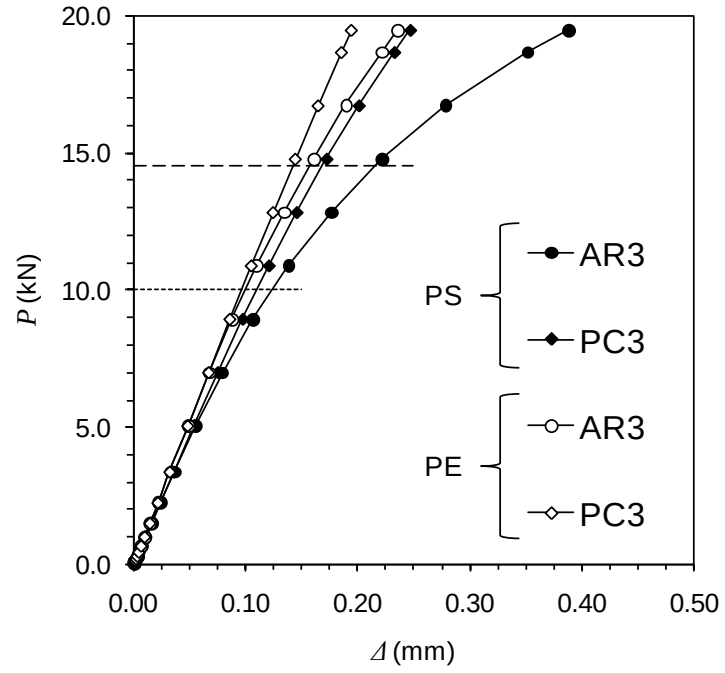


Figure 7.3: Load up response of the AR and PC materials predicted under plane stress and plane strain conditions

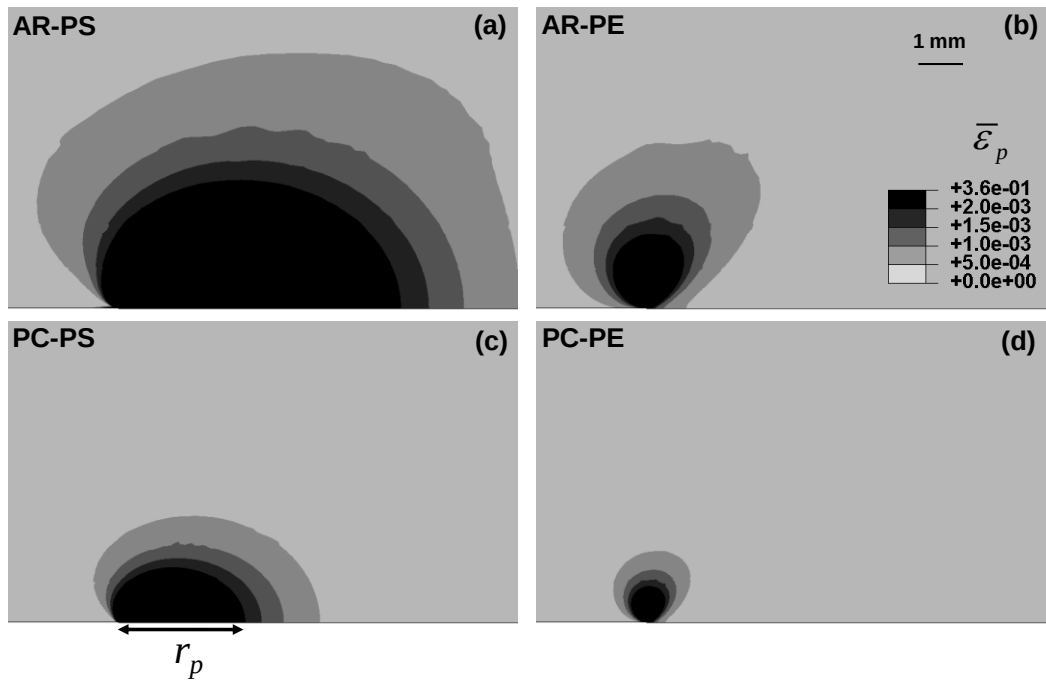


Figure 7.4: The plastic zone size predictions for (a) AR-PS (b) AR-PE (c) PC-PS and (d) PC-PE materials

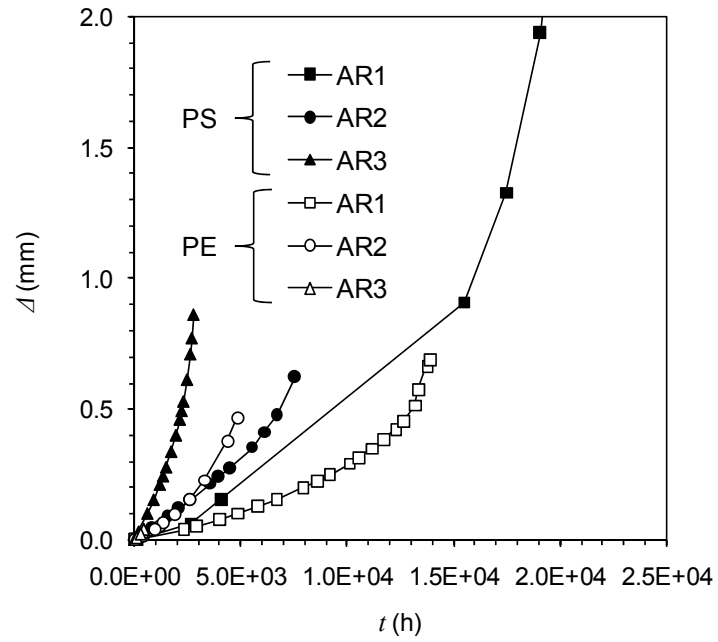


Figure 7.5: Prediction of the LLD for the AR material during CCG simulations performed under plane stress and plane strain conditions

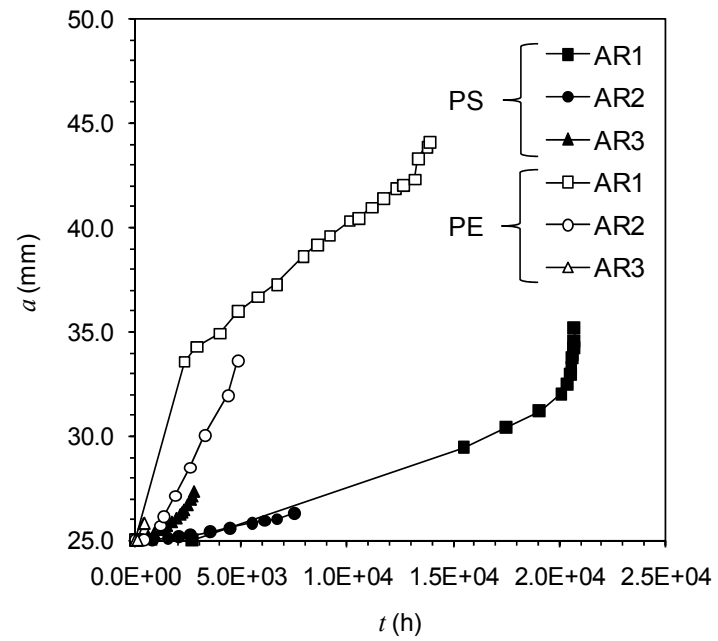


Figure 7.6: Crack length predictions for the AR material during CCG simulations performed under plane stress and plane strain conditions

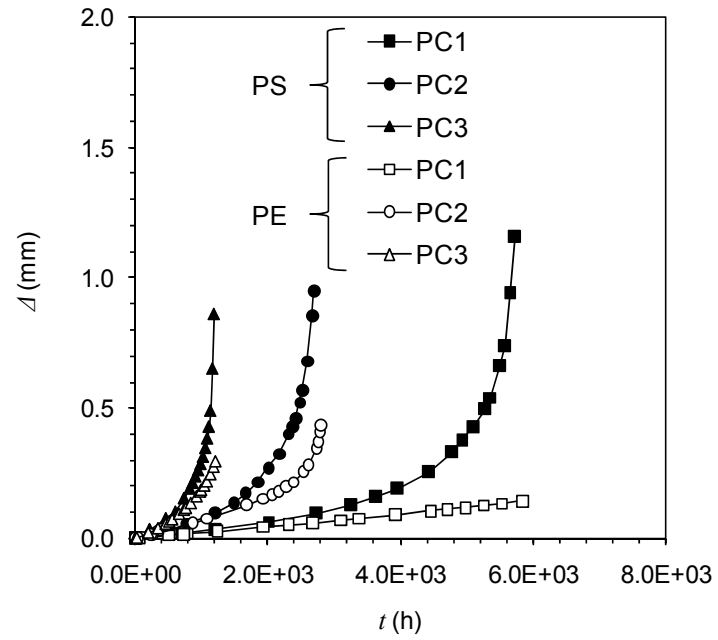


Figure 7.7: Prediction of the LLD for the PC material during CCG simulations performed under plane stress and plane strain conditions

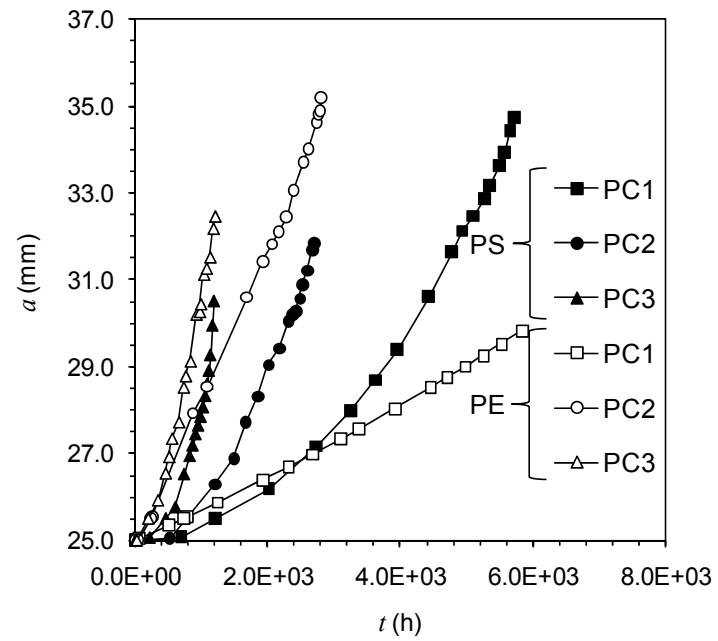


Figure 7.8: Crack length predictions for the PC material during CCG simulations performed under plane stress and plane strain conditions

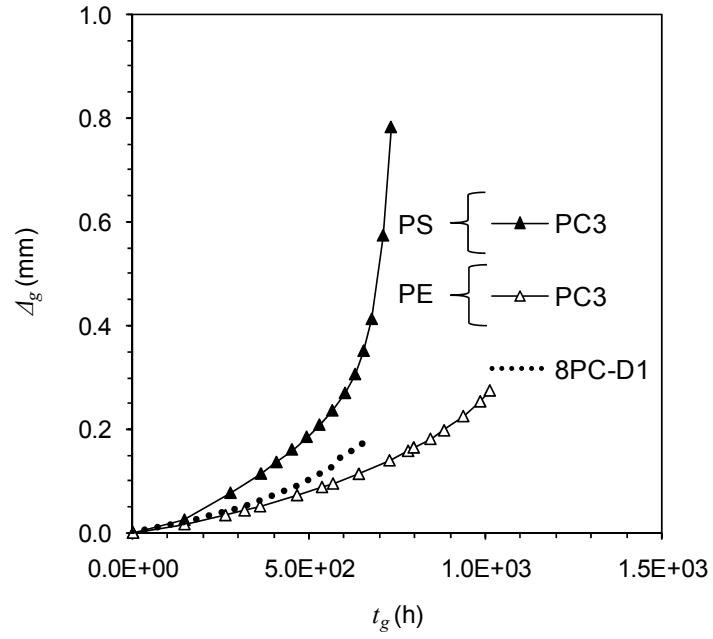


Figure 7.9: Comparison of the LLD predicted by FE simulations to the experimental data, under the initial stress intensity factor of around $K(a_0) \approx 34 \text{ MPa}\sqrt{\text{m}}$

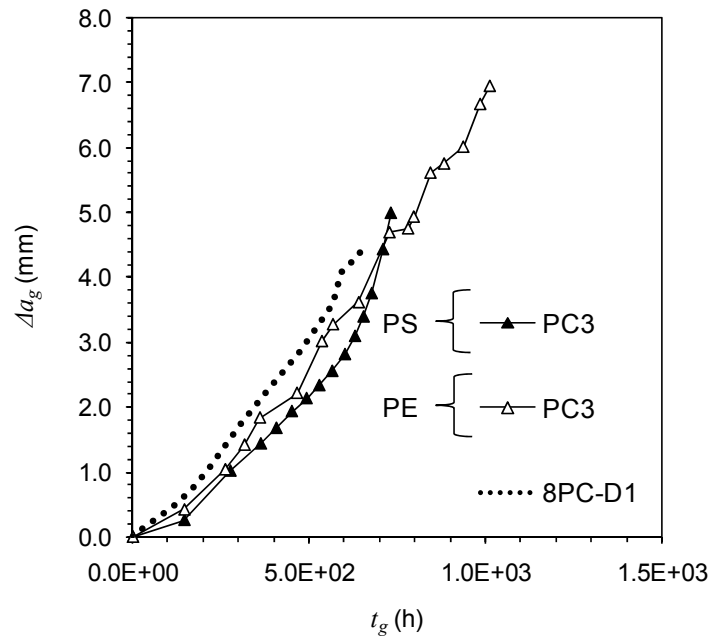


Figure 7.10: Comparison of the crack extension predicted by FE simulations to the experimental data, under the initial stress intensity factor of around $K(a_0) \approx 34 \text{ MPa}\sqrt{\text{m}}$

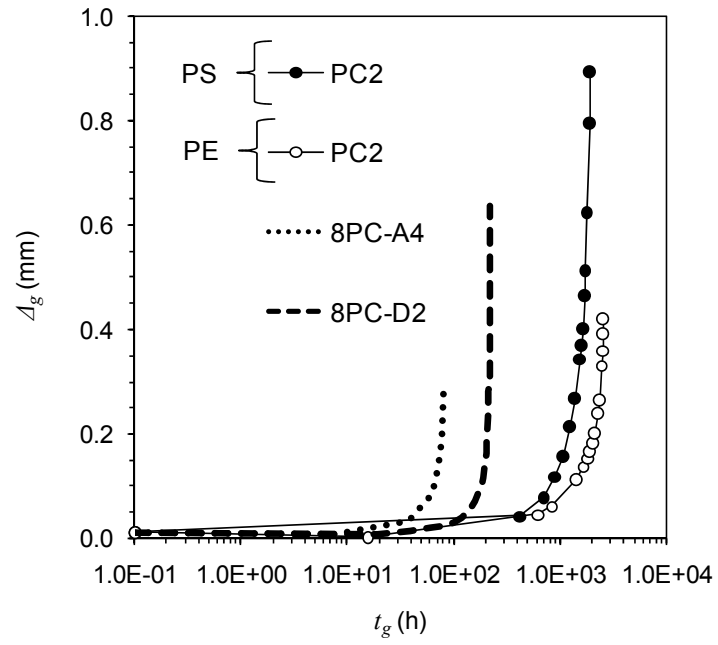


Figure 7.11: Comparison of the LLD predicted by FE simulations to the experimental data, under the initial stress intensity factor of $K(a_0) = 25 \text{ MPa}\sqrt{\text{m}}$

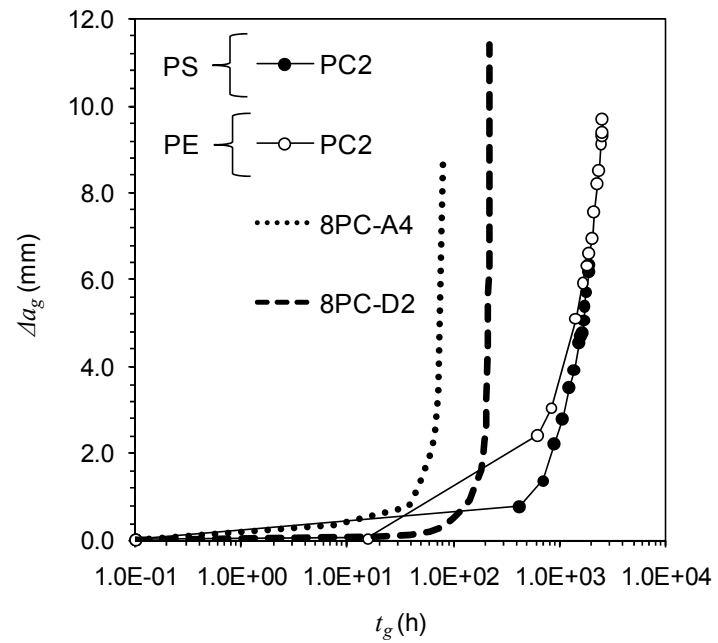


Figure 7.12: Comparison of the crack extension predicted by FE simulations to the experimental data, under the initial stress intensity factor of around $K(a_0) = 25 \text{ MPa}\sqrt{\text{m}}$

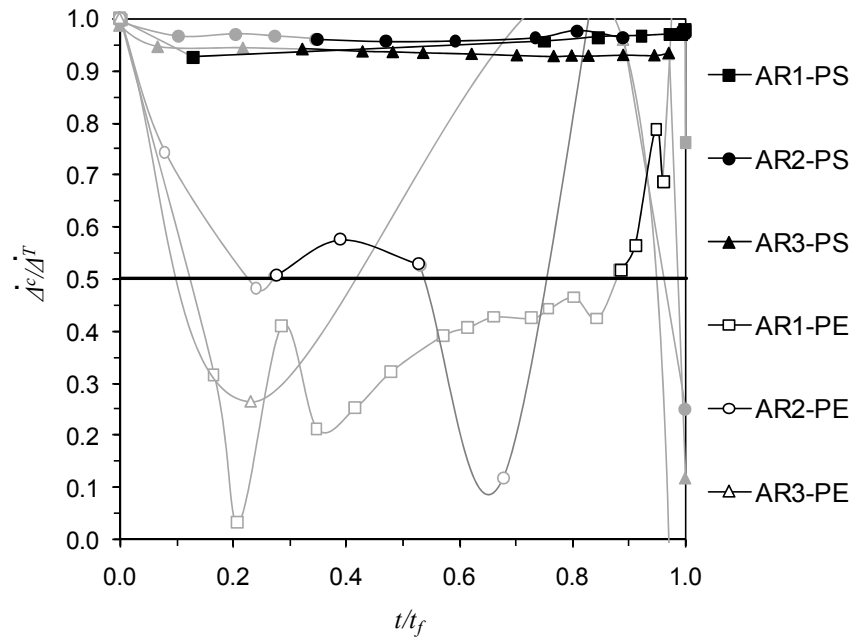


Figure 7.13: Illustration of ASTM E1457 C^* validity criteria for the FE results on the AR material

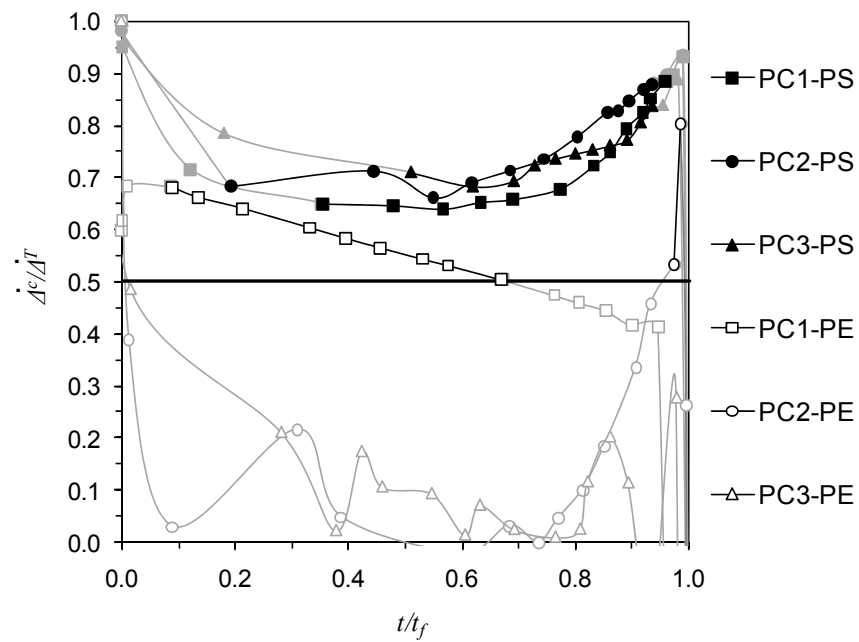


Figure 7.14: Illustration of ASTM E1457 C^* validity criteria for the FE results on the PC material

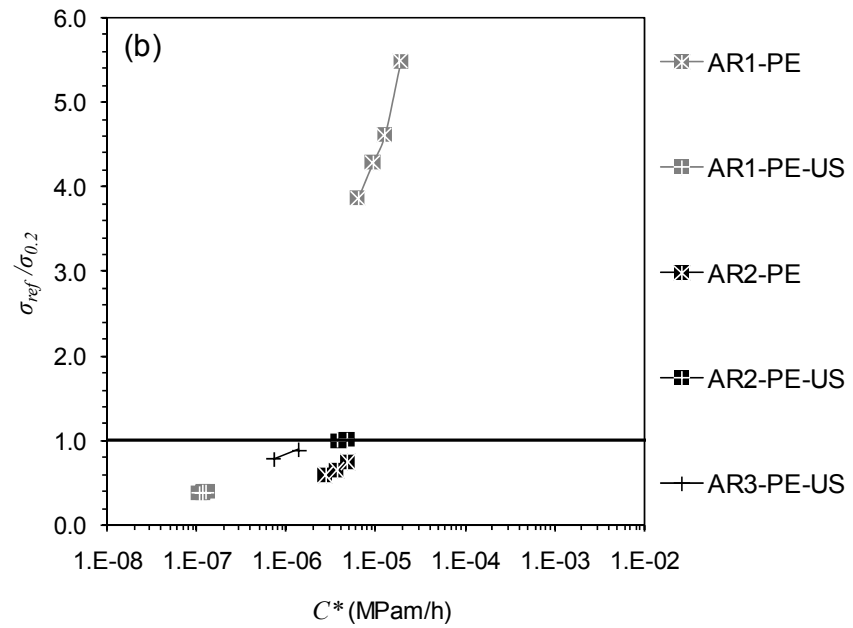
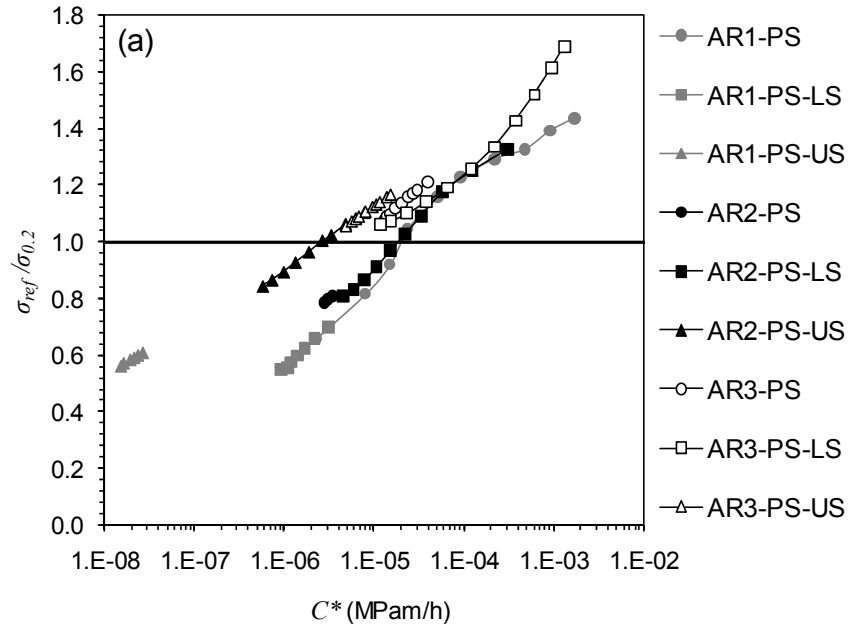


Figure 7.15: Variation of the normalised reference stress during CCG numerical tests performed on the AR material under (a) Plane stress and (b) plane strain conditions

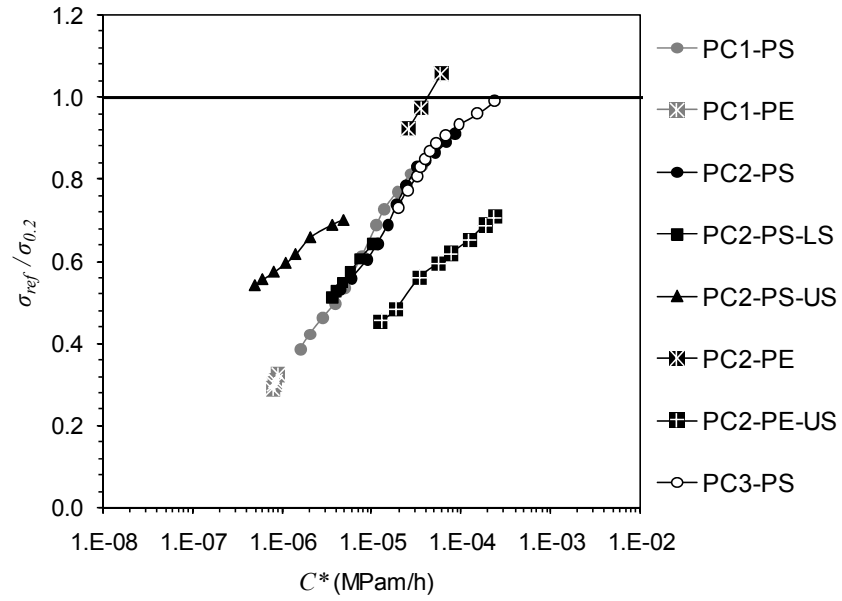


Figure 7.16: Variation of the normalised reference stress during CCG numerical tests performed on the PC material under plane stress and plane strain conditions

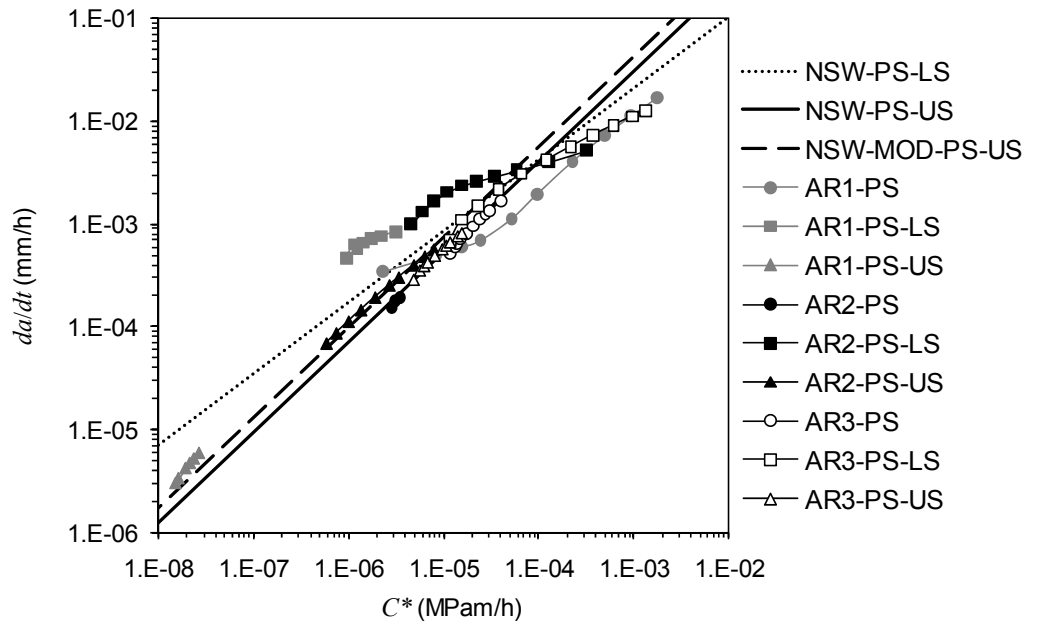


Figure 7.17: Comparison of the CCG predictions from the FE to the NSW and NSW-MOD models for the AR material under plane stress conditions

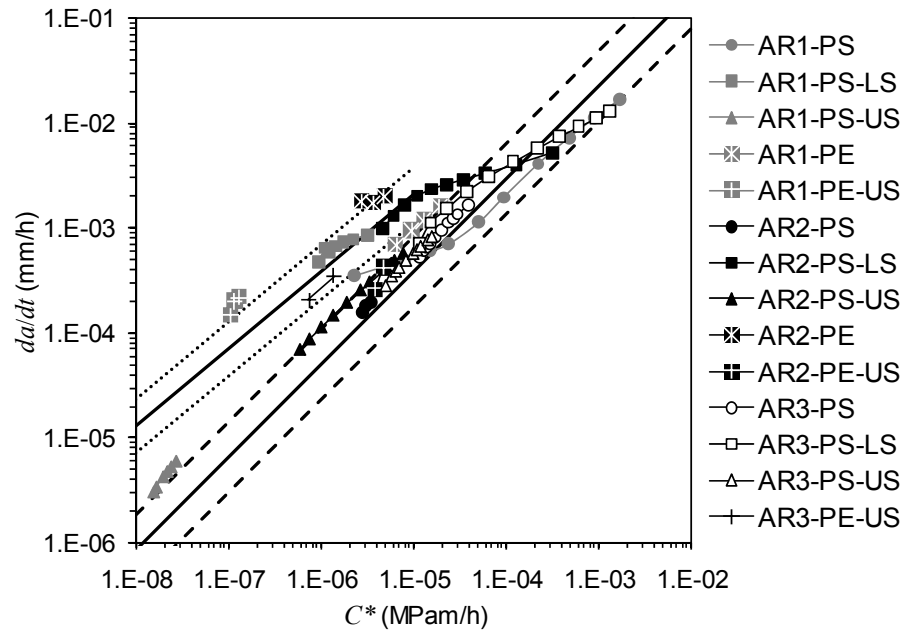


Figure 7.18: Comparison of CCG predictions to experimental data for AR material

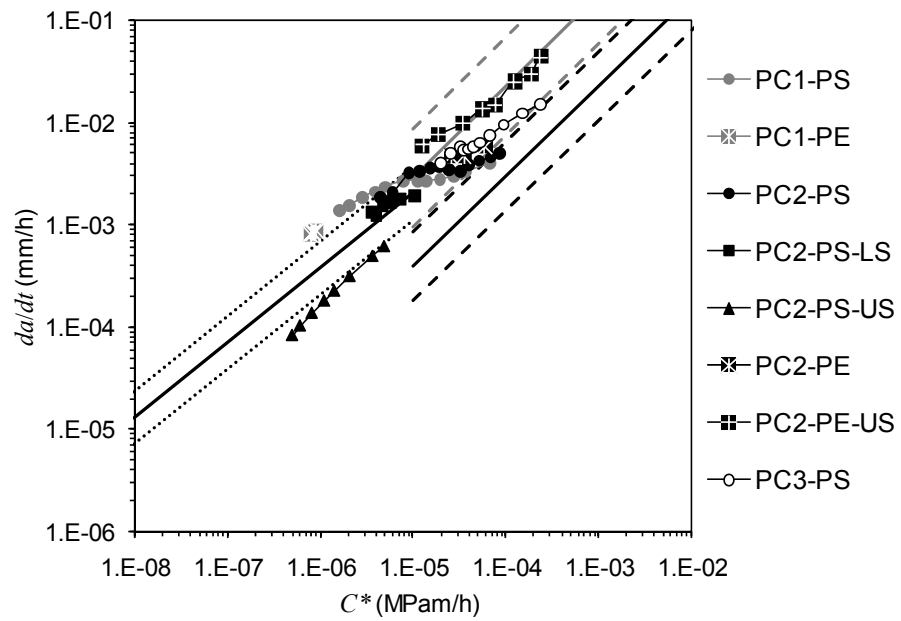


Figure 7.19: Comparison of CCG predictions to experimental data for PC material

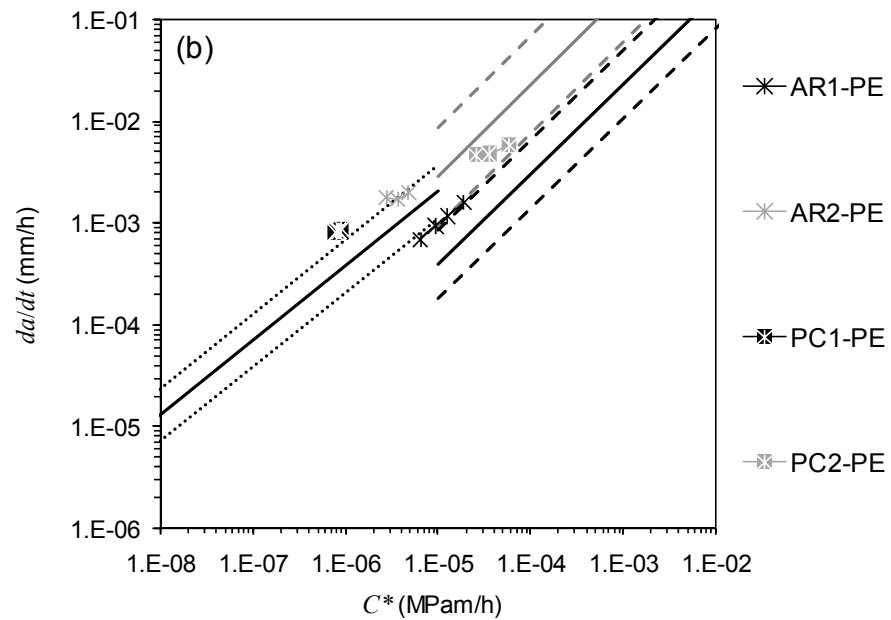
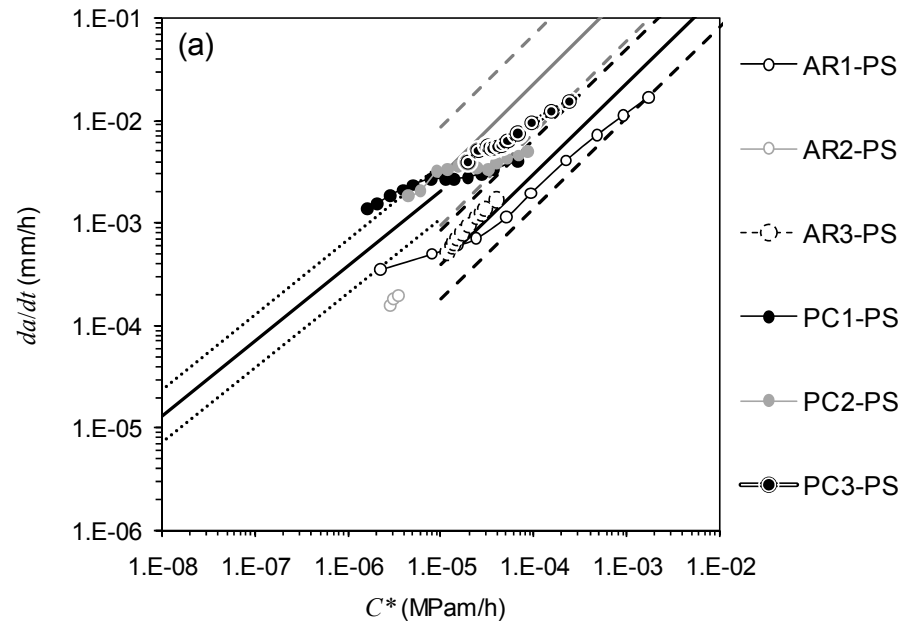


Figure 7.20: Comparison of CCG predictions for AR and PC material, made using stress dependent ductility model, at different loads under (a) plane stress (b) plane strain conditions

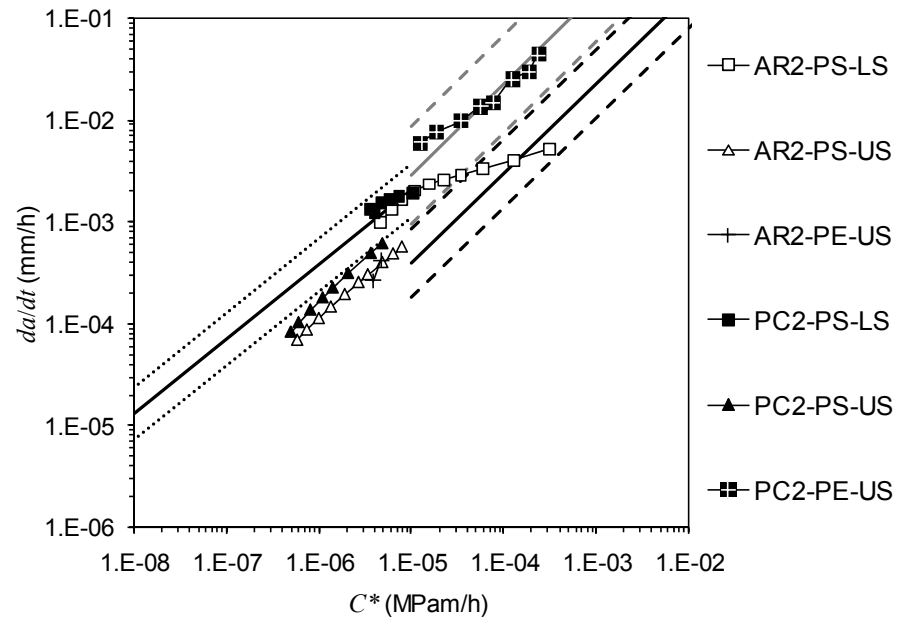


Figure 7.21: Comparison of CCG predictions for AR and PC material, made using constant ductility, at a given load ($P = 14.5$ kN)

Chapter 8

Microstructural Study of Compressive Plastic Pre-straining Effects on Intergranular Creep Damage

8.1 Introduction

As shown in Chapter 5 and Chapter 6, compressive plastic pre-strain induced in Type 316H stainless steel at room temperature, significantly influences the tensile, creep deformation and crack growth behaviour of the material. In order to examine the pre-straining effects on the mechanical and creep behaviour of 316H, optical and scanning electron microscopy (SEM) studies have been performed on AR (*i.e.* un-compressed) and 8% pre-compressed material.

The effect of tensile pre-strain on creep behaviour of stainless steels has been studied in the past mainly to provide a better understanding of the reheat cracking in weld materials. For example, the influence of tensile hot rolling on the creep behaviour of AISI 316 SS has been investigated in [14] where the creep ductility and creep strain rate were found to decrease as a result of hot tensile pre-strain. Pre-conditioning techniques were also previously employed to replicate and predict the materials behaviour in weld affected zones under service conditions (see *e.g.* [140]). In addition, the results from high temperature tests on smooth and notch bar samples and fracture specimens manufactured from pre-strained material have been reported in the literature and the observations of intergranular creep damage formation and crack growth behaviour in pre-conditioned material are detailed in [96, 141, 142].

In this chapter, the influence of material pre-conditioning by 8% plastic strain in compression at room temperature on the micro structure of 316H stainless steel has been studied and the observations are compared to AR material. The microstructures in 3 orthogonal directions relative to the pre-compression axis (*i.e.* parallel and perpendicular to the pre-compression direction) have been examined to aid the understanding of why PC material exhibits different creep failure behaviour compared to the AR material. Further microstructural studies have also been performed to

examine the fracture region of PC and AR material subsequent to failure in high temperature tensile, room temperature tensile and uniaxial creep rupture tests. The results are discussed in terms of intergranular and transgranular damage caused by the compression process and the importance of microstructural changes on the mechanical and creep behaviour of the material.

8.2 Microstructural Study

8.2.1 SEM Sample Preparation

SEM samples with 3 different orientations, with respect to the pre-compression axis, were extracted from the mid height of a pre-compressed block, which corresponds to the crack plane position in a fracture specimen or the mid length of the gauge region in uniaxial tensile and creep specimens (see Figure 8.1). Another three samples were also extracted from an uncompressed block of as-received material. Furthermore, samples have been extracted from the mid thickness of the uniaxial round bar specimens made of AR and PC materials subsequent to tensile and uniaxial creep test completion. The samples for the SEM study were ground, polished and electro-etched using 10% Perchloric Acid in Methanol electrolyte at -20 °C.

8.2.2 Pre-compression Effects in Different Orientations

The micro structures of the PC material have been examined in three orientations (along two planes parallel and one plane perpendicular to pre-compression direction), as illustrated in Figure 8.1, and compared to the corresponding grain structures in the AR material. Note that in Figure 8.1, the pre-compression direction is along the z -axis. The SEM pictures taken from the AR and PC materials along different orientations are shown and compared in Figure 8.2–Figure 8.7. The local coordinate in these figures is shown on an inset to indicate the sample orientation relative to the pre-compression direction. The pre-straining direction is indicated by the arrows shown in these figures. In each figure a selected region is identified by a bounding box and is shown in higher magnification on the right hand side of the figure. SEM backscatter pictures have been presented at two magnifications for each case— a lower magnification is used to study grain substructures and transgranular damage and a higher magnification is used to look at intergranular damage. The substructures inside the grains can also be seen in the backscatter SEM pictures in addition to the grains slip bands and directions.

By comparing the AR and PC material in the x - y plane in Figure 8.2 and Figure 8.3, respectively, it can be observed that significant intergranular damage is introduced in the x - y plane, orthogonal to the pre-straining direction, when the material is pre-

compressed to 8% plastic strain. However, there are no particular intragranular (*i.e.* transgranular) voids formed in this plane as a result of the pre-compression process. Slip bands are clearly observed in Figure 8.3 for the PC material, which are not apparent in the AR material, as expected. The same observation has also been made and reported in [143], where heat treatment and 20% plastic pre-straining was applied on 316H material. It can be seen in Figure 8.2 that the substructures are diffuse and stable in the AR material whereas pre-compression, Figure 8.3, has introduced more defined and developed substructures. It must be noted that the substructures of the PC material are potentially non-stable and may be subjected to changes by introducing dislocations.

Figure 8.4 and Figure 8.5 show the grain microstructures of the AR and PC material, respectively, in the x - z plane, along the pre-compression axis. There is no evidence of any intergranular or transgranular damage for the AR material oriented in this plane, as seen in Figure 8.4. Similar to the x - y plane, there is no transgranular damage observed but the intergranular damage in the PC material is significant, especially closer to the triple junctions around intergranular particles where local stresses are generally higher. No particular transgranular or intergranular damage has been observed in the AR and PC material oriented in the y - z plane, as shown in Figure 8.6 and Figure 8.7, respectively.

In general, the grain microstructural images of the PC material show that major intergranular damage is introduced along the grain boundaries in the x - y and x - z planes by pre-compressing the material to 8% plastic strain at room temperature. This indicates that the planes which are oriented along the width of the block, x -axis, are damaged the most on the grain boundaries where some intergranular damages may pre-exist in the AR material as a result of service exposure.

In order to measure the average grain size of the material, microscopic pictures were captured on the ground, polished and electro etched surface of the AR and PC samples. Image- J software was used to calculate the average grain size by analysing the microstructural pictures. Similar average grain diameter of approximately 50 μm was found in both AR and PC materials. The comparison of the grains before and after pre-compression signifies that the average grain size remains unaffected by pre-straining to 8% compressive plastic strain and thus similar range of grain size is observed in AR and PC materials.

8.2.3 The Influence of Pre-compression on Hardness

To investigate the influence of pre-compression on the hardness of the material oriented along different directions, with regard to the pre-straining axis, Vickers micro hardness tests were performed on the samples shown in Figure 8.3, Figure 8.5 and Figure 8.7. The hardness values found in these orientations were 209 ± 10 HV, 227 ± 11 HV and 202 ± 6 HV, respectively. All micro hardness indentation tests were carried out under 5kgf force with 20 seconds hold time at the maximum applied load. The indentations were repeated 10 times on each sample and the mean values and the uncertainties calculated as double the standard deviation (2SD) have been reported. The indentation tests on the PC material in different orientations showed that the hardness values of the PC material in all 3 planes are significantly larger than the parent material, which had the average hardness of 142 HV. Interestingly, it has been noted that the hardness value of the PC material is similar to that of reported for the heat affected zone (HAZ) material in [144] (220 HV). This may indicate similar level of hardening effects in the HAZ and 8% PC material.

8.2.4 Room and High Temperature Tensile Specimens

As explained in Chapter 5, tensile tests have been performed on the AR and PC samples oriented along the z -direction (parallel to the pre-compression direction) at both room temperature (RT) and high temperature (HT) (*i.e.* 550 °C). It was explained in Chapter 5 that hardening effects are exhibited, as a result of compressive pre-straining, in the room temperature and high temperature tensile behaviour of the material, due to the higher dislocation density in PC material. It was also seen that the failure strain in the PC material is similar to that of AR, especially at RT. The failure strain is however somewhat lower in the PC material, specifically in high temperature tensile tests (see Figure 5.1(a)).

The post-test microstructures, in the vicinity of the failure region, of the AR and PC tensile specimens at both RT and 550 °C are shown in Figure 8.8 and Figure 8.9, respectively. The loading direction in these tests (which is parallel to the pre-straining direction in the PC material) is indicated by the arrows shown. It can be seen that the fracture surface of the AR and PC materials at both temperatures are very similar, which may indicate that the failure processes (and subsequently failure strain) in both materials are therefore similar. Generally no qualitative differences can be seen between the damage and fracture region for the AR and PC tensile specimens.

8.2.5 Uniaxial Creep Rupture Specimens

It was shown in Chapter 5 that pre-compressing the material to 8% plastic strain decreases the rupture time and failure strain in uniaxial creep rupture tests, however the creep strain rate remains almost unchanged. In order to gain a better understanding of how the pre-compression process affects uniaxial creep behaviour of 316H SS, SEM pictures have been captured at the mid thickness fracture region of the AR and PC specimens tested at 257 MPa, as shown in Figure 8.10(a) and Figure 8.10(b), respectively. A significantly higher damage density specifically in the form of micro cracks is observed in the vicinity of the failure region of the PC material. These micro cracks are clearly intergranular and normal to the principal stress direction, as detailed in Figure 8.11(b) where the principal stress direction is signified by the arrow shown. No significant micro cracks are observed for the AR material in Figure 8.11(a). This indicates that the prior grain boundary damage introduced by pre-compression at room temperature can facilitate and promote the formation of micro cracks along the grain boundaries when the material is subjected to creep conditions. Therefore, the intergranular creep failure mechanism is initiated sooner in PC material, and has a smaller creep strain at failure compared to that of AR material due to the pre-existing grain boundary damage in the PC material. Similar observations have been made in [14] for pre-strained material under tensile hot rolling. The formation of premature voids on the grain boundaries, which influence creep properties of the material, is also observed and reported in [145] where the material is pre-tensioned to sufficiently large percentage of strain at room temperature and tested under constant stress creep conditions.

In general, the significance of intergranular creep damage in PC material can be linked to the fact that the number of voids on the grain boundaries is increased by room temperature pre-strain. This may be interpreted as higher level of micro stresses along the grain boundaries in PC material which can increase the void nucleation and growth rate under creep conditions. The ease of creep damage formation and development in pre-strained material therefore eventually reduces the creep strain at failure.

8.3 Discussion

It was shown that the mean hardness of the pre-compressed material in the x - z plane is the largest compared to the other two orientations. Also the average hardness value in the x - y plane is found relatively larger than that of measured in the y - z plane. This may imply that the hardening effects are more significant in the tensile repose of the PC material with the loading axis normal to the pre-compression direction

(combination of the x - z and x - y planes) compared to the specimens with the loading axis parallel to the pre-straining direction (combination of the x - z and y - z planes) as shown and discussed in [129, 130]. In other words, knowing that the tensile specimens in both orientations, with respect to the pre-compression axis, consist of the x - z plane, the uniaxial samples with the x - y plane (thus PC-Normal tensile specimens) are expected to exhibit larger hardening effects compared to the samples with the y - z plane (PC-Parallel tensile specimens).

Furthermore, it has been shown in [129] that the failure strain measured in tensile tests performed on the samples extracted along the x -direction in Figure 8.1 (*i.e.* perpendicular to the pre-compression direction) is 35% less at RT and 28% less at 550 °C than those of observed in the specimens orientated along the pre-compression direction. This may be explained by examining the microstructures shown in Figure 8.2–Figure 8.7. PC samples extracted along the x axis have their tensile axes parallel to the x - y and x - z planes (Figure 8.3 and Figure 8.5, respectively) and are therefore expected to have significantly higher initial dislocation densities and more intergranular damage present in the principal stress direction than those of extracted along the z axis, which have their tensile axes parallel to the x - z and y - z planes (Figure 8.5 and Figure 8.7, respectively).

As shown in Chapter 6, the CCG rate of the PC material is generally higher than that of AR for a given value of the crack tip fracture mechanics parameter C^* and the data follow the same trend as the upper bound long term AR. Therefore, the micro structural failure mechanisms in PC material and long-term tests on AR material may be comparable. As explained in Chapter 6 and shown in [24], a systematic change from ductile intergranular to brittle intergranular may occur in the cracking mode when the test duration is increased in CCG tests on C(T) specimens made of AR material. In brittle intergranular mode, deformation of the grains is negligible whereas the ductile intergranular mode may be characterised by crack growth in conjunction with deformation of the grains [24]. As seen in Figure 8.11(b), no significant grain deformation can be observed in the uniaxial creep tests on PC material. Knowing that the same failure mechanism is expected in creep rupture and creep crack growth, brittle intergranular cracking mode is expected to be exhibited in the PC material. This can be confirmed by Figure 6.19 presented in Chapter 6 where the cracking mode in a CCG test on PC material is shown.

8.4 Summary

- The process of pre-compression to 8% plastic strain at room temperature introduces intergranular voids and damage on the grain boundaries.
- The influence of intergranular damage induced by pre-compression becomes more significant when the material is tested under creep conditions at elevated temperatures.
- The formation of creep damage on the grain boundaries is favoured by pre-compression, which leads to the appearance of further intergranular voids on the grain boundaries.
- The average hardness values along different orientations in the PC material were found larger than that of measured in the AR material.
- The lack of significant grain deformation in uniaxial creep rupture tests on PC material suggests brittle intergranular cracking mode to occur in the CCG tests.

8.5 Figures

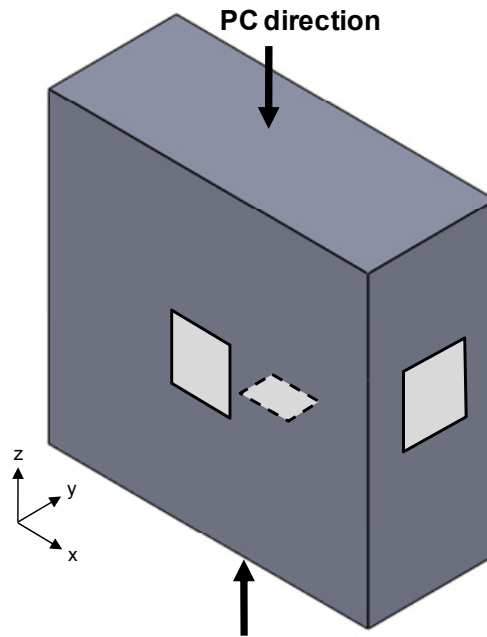


Figure 8.1: A schematic illustration of the SEM samples orientations

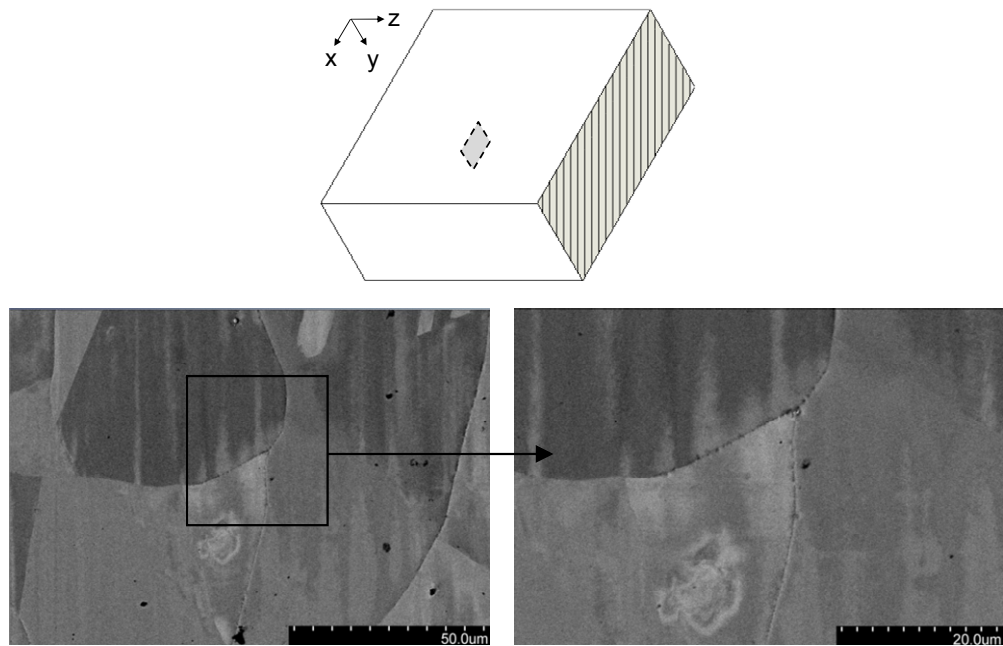


Figure 8.2: Backscatter SEM pictures of AR material, in the x - y plane normal to the axial direction, at two magnifications

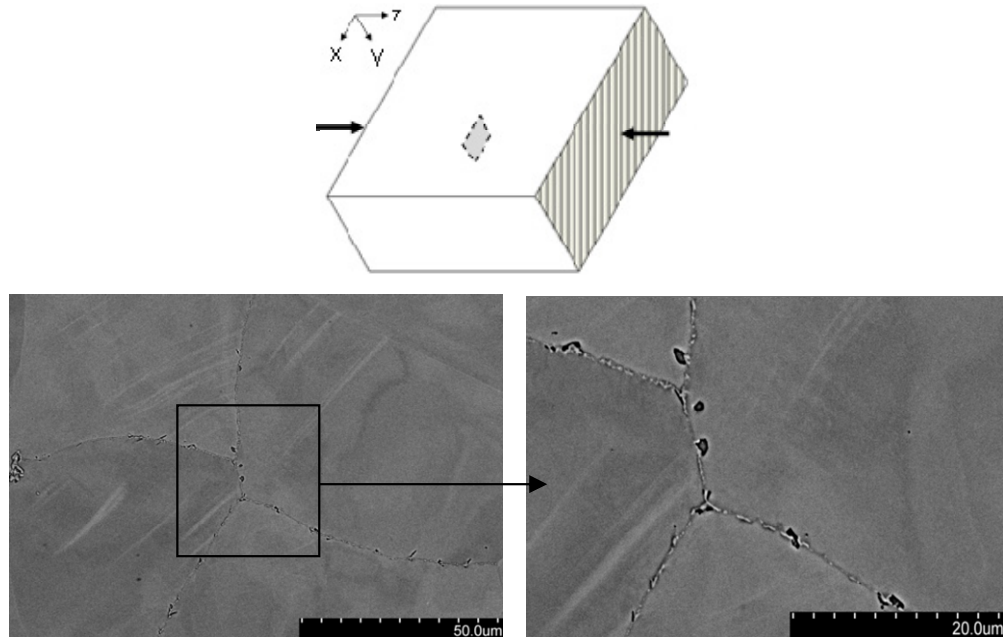


Figure 8.3: Backscatter SEM pictures of PC material, in the x - y plane normal to the axis of pre-compression, at two magnifications

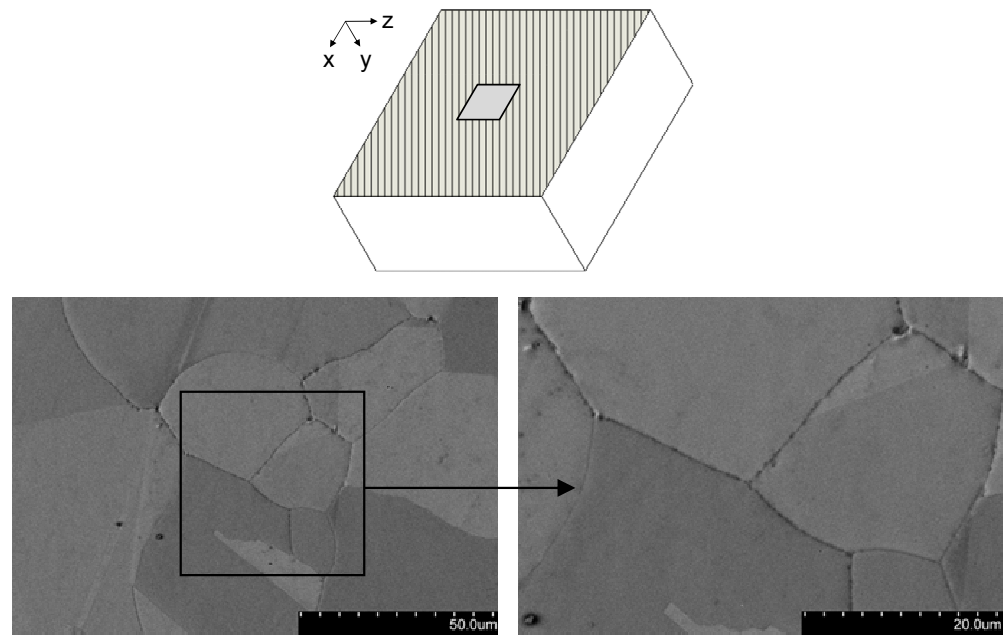


Figure 8.4: Backscatter SEM pictures of AR material, in the x - z plane parallel to the axial direction, at two magnifications

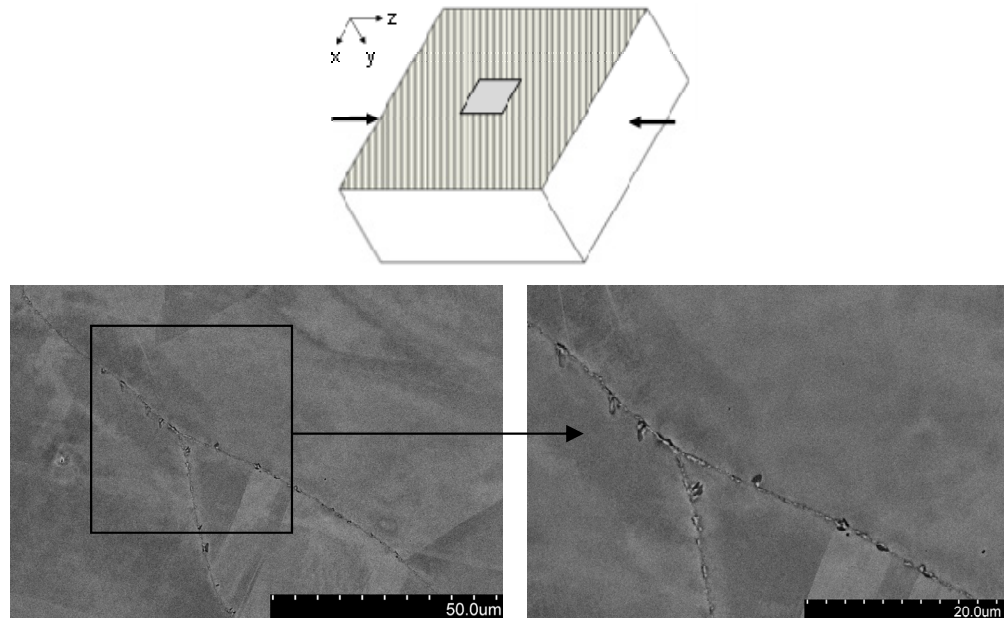


Figure 8.5: Backscatter SEM pictures of PC material, in the x - z plane parallel to the axis of pre-compression, at two magnifications

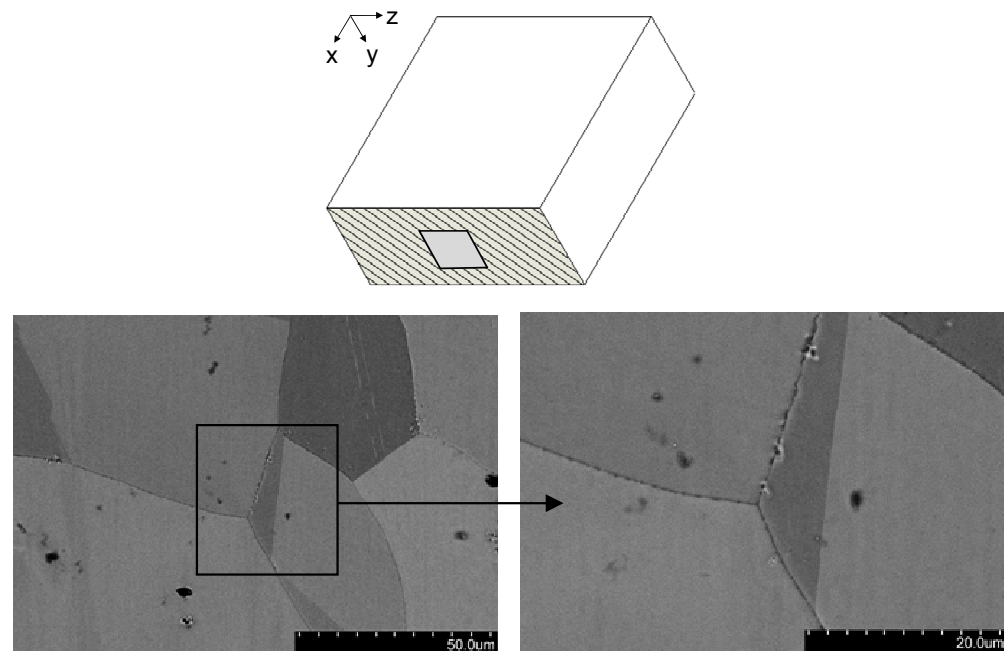


Figure 8.6: Backscatter SEM pictures of AR material, in the y - z plane parallel to the axial direction, at two magnifications

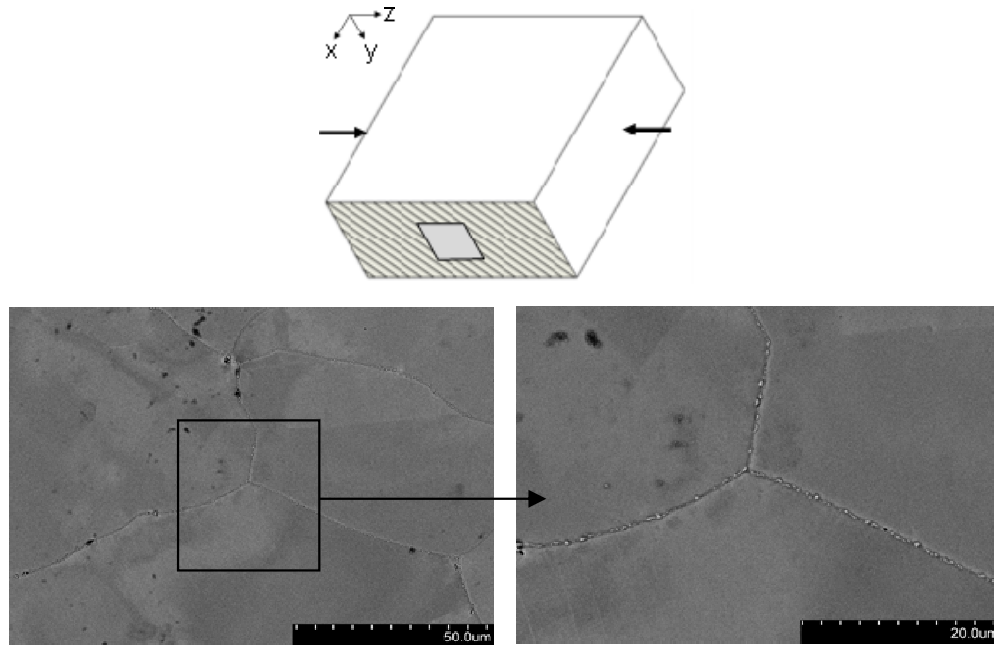


Figure 8.7: Backscatter SEM pictures of PC material, in the y - z plane parallel to the axis of pre-compression, at two magnifications

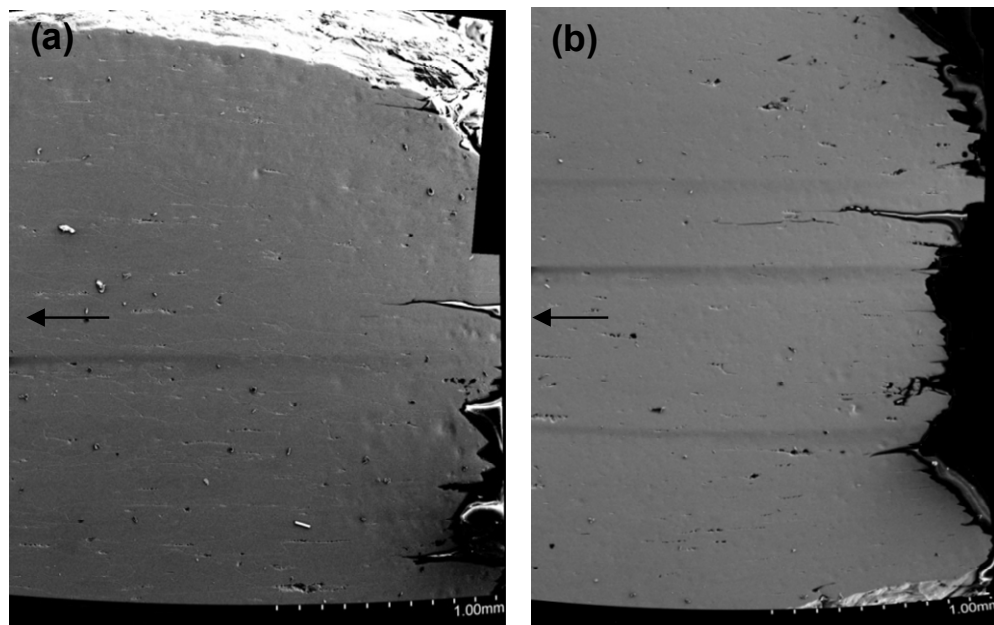


Figure 8.8: Fracture region of (a) AR (b) PC, tensile samples tested at room temperature (Secondary Electron Beam)

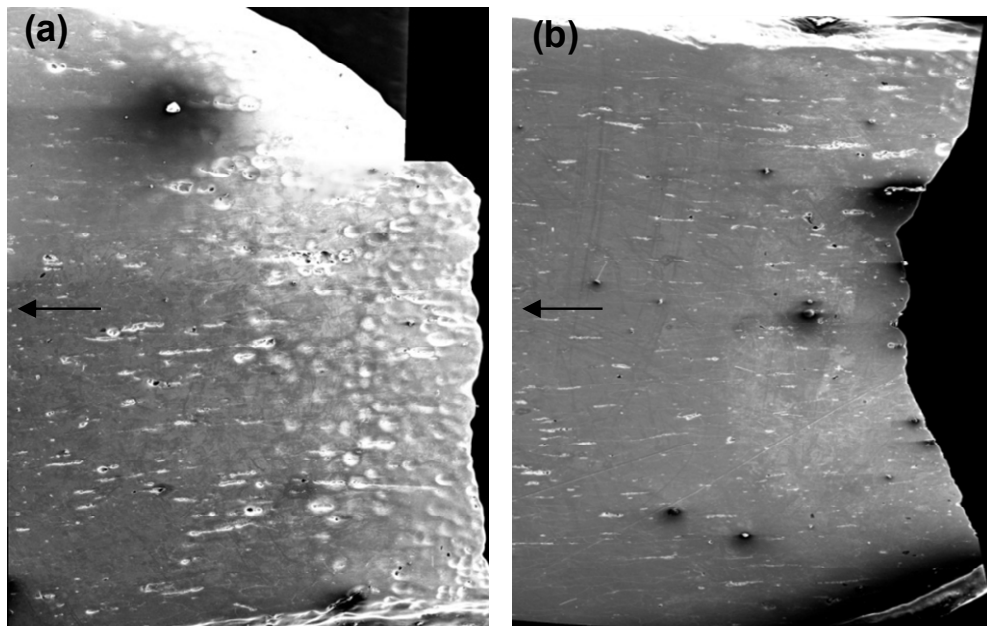


Figure 8.9: Fracture region of (a) AR (b) PC, tensile samples tested at 550 °C
(Secondary Electron Beam)

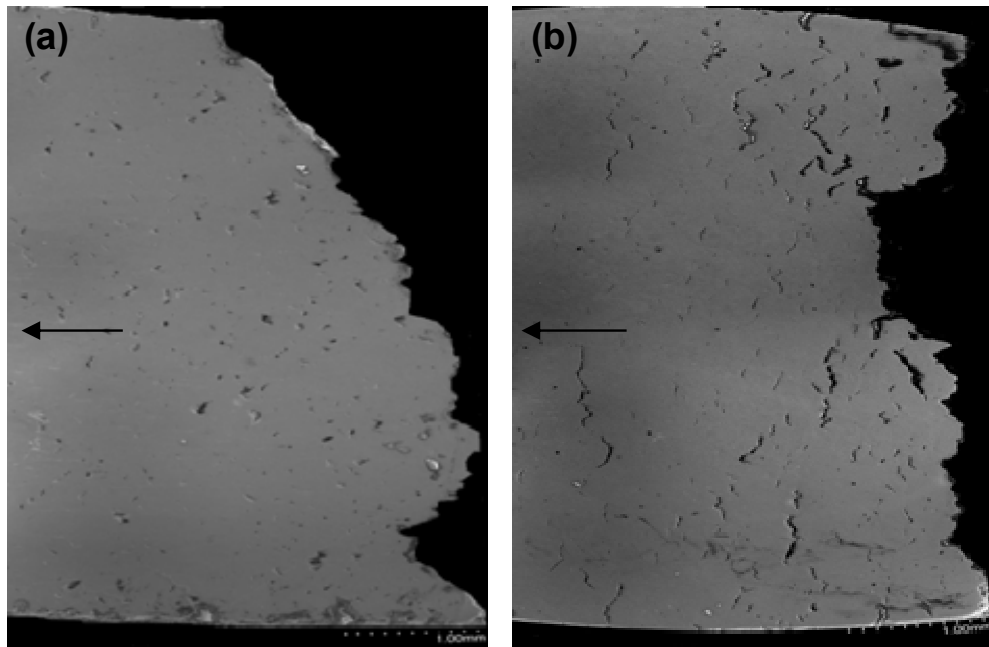


Figure 8.10: Fracture region of (a) AR (b) PC, uniaxial creep rupture samples tested
under 257 MPa (Secondary Electron Beam)

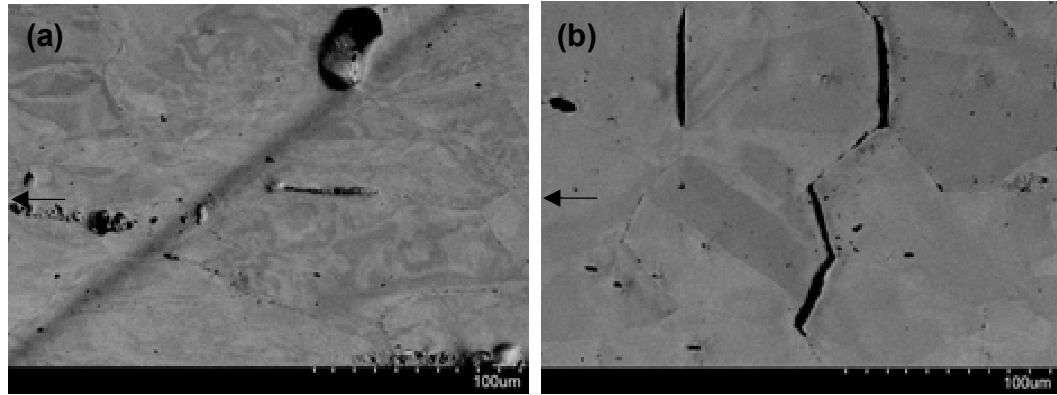


Figure 8.11: Backscatter SEM pictures of (a) AR (b) PC, uniaxial creep rupture samples tested under 257 MPa

Chapter 9

The Influence of Inelastic Damage on Fatigue Crack Growth and Fracture Toughness

9.1 Introduction

The main aim of this research is to investigate the influence of prior creep damage, which may be introduced into the material under service conditions at elevated temperatures, on subsequent fatigue and fracture toughness, which are the dominant failure mechanisms at lower temperatures. Creep damage local to the crack tip has been introduced into the material by interrupting CCG tests on compact tension, C(T), specimens at 550 °C. However to avoid any ambiguity in the data, fracture toughness and fatigue tests have been performed at room temperature.

As previously explained in Chapter 3, the yield stress of Type 316H stainless steel is relatively low at 550 °C ($\sigma_{0.2} \approx 170$ MPa). In order to study the influence of creep damage alone on fatigue and fracture toughness, blocks of 316H stainless steel were uniformly pre-compressed to 8% plastic strain at room temperature prior to specimen manufacturing. It was shown in Chapter 5 and Chapter 6 that pre-compression hardens the material, as seen in the load up response of the material in uniaxial and creep crack growth tests, and limits the extent of plasticity at the crack tip (see FE prediction of the pre-compression effects on the plastic zone size in Chapter 7). Furthermore, it was shown in Chapter 8 that additional voids may form on the grain boundaries as a result of pre-straining an ex-service block of 316H stainless steel to 8% plastic strain at room temperature. Therefore running CCG tests on PC material introduces a combination of prior plastic and subsequent creep damage, which is referred to as inelastic damage in this work, into the material.

In this chapter, the process of introducing creep damage into the material is explained in detail and fracture toughness and fatigue test results on the creep damaged material are presented. In order to differentiate the influence of pre-compression from creep damage, fracture toughness and fatigue tests have also been conducted on the AR and PC material and the results are compared to those of achieved from the creep damaged material. Further shown in this chapter are the ND

measurement results on an interrupted CCG sample to provide an estimate of the residual stresses ahead of the final crack tip in LCD specimens. In addition, the micro hardness test results along the main crack path and ahead of the final crack tip in an interrupted CCG test specimen are presented.

9.2 Local Creep Damage Tests

In order to introduce creep damage local to the crack tip, CCG tests have been performed on C(T) specimens made of PC material with $a_0/W = 0.35$ and interrupted at the approximate normalised crack length of $a/W \approx 0.5$. As discussed in Chapter 6, three CCG tests were performed on C(T) specimens with the initial normalised crack length and all other dimensions the same as the LCD specimens. The results from these tests, which are referred to as CCG calibration tests, were used to characterise the CCG behaviour of the material and thus estimating the PD calibration parameters (see Eqn (4.1)) at the normalised crack length of $a/W = 0.5$.

It was shown in Chapter 6 that fast crack growth rates, especially towards the end of the test, were observed in the CCG test on 8PC-A1 sample loaded under the initial stress intensity factor of $K(a_0) = 30.6 \text{ MPa}\sqrt{\text{m}}$. This indicates that running local creep damage tests under $K(a_0) = 30.6 \text{ MPa}\sqrt{\text{m}}$ will leave a very limited time to interrupt a LCD test at the desired crack length. Further shown in Chapter 6 was that another two CCG tests on 8PC-A2 and 8PC-A3 samples were performed at $K(a_0) = 25.5 \text{ MPa}\sqrt{\text{m}}$ and reasonable PD rates were found at the desired crack length at which the local creep damage tests are to be interrupted ($a/W = 0.5$). Thus the results from 8PC-A2 and 8PC-A3 tests were used as the CCG calibration test data and the PD rates corresponding to $a/W = 0.5$ in these two tests were used to stop LCD tests at the estimated normalised crack length of $a/W \approx 0.5$. The test duration in the CCG tests under $K(a_0) = 25.5 \text{ MPa}\sqrt{\text{m}}$ was also found reasonably short, allowing the research to be done in the allocated timescale. Furthermore, relatively small normalised reference stress values of less than unity, and thus limited plasticity, were found at the initial crack length ($a_0/W = 0.35$) and also at the normalised crack length where the LCD specimens are to be interrupted ($a/W = 0.5$), when $K(a_0) = 25.5 \text{ MPa}\sqrt{\text{m}}$ was applied. Therefore, running LCD tests at the initial stress intensity factor of $K(a_0) = 25.5 \text{ MPa}\sqrt{\text{m}}$ ensures that plasticity doesn't affect the CCG behaviour of the PC material on the initial loading and also no significant plastic damage will exist in locally creep damaged specimens at the final crack length. Hence the influence of creep damage alone, with no significant contribution of plasticity, can be studied on the subsequent fracture toughness and fatigue crack growth behaviour of the material.

As described in Chapter 3-Section 3.5.1, interrupting the LCD tests at the exact normalised crack length of $a/W = 0.5$ is challenging especially because of the accelerated CCG rates at larger crack lengths in the PC material. However, any approximate a/W of close to 0.5 will be suitable (as it would be in the valid range of crack length specified in the standard test methods) for performing fracture toughness and fatigue tests on LCD specimens. Note that although the LLD rates corresponding to the normalised crack length of $a/W = 0.5$ have been calculated for the CCG calibration tests on 8PC-A2 and 8PC-A3, but the PD rates which reflect a direct measure of the crack length have been found to provide a more accurate estimate of the final crack length. Thus the LCD tests were stopped, at the earliest convenient time, when the PD rates reached the values of those in CCG calibration tests corresponding to $a/W = 0.5$.

Four LCD tests have been performed in total. These specimens are denoted as LCD1–4. Two of these specimens, LCD1 and LCD2, have been used for fracture toughness testing whilst fatigue crack growth tests were performed on LCD3 and LCD4. The experimental results from the four LCD tests are presented next and the PD and also the LLD data from these tests are compared to those of obtained from the CCG calibration tests (8PC-A2 and 8PC-A3).

9.2.1 Experimental Results

CCG calibration test results have been used to estimate the PD rate at which the normalised crack length of around 0.5 is attained in LCD specimens. Comparison of the PD data, PD rate, LLD data and LLD rate variations against time in CCG calibration tests to those of obtained from LCD specimen are presented in Figure 9.1(a), Figure 9.1(b), Figure 9.2(a) and Figure 9.2(b), respectively. In these figures the CCG calibration tests and LCD data are shown in solid grey and hollow black symbols, respectively. Also seen in these figures are the solid and dashed grey lines for the data corresponding to the normalised crack length of $a/W = 0.5$ in 8PC-A2 and 8PC-A3 CCG calibration tests, respectively, which were evaluated post testing, by breaking open 8PC-A2 and 8PC-A3 specimens and calibrating the PD data (see Chapter 4-Section 4.1.2).

The variation of the PD data against time has been compared between the CCG calibration tests and the LCD specimens in Figure 9.1(a). As seen in this figure, the change in the PD data, ΔPD , at the time that the tests are interrupted are similar in LCD1 and LCD4 specimens. Similar values of ΔPD , which are smaller than those of seen for LCD1 and LCD4 specimens, are also observed in LCD2 and LCD3 specimens.

Further seen in this figure is that the final ΔPD values in LCD1 and LCD4 samples fall in between those of corresponding to $a/W = 0.5$ obtained from the tests on 8PC-A2 and 8PC-A3 CCG calibration specimens. It can be also observed that LCD2 and LCD3 samples have the final ΔPD values very close to that of obtained from 8PC-A3 specimen at the time corresponding to $a/W = 0.5$.

The PD rate data, dPD/dt , for the CCG calibration tests on 8PC-A2 and 8PC-A3 samples are plotted against time in Figure 9.1(b) and the results are compared to LCD specimens. As seen in this figure, different PD rates have been attained in 8PC-A2 and 8PC-A3 tests at the a/W of 0.5. This difference in the PD rates can be attributed to the inherent experimental errors which may be encountered in the tests. Note that because of the accelerating CCG in PC material at larger crack lengths, significant increase in the PD rate may occur in a relatively short period of time.

The more conservative measure of the PD rate obtained from specimen 8PC-A2 has been used to interrupt the LCD tests. As seen in Figure 9.1(b), LCD1, LCD2 and LCD4 tests have been stopped at the PD rate of around 5×10^{-4} (mV/h) which corresponds to the $a/W = 0.5$ in the CCG test on 8PC-A2 sample. However LCD3 test was stopped at a slightly lower PD rate. The crack length measurements have been performed on the LCD specimens after break open, subsequent to fracture toughness and fatigue crack growth test completion. The measured creep crack lengths in LCD specimens, a_c , are summarised in Table 9.1 and taken as the initial crack length, a_i , in subsequent fatigue crack growth and fracture toughness tests on these samples. Also seen in Table 9.1 are the values of the stress intensity factor, $K(a_c)$, and plastic zone size estimates, $r_{p,c}$, at the point that CCG tests on LCD specimens were interrupted.

A comparison has been made between the LLD, Δ , of the CCG calibration tests and the LCD specimens shown in Figure 9.2(a). As seen in this figure, the total LLD values measured from LCD1 and LCD2 are generally very close to that of obtained from 8PC-A2 specimen at a time corresponding to $a/W = 0.5$. Smaller LLD is measured in LCD3 which falls close to that of measured on the 8PC-A3 sample at a time corresponding to $a/W = 0.5$. Also seen in this figure is that the largest LLD has been obtained from the LCD4 sample.

Finally the LLD rate, $d\Delta/dt$, from the CCG calibration tests are plotted against time in Figure 9.2(b) and the data are compared to those of obtained from LCD specimens. As seen in this figure, the LLD rate at which the LCD4 is interrupted is almost the same as that of corresponding to $a/W = 0.5$ in 8PC-A2 CCG test. Also it can be seen in this figure that the LLD rate in LCD2 is close to that of seen in 8PC-A3 at a time

corresponding to $a/W = 0.5$. Further seen in Figure 9.2 is that the LLD rates in LCD1 and LCD3 at the time of interruption are below those of corresponding to $a/W = 0.5$ in 8PC-A2 and 8PC-A3 specimens, with a faster acceleration in LCD1 compared to LCD3. However, as previously shown in Figure 9.1(b) correct PD rate trends were observed in these two LCD specimens.

9.2.2 Neutron Diffraction Measurements

During a CCG test, the specimen is loaded in tension at 550°C and unloaded at the end of the test after some amount of crack extension. The process of loading (*i.e.* pre-tension), crack propagation and unloading, which is schematically shown in Figure 9.3, may lead to a significant compressive residual stress field being generated at the crack tip, which may subsequently affect the fracture toughness and fatigue crack growth test results on LCD specimens.

ND measurements have been performed on 8PC-A2 specimen to quantify the extent of residual stresses ahead of the final crack tip subsequent to CCG test completion. The results from ND measurements on this sample provide an approximate measure of the residual stresses ahead of the final creep crack tip in local creep damage specimens. Note that the dimensions, loading conditions and initial normalised crack length in 8PC-A2 sample were the same as the LCD specimens. As explained previously in Chapter 6, the loading axis was parallel to the pre-compression direction in 8PC-A2 (with 25 mm thickness) and the sample was loaded under the initial stress intensity factor of $K(a_0) = 25.5 \text{ MPa}\sqrt{\text{m}}$.

The plastic zone size estimates, $r_{p,f}$, and the normalised reference stress values calculated at the final crack length for the CCG test performed on 8PC-A2 are given in Table 9.2. Also shown in this table are the normalised final crack length and the stress intensity factor at the end of the test. As seen in this table, relatively low values of normalised reference stress ($\sigma_{ref}/\sigma_{0.2} \leq 1$) are calculated for 8PC-A2 at the final crack length and thus limited plasticity is expected in this specimen.

Neutron diffraction measurements were performed on the instrument E3 at Helmholtz-Zentrum Berlin (HZB). Measurements were taken at a number of positions in the plane of the crack, at the mid height and mid thickness of the specimen as illustrated in Figure 9.4. Note that the measurement line covered the creep damage zone which formed ahead of the final crack tip when the CCG test was interrupted. Therefore, a measure of the residual stress at the crack tip in the CCG specimen examined will provide an approximate measure of the residual stresses in LCD samples and helps to interpret the fracture toughness and fatigue test data on the LCD

specimens. The {311} peak was measured, as recommended for this steel in [111]. Note that E and ν values from Kroner model [111] have been employed in residual stress calculations for {311} peak. Due to the relatively large grain size of this material, which is on average around 50 μm as shown in Chapter 8, a gauge volume of $3 \times 3 \times 3 \text{ mm}^3$ was used. Residual strain and stress were measured along three orthogonal directions (*i.e.* principal planes) for this C(T) specimen (see Figure 9.4).

The residual strain and stress measurement results on specimen 8PC-A2 are shown in Figure 9.5(a) and Figure 9.5(b), respectively. The strain free reference value was determined by applying a force balance across the specimen. The error bars shown indicate the error associated with the peak fit as described in [110]. Note that part of the gauge volume was located behind the final crack tip position in the first measurement, and thus some additional error may be associated with the measurements at the crack tip. In Figure 9.5(b), a compressive residual stress of around -100 MPa is observed at the final crack tip position in the “yy” direction (*i.e.* normal to crack plane and parallel to the loading direction as seen in Figure 9.4) in specimen 8PC-A2. It can be seen in this figure that the compressive residual stress field along this direction is relatively small and only exists within a distance of around 1 mm from the crack tip. The compressive residual stress zone size in the “yy” direction is found in between the plastic zone size estimates made under PS and PE conditions (see Table 9.2). Further seen in Figure 9.5(b) is zero residual stress along the “zz” (through thickness) direction at the crack tip. In the in-plane “xx” direction negligible strain and stress values were measured at the crack tip, as expected. Furthermore, a tensile strain is observed at the crack tip in the “zz” direction which is around 0.3-0.5 times less than the compressive strain measured in the “yy” direction. As shown in Figure 9.5, peak tensile stress of around 200 MPa has been measured ahead of the crack tip in specimen 8PC-A2 along the “yy” direction. In addition, the peak stresses of around 100 MPa and 140 MPa are also observed in the “zz” and “xx” directions, respectively.

Specimen 8PC-A2 is nominally identical to LCD specimens and also the CCG test on this sample was stopped at a final crack length close to those in LCD samples. Hence it is expected that similar to 8PC-A2 sample, a compressive residual stress field of around -100 MPa forms at the final creep crack tip along a direction normal to the crack plane in LCD specimens. A compressive crack tip stress field in LCD specimens may inhibit early stage fracture toughness and fatigue crack growth behaviour of the material and thus need to be considered in the subsequent fracture test data analyses.

9.2.3 Micro Hardness Measurements

The hardness of a material is influenced by dislocation density hence plastic and creep strain. Therefore, hardness measurements may be used as an indicator of the extent of creep damage in the material. Indentation tests have been previously conducted by other researchers (*e.g.* [14, 146]) to specify the strained hardened zones which may be induced by the fabrication processes or service history effects on header components made of 316H SS. To examine the influence of creep damage on the hardness of the PC material, Vickers micro hardness measurements were performed on a section of the 8PC-A2 specimen post RS measurement and the results were compared to those values measured on the PC material along different orientations with respect to the pre-compression direction (see Chapter 8-Section 8.2.3). Micro hardness measurements were conducted along two lines parallel to the main creep crack path, with a distance of 0.5 mm above and below the specimen symmetry line. Indentations were carried out at the positions with 0.5 mm offset along the specified lines. These micro hardness linear profiles are illustrated in Figure 9.6. Furthermore, a micro hardness map at the final crack tip region, at a distance of 7.5 mm away from the initial EDM notch, was produced by performing sequence of indentations in an area ahead of the final crack length as seen in Figure 9.7. All micro hardness indentations were performed under 5kgf force with 20 seconds hold time under the maximum applied load. Note that the micro hardness variation scales are different in Figure 9.6 and Figure 9.7.

As seen in Figure 9.6, a small variation in the micro hardness values is generally observed in the linear profiles above and below the main crack path. The micro hardness values along these two lines are in the range of 195–210 HV. The average micro hardness value in the vicinity of the main creep crack path is generally close to those values measured on the PC material along different orientations illustrated in Figure 8.1 (see Chapter 8-Section 8.2.3). This may indicate that no significant hardening is induced as a result of the creep crack extension, in the material along the lines examined.

Slightly larger variation in the micro hardness values, compare to the linear hardness profiles, can be observed in the micro hardness map ahead of the crack tip shown in Figure 9.7. The micro hardness values ahead of the final crack tip are in the range of 185–215 HV with the lowest value found in the creep damage zone which comprises micro cracks on the grain boundaries. This may indicate that the creep damage reduces the hardness of the material close to the crack tip, but values similar to those of measured on the PC material are observed away from the creep damage

zone. These results show that micro hardness test can be employed as a valuable tool to provide a measure of the extent of strain hardening induced by plastic/creep deformation and also indicating the formation of creep damage which often appears in the form of intergranular micro cracks.

9.3 Fracture Toughness Tests

Fracture toughness (JIC) tests have been performed on two AR (JIC-AR1, JIC-AR2), two PC (JIC-PC1, JIC-PC2) and two LCD (LCD1, LCD2) specimens at room temperature, following the ESIS P2-92 standard. The loading/unloading rate was 1.3 mm/min in all tests. The AR and PC specimens were pre-fatigue cracked prior to fracture toughness testing to introduce sufficiently sharp crack tip into the samples. After pre-cracking, the AR and PC samples were side grooved to 20% of the total thickness. Relatively sharp crack tips are expected in LCD specimens since they were subjected to creep crack growth prior to fracture toughness testing. Though as shown in Chapter 6-Section 6.3.7, some crack discontinuities may be expected around the final crack tip in these samples.

Single specimen approach was employed to quantify the fracture toughness of the materials with different extent of damage. The general test method for performing fracture toughness tests was described in Chapter 4-Section 4.3. However, due to the specific testing methodologies employed in this work, additional data analysis procedures are outlined below.

9.3.1 Compliance Correction Calculations

The LLD data were measured using an external extensometry system in all fracture toughness tests examined. To exclude the extraneous compliance inherent in the testing configuration, the elastic unloading compliances associated with the specimen's LLD, $C_e^{Specimen}$, have been corrected using

$$C_e^{Specimen} = C_e^{Total} - C_e^{Extraneous} = C_e^{Total} - \left(\frac{1}{C_1} + \frac{1}{PC_2} \right) \quad (9.1)$$

where C_e^{Total} is the total elastic unloading compliance calculated using the raw LLD data, P is the applied load and C_1 and C_2 are the correction factors calculated by the testing software. The values of the correction factors for each test are given in Table 9.3. The details of the compliance correction calculations using C_1 and C_2 parameters are given in [147].

9.3.2 Fracture Resistance Curves

The dimensions of all the fracture toughness specimens are summarised in Table 9.4. Further given in this table are the average initial, a_i , and final, a_f , crack lengths in all specimens examined. These measurements have been made post testing by heat tinting (to mark the final crack length) and then breaking open the specimens. The average initial and final crack lengths have been measured using digital imaging approach and area calculations [24, 108]. As seen in Table 9.4, the average initial crack lengths normalised by the samples' width are in the range of 0.49–0.55 for the specimens tested. Also the normalised final crack lengths are in the range of 0.55–0.64. This means that all the fracture toughness samples were tested within the valid range of $0.45 \leq a/W \leq 0.65$. Also seen in Table 9.4 is that the AR and PC specimens were side grooved to 20% of the specimen thickness according to the ESIS P2-92 standard. However due to the maximum load carrying capacity of the creep machines, LCD specimens were side grooved to a larger percentage (30% in total). This small difference between the net thickness of LCD specimens compared to AR and PC samples is not expected to affect the fracture toughness results since specimen size independency checks have been performed in the analysis of all the fracture toughness test data.

The load *vs.* LLD data recorded during the fracture toughness tests are shown in Figure 9.8 which include the loading/unloading data from which the elastic compliances are calculated to provide an approximate measure of the instantaneous crack length. Similar responses can be seen in the linear elastic part of the load *vs.* LLD trends for all the specimens examined. Figure 9.8 shows that for a given load, larger LLD is observed in the AR specimens compared to those of obtained from the PC and LCD samples. This is due to the lower yield stress in the AR material. Also seen in this figure is that due to the hardening effects, a larger maximum load, P_{max} , is observed in PC specimens during the fracture toughness tests compared to those of measured for AR specimens. However due to the creep damage effects in LCD samples, the yield stress and P_{max} measured in these specimens fall below those of seen in PC samples. The lower yield stress and P_{max} observed in LCD specimens containing micro cracks ahead of the crack tip are consistent with the drop in the micro hardness measurement results in the presence of intergranular micro cracks as shown in Figure 9.7.

The fracture toughness data were analysed following the ESIS P2-92 standard testing and analysis method as detailed in Chapter 4-Section 4.3. The fracture resistance curves from the tests performed on the AR, PC and LCD specimens are

presented in Figure 9.9. Note that the crack extensions presented in this figure were approximated by employing the unloading compliances observed in Figure 9.8 and correcting them using Eqn (9.1). In Figure 9.9, the AR specimens are shown in solid symbols whereas the PC and LCD samples are shown in hollow symbols. The comparison of the R-curves for the AR and PC specimens in Figure 9.9 shows that the fracture energy corresponding to a given amount of crack extension, Δa , is larger in the AR material compared to the PC. This is due to the lower tensile strain at failure in the PC material compared to AR as shown and discussed in Chapter 5-Section 5.2.1

Comparing the J vs. Δa data for the PC material to those of obtained from the LCD samples in Figure 9.9 shows that although the R-curve trends in the LCD samples at larger crack extensions are similar to those observed for the PC material, but some noticeable crack growth is estimated in these specimens at relatively small values of the fracture energy J . This part of the R-curve observed in LCD specimens is referred to as a “tail” region. Note that in LCD specimens, the existence of micro cracks and unbroken ligaments ahead of the creep crack tip (and thus crack discontinuity) influences the unloading compliance measurements. Due to inaccurate unloading compliance values measured in the creep damage zone, artificial tail regions are appeared in the obtained R-curves for LCD samples.

To study the fracture resistance behaviour of the LCD material in the region outside the damage zone, the R-curves for the LCD specimens shown in Figure 9.9 are shifted back and the modified J vs. Δa trends are compared to the AR and PC R-curves in Figure 9.10. This figure shows that by excluding the tail regions (which are approximated by unloading compliance measurements to be around 3 mm) from the R-curves for LCD specimens, similar trends are observed between the LCD and PC specimens. The J vs. Δa plots for LCD specimens modified by removing the tail regions are referred to as “modified R-curves”. It can be seen in Figure 9.10 that although the R-curves are almost coincident for PC and LCD material in the earlier half of the total crack extension, but the J values, corresponding to a given crack extension in LCD specimens, fall slightly below PC material in the latter half of the total crack extension region. This can be attributed to the creep deformation and damage effects in LCD specimens which may lead to a drop in the failure strain of the material.

The fracture toughness data analysis on the AR, PC and modified LCD R-curves are presented next and the fracture parameters are quantified for each test. The equations of the line of best fits to the valid data points and the fracture toughness results for all tests are summarised in Table 9.5 and Table 9.6, respectively. To

differentiate the influence of pre-compression on the fracture behaviour of the material from creep damage effects, a comparison has been made between the AR and PC materials first and then the PC fracture toughness test data are compared to those of obtained from the modified LCD R-curves.

9.3.3 Pre-compression Effects on Fracture Toughness

The fracture toughness test data for the AR and PC specimens are analysed and the results are shown in Figure 9.11—Figure 9.14. The maximum valid crack extension, Δa_{max} , and the slope of the blunting line have been calculated using Eqn (4.18) and Eqn (4.19), respectively, for each test and the results are tabulated in Table 9.6. Note that the tensile data and the specimen dimensions required for analysing the fracture toughness test data from the AR and PC specimens have been taken from Table 5.1 and Table 9.4, respectively.

As seen in Figure 9.11(a), the blunting line has been constructed for JIC-AR1 specimen and the exclusion lines parallel to the blunting line with the offsets of 0.1 mm and Δa_{max} have been plotted on the J vs. Δa graph. The data points falling in between these two exclusion lines are considered valid and shown in black symbols. However, those data behind the 0.1 mm or in front of the Δa_{max} exclusion lines are considered invalid and shaded in grey. After applying the fracture resistance validity criteria, the line of best fit has been made to the valid data points as seen Figure 9.11(a). The equation of regression line for the valid data points is given in Table 9.5.

After verifying the valid data points, the maximum allowable J value, J_{max} , and subsequently the Ω parameter have been calculated using Eqn (4.25) and Eqn (4.26), respectively, and the results are summarised in Table 9.6. As seen in Table 9.6, the value of Ω parameter for the fracture toughness test on JIC-AR1 specimen is larger than 10, thus the valid data points which fall below the J_{max} exclusion line may be considered a material property, independent of the specimen size.

An offset line parallel and 0.2 mm away from the blunting line has been constructed as shown in Figure 9.11(b). The intersection between this offset line and the line of best fit to the valid data points is considered as the initiation resistance value of J , $J_{0.2/BL}$, as shown by a circle in Figure 9.11(b). It can be seen in this figure that the $J_{0.2/BL}$ value is smaller than the calculated J_{max} for this test. Also shown in Table 9.6 is that the slope of the line of best fit at the intersection point is smaller than half of the slope of the blunting line. Thus the obtained $J_{0.2/BL}$ value is considered valid and the

corresponding value of plane strain stress intensity, $K_{0.2/BL}$, is calculated for this test and given in Table 9.6.

Furthermore, the value of J at the intersection point between the constant crack growth line of 0.2 mm and the best fit curve to the valid fracture resistance data points, $J_{0.2}$, has been found and tabulated in Table 9.6. The obtained $J_{0.2}$ value for JIC-AR1 test is considered valid since it's smaller than the calculated J_{max} for this test and there is at least one valid data point between 0.2 mm and 0.4 mm constant crack extension lines.

Similar to JIC-AR1, the fracture toughness test data from the JIC-AR2, JIC-PC1 and JIC-PC2 specimens have been analysed as shown in Figure 9.12, Figure 9.13 and Figure 9.14, respectively. The equations of the line of best fits to the valid data points in each test are given in Table 9.5. Furthermore, the slope of the blunting lines and the values of Δa_{max} , J_{max} , $J_{0.2/BL}$, $K_{0.2/BL}$, Ω , $2dJ/d\Delta a_{(0.2/BL)}$ and $J_{0.2}$ for each test are summarised in Table 9.6. As seen in this table, the slopes of the blunting lines have been found very similar in the AR and PC materials. Though slightly larger values of J_{max} are generally observed in the specimens made of AR material compared to the PC samples. The values of Ω calculated for the tests on AR and PC specimens have been found larger than 10. Thus the valid data points which are below J_{max} in each of these tests can be considered independent of the specimen size.

Further seen in Table 9.6 is that the $2dJ/d\Delta a_{(0.2/BL)}$ values (double of the slope of the best fit curve at the intersection point with the 0.2 mm offset line parallel to the blunting line) are smaller than the calculated slopes of the blunting lines and also $J_{0.2/BL}$ values fall below J_{max} in each of these tests examined on the AR and PC materials. This means that the $J_{0.2/BL}$ values found in these tests are valid. Finally observed in Table 9.6 is that all the $J_{0.2}$ values found for the tests on AR and PC materials can be considered valid since they are below the corresponding J_{max} for each test. Comparison of the values of fracture toughness parameters obtained from the tests performed on the AR and PC materials show that $J_{0.2/BL}$ (and subsequently $K_{0.2/BL}$) and $J_{0.2}$ values are larger in the AR material compared to the PC. This means that greater amount of fracture energy is needed to initiate the crack in the AR material due to the larger failure strain compared to the PC material.

9.3.4 Creep Damage Effects on Fracture Toughness

The fracture toughness data analysis has been performed on the modified R-curves obtained from the tests on LCD1 and LCD2 specimens as illustrated in Figure 9.15 and

Figure 9.16, respectively. The lines of best fits to the valid data points, after applying the exclusion of the data behind 0.1 mm and beyond Δa_{max} lines, have been shown in Figure 9.15(a) for LCD1 and in Figure 9.16(a) for LCD2 specimen and the equations are given in Table 9.5. Also the J_{max} exclusion lines for these tests, the 0.2 mm offset lines with the slope of the blunting line and the $J_{0.2/BL}$ values are presented in Figure 9.15(b) for LCD1 and Figure 9.16(b) for LCD2 specimen. The slope of the blunting lines and the values of Δa_{max} , J_{max} , $J_{0.2/BL}$, $K_{0.2/BL}$, Ω , $2dJ/d\Delta a_{(0.2/BL)}$ and $J_{0.2}$ for these two tests on the LCD samples are summarised in Table 9.6. It can be seen in this table that the Ω values obtained from these two tests are larger than 10, thus the valid data points which fall below J_{max} can be considered independent of the specimens size. Also seen in this table is that $2dJ/d\Delta a_{(0.2/BL)}$ is smaller than the slope of the blunting line and $J_{0.2/BL}$ values are smaller than the corresponding J_{max} in both tests on LCD1 and LCD2 specimens. Thus the obtained values of $J_{0.2/BL}$ for these two tests are considered valid and consequently plane strain stress intensity factors, $K_{0.2/BL}$, are calculated. Furthermore, $J_{0.2}$ values found in these two tests are smaller than J_{max} and at least one data point falls in between 0.2 and 0.4 mm constant crack extension lines as can be seen in Figure 9.15(b) and Figure 9.16(b) for LCD1 and LCD2 specimens, respectively. Therefore, $J_{0.2}$ values obtained from these two tests can also be considered valid.

The fracture toughness test results from the LCD samples are compared to those of obtained from the PC specimens. As seen in Table 9.6, the slope of the blunting lines are found the same in the PC and LCD specimens since the interrupted CCG test specimens were made of the pre-compressed material and thus have the same properties. It can be observed in this table that the Δa_{max} values calculated for the PC specimens are larger than those for the LCD samples. Also similar values of J_{max} have been found in LCD and PC materials, though the values in PC specimens are slightly larger.

Further seen in Table 9.6 is that generally larger values of $J_{0.2/BL}$ (and subsequently $K_{0.2/BL}$) and $J_{0.2}$ are found in the PC specimens compared to LCD samples. This implies that when the modified R-curve (which describes the fracture behaviour of the LCD material outside the creep damage zone) is employed in the fracture toughness data analysis on the LCD specimens, the fracture energy may be considered lower in the material with creep damage compared to the PC material with no creep damage. The decrease in the fracture energy of the LCD material is not only observed at small values of Δa (*i.e.* 0.2 mm), but as previously shown in Figure 9.10, considerably lower amount of fracture energy is required for the LCD specimens also at larger crack extensions (in regions away from the creep damage zone). This indicates that due to

the smaller failure strain in LCD specimens compared to the PC material, a lower fracture energy is needed for the crack to initiate and propagate in these samples.

An alternative approach may also be applied to analyse the fracture toughness data obtained from LCD specimens assuming that the tail regions in J vs. Δa trends are real. In this approach, the inaccuracy in the instantaneous crack lengths estimated from the unloading compliance measurements may be considered to be distributed within the total crack extension and not only occurring at the initial crack length located in the creep damage zone. By applying correction factors, obtained using the values measured from the fracture surface, on the initial and final crack length estimates the fracture resistance curves for LCD specimens can be re-generated as shown in Figure 9.17. Using the corrected R-curves shown in Figure 9.17 for LCD samples, the new fracture toughness values are found as $J_{0.2/BL} = 0.04$ and 0.03 MPam for LCD1 and LCD2 specimens, respectively.

9.3.5 Fractography on Fracture Toughness Specimens

Macroscopic pictures taken from the fracture surfaces of the AR, PC and LCD fracture toughness specimens have been shown in Figure 9.18, Figure 9.19 and Figure 9.20, respectively. It can be seen in Figure 9.18 and Figure 9.19 that three regions can be observed on the fracture surface of the AR and PC specimens. The first region is the pre-fatigue crack zone, followed by crack growth region by fracture toughness and finally the fast fracture (due to break open) zone. Figure 9.20 shows that on the fracture surface of the LCD specimens four regions can be observed which are the EDM notch, creep crack growth region, crack extension zone due to fracture toughness and finally the fast fracture (due to break open) region.

Figure 9.18(b) and Figure 9.19(b) show that pre-fatigue cracking in specimens JIC-AR2 and JIC-PC2 was relatively non-uniform. This can be due to misalignment in the test set up. The discrepancy between the R-curves obtained from the JIC-AR1 and JIC-AR2 samples, as seen in Figure 9.9, can be due to non-uniform pre-cracking in sample JIC-AR2. Therefore the results from JIC-AR1 are expected to be more reliable. However, the R-curve obtained from the test on JIC-AR2 is found more continuous. Although non-uniform pre-crack is observed in JIC-PC2 sample, but the J vs. Δa trends observed for both PC specimens in Figure 9.9 are almost coincident suggesting that non-uniform pre-cracking may have more significant consequences on the AR material compared to the PC.

A detailed micro structural study of the cracking mode has been performed by capturing SEM pictures on the fracture surface of the LCD2 specimen. These images are shown in Figure 9.21. As seen in this figure, the cracking mode in the CCG region is intergranular whereas transgranular deformation can be observed in the fracture toughness area. Further observed in the SEM pictures of the fracture toughness zone in Figure 9.21 are the transgranular voids which have plastically grown as a result of fracture toughness testing. In Figure 9.21 the SEM pictures taken along a line corresponding to the average creep crack length measured on the fracture surface show that intergranular deformation is dominant in this region, though some transgranular void growth can also be observed along the examined line. This indicates that an interaction between both failure mechanisms may be expected at the end of creep/beginning of fracture toughness where the unbroken ligaments introduced by CCG are linked up under fracture toughness testing conditions. Also shown in Figure 9.21 are the SEM pictures captured along a line located in the fracture toughness region with an offset of around 1 mm from the average creep crack length. As seen in Figure 9.21, the dominant failure mechanism in this examined region is transgranular crack growth and the plastic void growth is apparent in all the pictures taken along this line. No evidence of intergranular creep damage has been found in this region.

9.4 Fatigue Crack Growth Tests

High frequency fatigue crack growth tests were performed on three AR (FCG-AR1, FCG-AR2 and FCG-AR3), three PC (FCG-PC1, FCG-PC2 and FCG-PC3) and a LCD (LCD3) specimen at room temperature following the ASTM E647 [107] standard. All FCG tests were conducted at an R -ratio (see Eqn (2.126)) of $R = 0.1$ and frequency of $f = 10$ Hz. Based on the existing FCG data in the literature, an initial stress intensity factor range of $\Delta K = 18.5 \text{ MPa}\sqrt{\text{m}}$ was chosen for the tests on AR and PC specimens. The FCG test on LCD3 specimen was also started at the same initial ΔK . However no crack initiation was observed at this initial ΔK and thus the stress intensity factor range was incrementally increased to $\Delta K = 23.5 \text{ MPa}\sqrt{\text{m}}$ at which the crack started to initiate and grow in LCD3 specimen. Although the initial ΔK was different in the FCG test on LCD3 specimen compared to AR and PC samples, the $d a/d N$ vs. ΔK trends are still comparable due to the same R -ratio and frequency applied in all tests. The instantaneous crack length in all tests was estimated using the unloading compliance measurements, similar to the methodology employed in fracture toughness tests. In order to introduce sufficiently sharp crack tip into the FCG specimens, the samples made of AR and PC materials were pre-fatigue cracked prior to FCG testing. The initial crack length in these specimens was around 22 mm and they were pre-cracked to

approximately 25 mm prior to FCG testing. The AR and PC specimens were heat tinted to mark the starter crack tip for fracture surface measurements subsequent to test completion.

The dimensions of all FCG specimens are summarised in Table 9.7. As seen in this table, the dimensions of the AR and PC specimens were chosen the same as those of LCD specimens for comparison purposes. Note that similar to LCD specimens, the AR and PC samples were side grooves by 15% each side prior to FCG testing. The specimens were broken open subsequent to test completion and the average crack lengths before and after FCG testing were measured at the fracture surface using digital imaging technique. These measurements are given in Table 9.7 and compared to those of estimated from the unloading compliance at the beginning and end of the fatigue crack growth tests. As seen in the table, acceptable estimates were provided by the unloading compliance for the AR and PC materials especially for larger crack lengths (at the end of the FCG tests). However the percentage error shown in Table 9.7 implies that some uncertainties were encountered in crack length estimates using unloading compliance measurements at the beginning and end of the FCG test on LCD3 specimen. As seen in this table, the largest error is observed when the initial crack length in the FCG test on the LCD3 specimen is estimated using the unloading compliance. This can be attributed to the presence of micro cracks and unbroken ligaments in the creep damage zone of LCD specimens which are expected to have a significant influence on the unloading compliance measurements. Note that the initial normalised crack length was around $a/W \approx 0.5$ in all FCG tests, as shown in Table 9.7. As seen in this table, the FCG test on LCD4 will be performed as a future work and therefore the crack length measurements and estimates are not available yet.

The variation of the crack length, a , (estimated from the unloading compliance) is plotted against the number of cycles, N , for the FCG tests performed on the AR and PC materials in Figure 9.22. The data from the AR and PC materials in this figure are shown in grey and black lines, respectively. Comparison of the fatigue crack growth trends for the specimens made of AR and PC materials (which all have the same dimensions and tested under identical cyclic loading conditions) shows that for a given N , larger crack extensions are observed in the AR material compared to the PC for $N > 4.0 \times 10^4$. This can be attributed to the hardening effects and larger yield stress in the PC material which may limit the extent of plastic/fatigue damage at low load levels in cyclic tests. Also seen in Figure 9.22 is that accelerated fatigue crack growth is exhibited in all tests performed on the AR and PC materials.

The crack length a vs. N trends for the AR and PC materials are compared to that of obtained from the LCD3 specimen in Figure 9.23. Note that the N axis in this figure is plotted in log scale to provide a clear comparison between AR, PC and LCD specimens. As seen in Figure 9.23, the fatigue crack growth in AR and PC specimens occurs immediately and the initiation time can be assumed negligible in these tests. This is due to pre-fatigue cracking prior to performing FCG tests and thus introducing infinitely sharp crack tip into the samples. However, due to the change in cracking mode (from intergranular to transgranular) and also the presence of micro cracks and unbroken ligaments (and thus discontinuous crack path) at the crack tip in LCD3 specimen, a substantial delay in crack initiation and early stage fatigue crack extension is observed. Subsequent to fatigue crack initiation in LCD3 specimen, accelerated cracking can be seen in this specimen at large number of cycles. The possible factors which may influence the crack initiation in LCD specimens are described in more detail in discussion section.

The variation of maximum stress intensity factor in a cycle, K_{max} , against the crack length is illustrated in Figure 9.24 for the FCG tests performed on AR, PC and LCD specimens. As seen in this figure, for a given value of crack length, considerably larger K_{max} is applied in LCD3 specimen compared to those of observed in AR and PC materials. As explained previously, due to the difficulties encountered to initiate the crack in LCD3 specimen, a larger initial ΔK was applied in this test compared to those of performed on the AR and PC materials. This led to a higher K_{max} for a given crack length in LCD3 specimen compared to other tests, though due to the same R -ratio applied in all tests the fatigue crack growth trends (da/dN vs. ΔK) are still comparable. Further seen in Figure 9.24 is that the K_{max} vs. a trends for the AR and PC specimens are almost coincident. This is because of the similar initial ΔK and the same R -ratio applied on these specimens which results in similar K_{max} trends for these samples.

Finally seen in Figure 9.24 is that the initial K_{max} value applied on LCD3 specimen ($K_{max}(a) \approx 26.0 \text{ MPa}\sqrt{\text{m}}$) was severely lower than that of previously applied on this sample at the end of the CCG test ($K(a_c) = 37.0 \text{ MPa}\sqrt{\text{m}}$ as seen in Table 9.1). Therefore, prior creep damage is expected to significantly influence the early stage fatigue crack growth behaviour of the material especially up to the point that $K_{max} < 37.0$. Note that in order to investigate LCD effects on the subsequent FCG behaviour of the material, it is important to apply sufficiently low loads during cyclic tests to avoid introducing additional inelastic damage into the material.

9.4.1 Pre-compression Effects on Fatigue Crack Growth

The fatigue crack growth rate per cycle, da/dN , has been correlated with the stress intensity factor range, ΔK , in Figure 9.25(a) and Figure 9.25(b) for the tests performed on the AR (FCG-AR1–3) and PC (FCG-PC1–3) specimens, respectively. The data in Figure 9.25 are plotted in log-log axes and the da/dN data presented have the unit of mm/Cycle. As seen in Figure 9.25, the FCG data for the AR and PC specimens appear as a straight line with almost coincident trends when presented in log-log axes.

The lines of best fits have been made to each data set within the linear steady state Paris law region. These trend lines are shown in Figure 9.26 (AR and PC materials shown in grey and black lines, respectively) and the fatigue crack growth power law constants (see Eqn (2.130)) obtained from these regression fits are given in Table 9.8. Also included in this table are the R^2 values indicating the accuracy of each mean fit to the data points. As seen in Table 9.8 all R^2 values are close to unity and thus acceptable fits have been made to the data. Table 9.8 shows that the FCG power law coefficient, C , and exponent, m , are almost the same in both AR and PC materials ($C = 0.4 \pm 0.1 \times 10^{-8}$ and $m = 3.2 \pm 0.1$). This means that the pre-compression process has insignificant effects on the fatigue crack growth behaviour of the material at room temperature.

9.4.2 Creep Damage Effects on Fatigue Crack Growth

The FCG data from LCD3 specimen are compared to the mean fits to the AR and PC data in Figure 9.26. As seen in this figure, significantly lower fatigue crack growth rate per cycle of around two orders of magnitude smaller than those of received from the AR and PC materials is obtained in the local creep damage specimen at the early stage FCG. However, as the crack grows the FCG trend for LCD3 specimen tends to the line of best fit to the AR and PC materials at large values of ΔK . In other words, a convergence between the FCG trends of the LCD3 specimen and those of obtained from the AR and PC materials can be observed away from the initial creep damage zone. This may imply that the creep damage density is insignificant outside the initial crack tip zone and thus the FCG behaviour of the LCD specimen is similar to the PC material with no creep damage.

Also shown in Figure 9.26 are the increments of crack extension estimated using the unloading compliance measurements. As previously shown in Table 9.7, an approximate error of around 9% was found in the initial crack length estimates using the unloading compliance approach in LCD3 specimen, though acceptable estimates were made at

the final crack length. Comparison of the estimated and measured fatigue crack extension for LCD3 specimen in Table 9.7 shows that the unloading compliance approach over estimates the total fatigue crack extension by around 3 mm due to the existence of micro cracks/unbroken ligaments around the creep crack tip in LCD specimen. Therefore the early stage FCG trend corresponding to the first 3 mm of crack extension estimated by the unloading compliance technique in Figure 9.26 may be considered unrealistic for LCD3 specimen. However, the rest of the FCG curve for the LCD3 sample, which is found closer to the reference FCG trends from the PC material, can be assumed valid. This implies that although lower initial crack propagation rates of around two orders of magnitude are observed in LCD3 specimen compared to the PC material, but in reality the drop in the da/dN due to the creep damage effects ahead of the crack tip is only around an order of magnitude (see Figure 9.26).

The FCG trend for LCD3 specimen in Figure 9.26 suggests that local creep damage severely influences the initiation and early stage fatigue crack growth behaviour of the material. It can be seen in Figure 9.26 that as a result of creep damage formation (in the form of micro cracks and unbroken ligaments) at the crack tip zone, the threshold region of the da/dN vs. ΔK FCG trend (which was schematically shown in Figure 2.12 and is referred to the region that fatigue damage builds up ahead of the crack tip) is signified for the case of LCD3 specimen. Note that sufficiently sharp transgranular crack tip was introduced into the AR and PC specimens by pre-fatigue cracking them before the FCG testing. Therefore, the FCG behaviour of the AR and PC materials can be described by the fracture mechanics K parameter. Knowing that micro cracks and unbroken ligaments are present at the crack tip in LCD specimens, K parameter may still be applied to characterise the initiation and early stage FCG behaviour of the LCD material assuming a sharp, though discontinuous, crack tip. As the crack grows and the value of ΔK enlarges (*i.e.* $K_{max} \gg K(a_c)$), the fatigue crack growth da/dN vs. ΔK trend observed in the LCD specimen becomes more comparable to those of obtained from AR and PC materials due to the continuous and sharp cracking mode in all specimens.

9.4.3 Fractography

Macroscopic fracture surface pictures for the AR, PC and LCD specimens after FCG test completion are shown in Figure 9.27, Figure 9.28 and Figure 9.29, respectively. As seen in Figure 9.27 and Figure 9.28, three regions can be identified on the fracture surface of the AR and PC specimens. The first region in these samples is the pre-fatigue cracking zone (which is marked by heat tinting prior to FCG testing), followed by

the fatigue crack growth region and finally the fast fracture (due to break open) zone. The fracture surface of the LCD3 specimen in Figure 9.29 shows that the four regions distinguished in this sample are the pre-crack introduced by an EDM notch, CCG region (which is dark due to oxidation at elevated temperatures), FCG zone and finally the fast fracture (due to break open) region.

It can be observed in Figure 9.27(a) and Figure 9.28(c) that pre-fatigue cracking in specimens FCG-AR1 and FCG-PC3 was non-uniform possibly due to misalignments in the test set up. Therefore some error may be encountered in early stage FCG test results from these specimens. However, uniform fatigue crack extension is observed in the rest of the fracture surface region for both of these specimens. As a result of non-uniform pre-cracking, slightly higher FCG trends have been obtained from FCG-AR1 specimen as seen in Figure 9.25(a). However, observed in Figure 9.25(b) is that FCG-PC3 sample was insensitive to non-uniform pre-cracking since the FCG results are coincident with those of obtained from FCG-PC1 and FCG-PC2 specimens.

A 3 mm slice has been sectioned from the mid thickness of the LCD3 specimen for metallography purposes. This slice has been ground, polished and electro etched to perform a metallurgical study on the cracking mode under prior high temperature creep and subsequent room temperature fatigue crack growth testing conditions. The variation of the main crack path in LCD3 specimen subjected to different loading conditions is shown in Figure 9.30. The late stage CCG/early stage FCG zone is bounded in a box in Figure 9.30 and shown in a higher magnification for a better observation. Also shown in this figure are the average crack extensions due to CCG and FCG testing which are measured by digital imaging on the fracture surface of the specimen.

As seen in Figure 9.30, under static CCG testing conditions the cracking mode is intergranular whereas transgranular crack growth can be observed under cyclic loading FCG deformation. At the intermediate region, an interaction between CCG and FCG failure mechanisms is expected to occur. Therefore combined intergranular-transgranular crack growth can be observed at the early stage of FCG zone in Figure 9.30. This is due to existence of micro cracks (creep damage) on the grain boundaries ahead of the creep crack tip due to the prior CCG deformation. As a result of creep damage formation, although FCG deformation tends to be transgranular, but pre-existing damages on the grain boundaries introduce high stress concentrations and lead to an interaction between intergranular and transgranular deformations under cyclic loading conditions at the early stage FCG test on the LCD specimen. As seen in

Figure 9.30 dominant transgranular fatigue cracking mode can be observed further away from the creep damage zone where intergranular damage doesn't exist or is insignificant.

A more detailed SEM study of the cracking mode has been performed on the fracture surface of the LCD3 specimen and shown in Figure 9.31. As seen in this figure, intergranular and transgranular cracking modes are observed on the fracture surface in the CCG and FCG zones, respectively. However, the SEM pictures in Figure 9.31 taken along a line corresponding to the average creep crack length measured on the fracture surface show that a combination of intergranular and transgranular cracking modes is observed at the end of CCG/beginning of FCG region. This implies an interaction between intergranular and transgranular failure mechanisms in this region. Further shown in Figure 9.31 are the SEM pictures taken along a line in the FCG region with an offset of around 1 mm away from the average creep crack length measured on the fracture surface. As seen in Figure 9.31, although the dominant failure mechanism in the FCG zone is transgranular, some intergranular creep damage can be clearly observed along the grain boundaries in this region which is fairly close to the CCG area. The existence of these intergranular micro cracks ahead of the final creep crack tip indicates that the local creep damage may have influences on the subsequent FCG failure mechanism of the material, depending on the significance of the intergranular creep damage density in that region.

9.5 Discussion

All CCG tests were performed under nominally the same laboratory testing conditions, but still some discrepancy can be observed in the time data at which the desired PD rate is obtained in LCD specimens. This can be related to the experimental errors, specimen machining, material variability due to service conditions and possibly small differences between prior pre-compression levels of the blocks from which the specimens were manufactured.

Comparison of the load *vs.* LLD trend of the AR and PC specimens show that generally larger loads and smaller displacements are observed in the PC material. This can be associated with the hardening effects and lower tensile failure strain in the PC material. Also the higher fracture energy, J , in the AR compared to the PC for a given amount of crack extension indicates that there is larger resistance against fracture in the AR material. The difference between R-curve trends in the AR and PC materials may be associated with the intergranular voids introduced into the material by room temperature pre-compression prior to specimen manufacturing. In other words pre-

compression generally introduces some significant amount of damage into the material which subsequently influences the R-curve trends by decreasing the tensile strain at failure.

Comparison of the measured values of crack extension using digital imaging (see Table 9.4) to those estimated by the unloading compliance (see Figure 9.9) shows that similar results are found for JIC-AR1 specimen. However due to non-uniform pre-cracking in specimen JIC-AR2 some large discrepancy has been observed between the average measured and the estimated values of crack growth due to fracture toughness for this specimen. Similar comparison on the JIC-PC1 and JIC-PC2 samples shows that very good agreement is found between the estimated and measured values of crack extension for both specimens, although non-uniform pre-cracking is observed on the fracture surface of the JIC-PC2 sample.

Some uncertainties are apparent in the crack extension estimates made using the unloading compliance technique for LCD samples. Comparison of the measured values of crack extension using digital imaging on LCD specimens (see Table 9.4) to those predicted from the unloading compliance in Figure 9.9 shows that the total crack growth due to fracture toughness is over estimated by the initial R-curve generated using the raw data. In LCD specimens significant creep damage zone exists in the vicinity of the creep crack tip which comprises micro cracks and voids along the grain boundaries. Therefore, the largest error in crack length estimates using the unloading compliance is expected to occur at the initial crack length where the creep damage is present and thus to exclude the uncertainties encountered it is sensible to remove the tail regions from the R-curves for LCD specimens. In addition, some compressive residual stress of around -100 MPa is expected to exist at the crack tip along the loading direction. Therefore, difficulties may be encountered in the initial unloading compliance measurement within the creep damage zone in LCD specimens. Knowing that the residual stresses in LCD 316H specimens only exist within a small distance from the crack tip and they will be washed out at high loads, it is believed that creep damage effects are more influential on the fracture toughness behaviour of the material containing creep damage.

Comparison of the $J_{0.2/BL}$ in the AR and PC specimens shows that pre-compression reduces the fracture energy for a given amount of crack extension. Similarly, when the fracture response of LCD specimens is investigated in regions outside the creep damage zone by employing modified R-curves in the analyses, a drop in the fracture energy at the crack initiation stage is observed in LCD specimens compared to the PC

material. Figure 9.10 shows that by modifying the fracture resistance curves for LCD specimens, by excluding predicted crack extensions at lower values of J , a decrease in the fracture energy of LCD specimens is also observed in the latter half of the modified R-curves. Note that as explained in Chapter 3-Section 3.1.2, the global creep damage leads to an increase in the yield stress, decrease in the tensile strain at failure and a severe drop in the fracture energy of the material. Therefore, creep damage doesn't only affect the R-curve trends in the regions local to the initial crack tip, but the fracture behaviour of the material away from the creep damage zone is also altered in LCD specimens.

The fracture toughness values obtained from LCD specimens by assuming inaccuracies being encountered in unloading compliance measurements within the entire range of crack extension, and not only at the initial crack tip, and thus correcting the R-curves using initial and final crack length measurements on the fracture surface show that a more significant drop is observed in specimens containing creep damage compared to those of obtained from modified R-curves, when the tail regions are considered valid and taken into account in the analysis. Comparing Figure 9.9, Figure 9.10 and Figure 9.17 it can be observed that the fracture resistance of the LCD material is generally lower than the PC material. However by considering different approaches to correct the R-curve data obtained from LCD specimens, the fracture toughness values in the crept material decrease by different factors compared to those of obtained from PC material containing no creep damage. The largest factor is attained when the tail region is considered in the analysis and the R-curve trends are corrected using the initial and final crack length measurements made on the fracture surface whereas lower fracture toughness values by a smaller factor are obtained in the LCD material when the tail regions are removed and modified R-curves are employed in the analysis.

It was shown in Figure 9.22 that for a given number of cycles, the amount of crack extension in the AR material is larger than the PC specimens. However, seen in Figure 9.25 is that $daldN$ vs. ΔK trends are insensitive to the pre-compression and similar Paris law constants are found for the AR and PC materials. This is due to the fact that although room temperature pre-compression increases the yield stress of the material, but also the tensile strain at failure drops as previously shown in Figure 5.1(a). Therefore, under cyclic loading conditions competing higher yield stress and lower failure strain effects cancel out each other and the FCG behaviour of the PC material remains unchanged compared to the AR.

It was observed in Figure 9.23 that the crack initiation occurs after a considerably large number of cycles in the LCD3 specimen. Also seen in Figure 9.26 is that for the majority of the creep material fatigue life, the crack growth rate per cycle in the LCD3 specimen is severely lower than that of observed in the PC material (which contains no creep damage). Though at larger values of ΔK , where significant amount of fatigue crack extension has occurred and the crack tip is expected to be far from the initial local creep damage zone, the FCG behaviour of the LCD3 specimen tends to the FCG trends obtained from the PC material. Late initiation and small da/dN in the LCD material at early stage fatigue crack growth are due to a number of reasons which are explained and discussed below.

- i. Local creep damage effects: When creep damage local to the crack tip is introduced into the material by interrupting a CCG test, intergranular voids and micro cracks form ahead of the final crack tip. It is also believed that towards the end of the CCG test, some discontinuity occurs in the main creep crack path within a region close to the final crack tip. Therefore when a CCG test is interrupted, a high creep damage density in the form of voids and micro cracks is expected to exist behind the final discontinuous crack tip and also a creep damage zone with relatively lower extent of creep damage density is expected to form ahead of the final crack tip. A schematic illustration of the creep damage formation in an interrupted CCG tests is presented in Figure 9.32.

The micro cracks formed behind the final crack tip may have a significant contribution to the late initiation of the crack under subsequent cyclic fatigue loading conditions. In other words, these micro cracks need to link up with the main crack path before the subsequent FCG can start to happen. Furthermore, the formation of intergranular micro cracks behind and ahead of the final crack tip significantly influences the unloading compliance measurements. Therefore large percentages of error are introduced in the crack length estimates at the beginning of the FCG test on LCD specimens as seen in Table 9.7. Due to unbroken intergranular ligaments in the regions behind and in front of the final crack tip, the unloading compliance underestimates the initial crack tip. However, when a continuous crack path forms, relatively accurate estimations using the unloading compliance may be obtained outside the intergranular creep damage zone.

As seen in Table 9.7, the overestimated fatigue crack extension compared to the values measured from the fracture surface suggests excluding the first few millimetres (around 3 mm) of the estimated crack propagation using the unloading

compliance in Figure 9.26. This may help quantifying a more realistic value of da/dN at the beginning of the FCG test on the LCD3 specimen which would be approximately an order of magnitude lower than that of obtained from the FCG test on the PC material. Similar exclusion of the first few millimetres of crack extension due to large errors in the initial crack length estimates using the unloading compliance has also been applied to analyse the fracture toughness test data on LCD specimens in a more realistic manner.

- ii. Change in the cracking mode: As seen in Figure 9.30 and Figure 9.31, intergranular crack growth occurs under CCG testing conditions whereas the cracking mode is transgranular under high frequency FCG cyclic loading conditions. The interaction between the intergranular and transgranular cracking modes in the creep damage zones behind and in front of the final creep crack tip is expected to have substantial effects on the fatigue crack initiation and early stage FCG behaviour of the material. Although FCG is a transgranular process, but as seen in Figure 9.30 the presence of voids and micro cracks along the grain boundaries (formed due to prior CCG) facilitates intergranular crack growth even under cyclic loading conditions. Thus intergranular defects, which introduce large stress concentration along the grain boundaries in the creep damage zone, interact with the transgranular fatigue cracking mode and lead to late fatigue crack initiation and low fatigue crack growth rates per cycle at the early stage FCG test on LCD specimens.
- iii. Global creep damage effects: As discussed in Chapter 3-Section 3.1.2, it has been shown in the literature that the global creep damage introduced into an uncracked body leads to an increase in the yield stress and a drop in the tensile strain at failure of the material. The higher yield stress in the crept material may act as an additional resistance against crack initiation in the FCG tests on LCD specimens. However, further away from the initial creep crack tip the increase in the yield stress and the decrease in the tensile failure strain of the crept material leads to FCG trends similar to those of obtained from the PC and AR materials. In other words, due to similar effects of pre-compression and creep damage on the yield stress and tensile failure strain of the material, it is expected that the FCG trends in the globally creep damaged material will be similar to those of obtained from the AR material. Further tests on global creep damage material are required to confirm this trend.
- iv. Residual stress effects: As shown in Section 9.2.2, compressive residual stress of around -100 MPa is expected to exist at the crack tip of the LCD specimens along the loading direction. According to ASTM E647 [107] residual stresses may

significantly influence the fatigue crack growth behaviour of the material. Although as the crack initiates and advances the residual stress partially relieves and redistributes (by decreasing the peak residual stresses ahead of the crack tip), the remaining magnitude of residual stress may still have some effects on the FCG behaviour of the material. Especially in the case of compressive residual stresses, this effect becomes more significant. The compressive residual stress is superimposed with the applied cyclic stresses and results in lower effective cyclic stresses. This may particularly make difficulties for initiating a crack under cyclic loading conditions since the large magnitude of compressive residual stress may lead to crack closure even though a tensile cyclic load is applied on the sample. Therefore, as observed in FCG experiments on LCD specimens, the compressive residual stresses may influence the initiation and early stage FCG behaviour of the material. Note that the effective R -ratio in FCG tests on specimens containing tensile and compressive residual stress is higher and lower than the applied R -ratio, respectively. However, as the residual stress releases due to crack extension, the effective R -ratio and the applied R -ratio will converge. Therefore for the case of FCG tests on LCD specimens where compressive residual stresses exist at the crack tip, the effective R -ratio of smaller than 0.1 may lead to a lower da/dN at the early stage fatigue crack growth.

It has been shown in [148] that when tensile residual stresses are introduced in C(T) specimens (by means of mechanical non-uniform pre-compression on a specimen containing a blunt notch) made of 347 austenitic Type stainless steel and the specimens are tested in thermal soaking or primary loading creep conditions at 650 °C, a progressive reduction in the tensile residual stress peaks (measured by Neutron Diffraction) can be observed as the crack advances. Similar reduction in the peak residual stresses have been reported in [149] for combined loading creep crack growth tests on C-ring specimens made of 1Cr1Mo1/4V steel which were subjected to mechanical non-uniform pre-compression to introduce residual stresses prior to CCG testing. The significant drop in peak tensile residual stresses with the crack extension reported in [148, 149] indicates that progressive reduction in compressive residual stresses may be expected in FCG tests on LCD samples. Therefore, the remaining residual stresses in LCD specimens at each increment of fatigue crack extension may have some influences on the FCG behaviour of the material. Note that the effect of residual stress in room temperature FCG tests is considerably larger than high temperature CCG tests. This is because of the residual stress relaxation under creep conditions which doesn't occur in FCG tests

and thus although some residual stresses are released due to the crack extension but the remaining amount may still affect the FCG behaviour of the material.

9.6 Summary

- Creep damage local to the crack tip has been introduced into the material by interrupting CCG tests on C(T) specimens, loaded under $K(a_0) = 25.5 \text{ MPa}\sqrt{\text{m}}$, at 550°C .
- A compressive residual stress of around -100 MPa is expected at the crack tip of the LCD specimens in a direction normal to the crack plane. This residual stress at the crack tip need to be considered in the subsequent fracture toughness and fatigue crack growth analyses.
- The micro hardness map ahead of the final crack tip shows that creep damage slightly reduces the hardness of the material close to the crack tip, but values similar to those of measured on the PC material are observed away from the creep damage zone.
- Room temperature fracture toughness tests have been performed on two AR, two PC and two LCD specimens.
- Larger LLD in the fracture toughness tests on the AR material is observed compared to the PC and LCD specimens.
- $J_{0.2/BL}$, $K_{0.2/BL}$ and $J_{0.2}$ values are found larger in the AR material compared to the PC.
- The unloading compliance technique doesn't provide accurate estimates of the instantaneous crack length at early stages of fracture toughness tests on LCD specimens. This is due to the existence of unbroken ligaments and micro cracks/voids (and thus discontinuous crack) in the creep damage zone around the crack tip.
- Smaller values of $J_{0.2/BL}$, $K_{0.2/BL}$ and $J_{0.2}$ have been found in LCD specimens compared to the PC material when modified R-curves are employed in the analyses. Considerably lower amount of fracture energy is also required in the LCD material at larger crack extensions in regions away from the creep damage zone.
- By considering the tail regions being valid in the R-curves and correcting the crack length estimates using the measured values of initial and final crack lengths, significantly lower values of fracture toughness were obtained in the LCD material compared to the PC.

- High frequency FCG tests have been performed on three AR, three PC and a LCD specimen at room temperature under $R = 0.1$ and $f = 10$ Hz. The instantaneous crack length during FCG tests was estimated using the unloading compliance technique.
- At relatively large number of cycles, the crack extensions in the AR material are found greater than those of obtained from the PC material due to the higher level of yield stress in the pre-strained material.
- A substantial delay is observed in the crack initiation of the LCD3 specimen whereas the number of cycles required to initiate cracks in pre-fatigue cracked AR and PC specimens can be considered negligible.
- Due to the difficulties encountered in the fatigue crack initiation of the LCD3 specimen, a higher initial ΔK was applied on this sample and thus a larger K_{max} per cycle for a give crack length was applied on this specimen compared to the AR and PC materials.
- To avoid introducing additional inelastic damage into the LCD material during FCG tests, the initial K_{max} applied on the LCD3 specimen was chosen to be lower than that of previously applied on this sample during the CCG test, $K(a_c)$.
- The pre-compression process has insignificant effects on the fatigue crack growth behaviour of the material at room temperature.
- Lower fatigue crack growth rate per cycle is obtained in the LCD3 specimen compared to the AR and PC specimens at the early stage FCG. Though, the da/dN vs. ΔK trend for LCD3 specimen converges with the line of best fits to the AR and PC materials at large values of ΔK .
- The late crack initiation and low da/dN values found in the early stage FCG test on LCD3 specimen are expected to be because of:
 - i. Local creep damage effects around the creep crack tip which appear in the form of voids, micro cracks and unbroken ligaments (and thus crack discontinuity)
 - ii. Change in the cracking mode from intergranular (due to CCG deformation) to transgranular (due to FCG deformation)
 - iii. Global creep damage effects which lead to an increase in the yield stress and a drop in the tensile failure strain of the material
 - iv. Compressive residual stress effects at the crack tip

9.7 Tables

Table 9.1: Creep crack length measurements and plastic zone size estimates in LCD specimens

Test ID	a_c (mm)	$K(a_c)$ (MPa√m)	$r_{p,c}$ (mm) PS	$r_{p,c}$ (mm) PE
LCD1	27.1	43.3	5.92	1.97
LCD2	27.7	45.1	6.43	2.14
LCD3	24.7	37.0	4.35	1.45
LCD4*	-	-	-	-

*Test is yet to be done on this specimen

Table 9.2: Normalised reference stress values and plastic zone size estimates in specimen 8PC-A2, calculated at the final crack length

Test ID	P (kN)	$K(a_f)$ (MPa√m)	a_f/W	$\sigma_{ref} / \sigma_{0.2}$ PS	$\sigma_{ref} / \sigma_{0.2}$ PE	$r_{p,f}$ (mm) PS	$r_{p,f}$ (mm) PE
8PC-A2	18.3	40.25	0.52	1.0	0.7	2.06	0.69

Table 9.3: C_1 and C_2 correction constants for the fracture toughness tests

Test ID	C_1 (kN/mm)	C_2 (mm ⁻¹)
JIC-AR1	145.9	20.6
JIC-AR2	139.4	53.2
JIC-PC1	114.2	137.3
JIC-PC2	127.8	55.3
LCD1	251.9	42.2
LCD2	314.2	31.6

Table 9.4: Fracture toughness specimens dimensions

Test ID	W (mm)	B (mm)	B_n (mm)	a_i (mm)	a_f (mm)	Δa (mm)
JIC-AR1	50.0	25.0	20.1	25.3	28.1	2.8
JIC-AR2	50.1	25.0	20.0	26.0	29.3	3.3
JIC-PC1	50.1	25.0	20.0	26.6	29.5	2.9
JIC-PC2	50.0	25.0	20.0	24.7	27.3	2.6
LCD1	50.0	25.0	17.5	27.1	30.4	3.3
LCD2	50.0	25.0	17.5	27.7	31.9	4.2

Table 9.5: Equations of the line of best fits to the valid data points in fracture toughness R-curves

Test ID	J (MPam) =
JIC-AR1	$-0.1345\Delta a^2 + 0.9742\Delta a + 0.1037$
JIC-AR2	$-0.0610\Delta a^2 + 0.6586\Delta a + 0.0505$
JIC-PC1	$-0.0500\Delta a^2 + 0.4093\Delta a + 0.0952$
JIC-PC2	$-0.0815\Delta a^2 + 0.4596\Delta a + 0.0807$
LCD1	$-0.0571\Delta a^2 + 0.3696\Delta a + 0.0825$
LCD2	$-0.0635\Delta a^2 + 0.4282\Delta a + 0.0425$

Table 9.6: A summary of the fracture toughness test results

Test ID	Δa_{max} (mm)	J_{max} (MPam)	Blunting line slope	Ω	$J_{0.2/BL}$ (MPam)	$K_{0.2/BL}$ (MPa $\sqrt{m}$)	$2dJ/d\Delta a$ $_{(0.2/BL)}$	$J_{0.2}$ (MPam)
JIC-AR1	2.48	0.57	2.334	36.53	0.47	327.06	1.73	0.29
JIC-AR2	2.41	0.55	2.334	24.38	0.25	235.37	1.24	0.18
JIC-PC1	2.34	0.50	2.333	13.63	0.21	223.58	0.76	0.17
JIC-PC2	2.53	0.54	2.333	11.74	0.21	221.91	0.83	0.17
LCD1	2.29	0.49	2.333	10.63	0.18	207.54	0.68	0.15
LCD2	2.24	0.48	2.333	12.44	0.15	190.06	0.79	0.13

Table 9.7: Fatigue crack growth specimens dimensions

Test ID	W (mm)	B (mm)	B_n (mm)	a_i (mm)	a_f (mm)	Δa (mm)	a_i (mm)	a_f (mm)	% Error	
				Measured			Estimated		a_i	a_f
FCG-AR1	50.0	25.0	17.5	26.2	37.6	11.4	25.0	37.8	4.7	0.4
FCG-AR2	50.0	25.0	17.5	25.9	37.4	11.5	25.0	37.7	3.5	0.8
FCG-AR3	50.0	25.0	17.5	26.1	35.0	8.9	25.6	35.0	2.0	0.0
FCG-PC1	50.0	25.0	17.5	25.5	37.9	12.4	25.0	37.9	2.1	0.1
FCG-PC2	50.0	25.0	17.5	25.6	37.8	12.2	25.0	38.4	2.3	1.7
FCG-PC3	50.0	25.0	17.5	25.6	34.5	8.9	24.7	35.13	3.4	1.8
LCD3	50.0	25.0	17.5	24.7	34.1	9.4	22.4	34.9	9.4	2.5
LCD4*	50.0	25.0	17.5	-	-	-	-	-	-	-

*Test is yet to be done on this specimen

Table 9.8: Fatigue crack growth power law constants for the tests performed on the AR and PC materials

Test ID	C	m	R^2
FCG-AR1	5.34×10^{-9}	3.20	0.99
FCG-AR2	4.27×10^{-9}	3.20	0.99
FCG-AR3	3.04×10^{-9}	3.30	0.99
FCG-PC1	4.09×10^{-9}	3.17	0.99
FCG-PC2	3.73×10^{-9}	3.20	0.99
FCG-PC3	4.97×10^{-9}	3.12	0.99

9.8 Figures

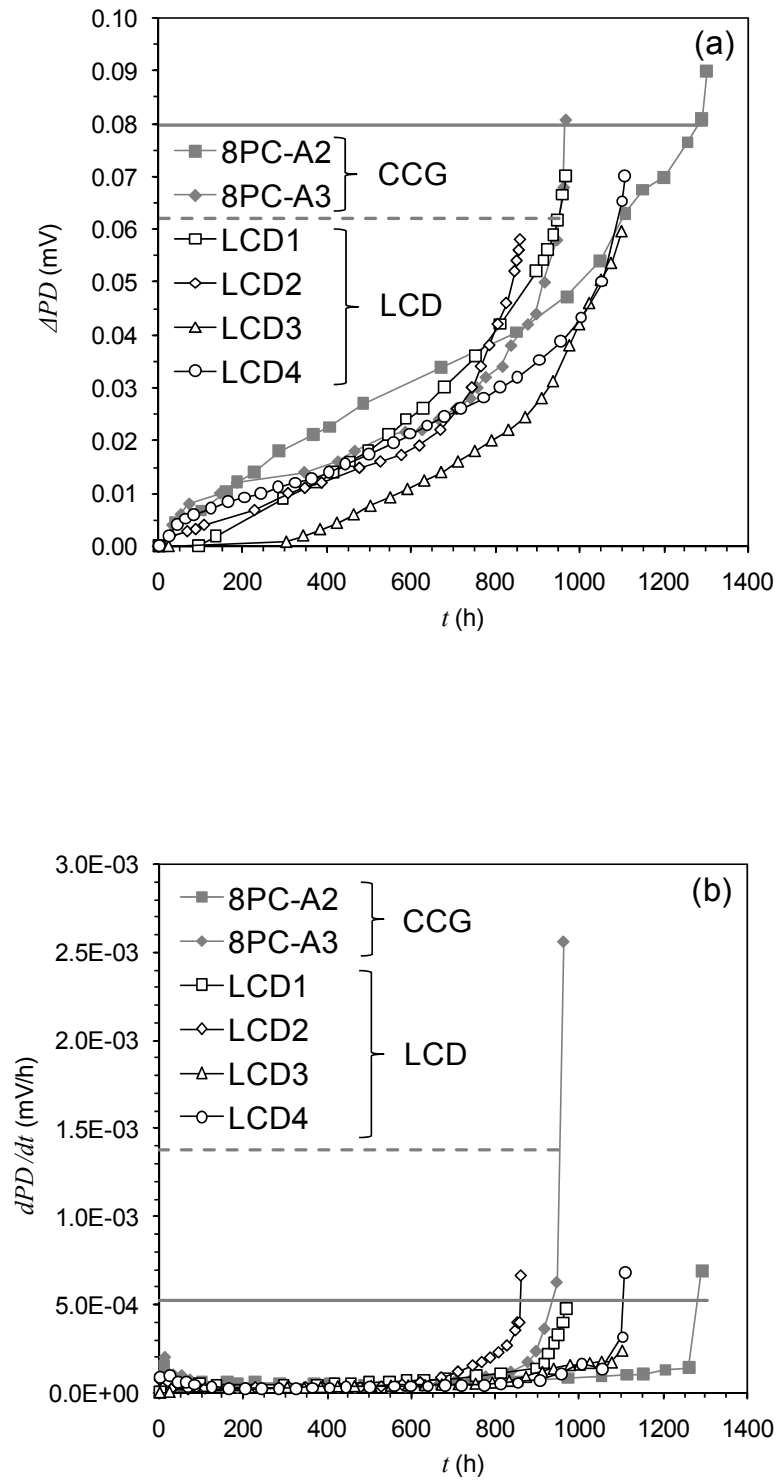


Figure 9.1: Comparison of the (a) ΔPD and (b) PD rate, in LCD specimens to the CCG calibration tests

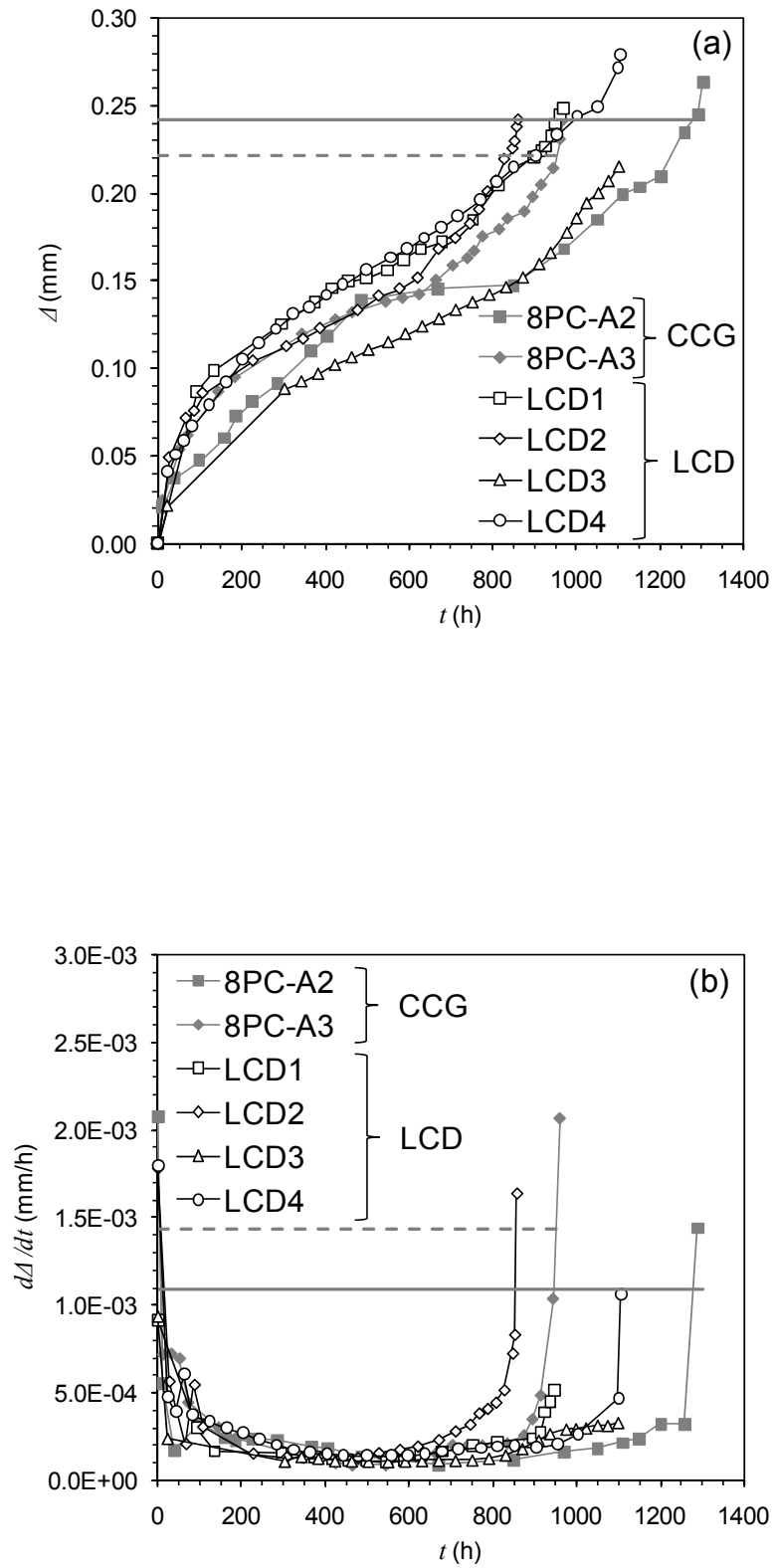


Figure 9.2: Comparison of the (a) LLD and (b) LLD rate, in LCD specimens to the CCG calibration tests

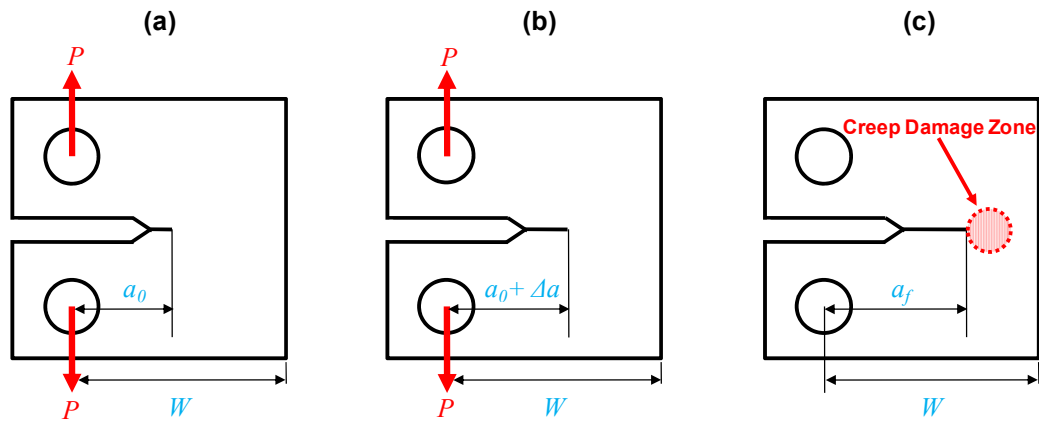


Figure 9.3: Three stages of a CCG test (a) Loading under tension at the initial crack length (b) Creep crack growth under a constant load (c) Unloading and formation of a creep damage zone ahead of the final crack length

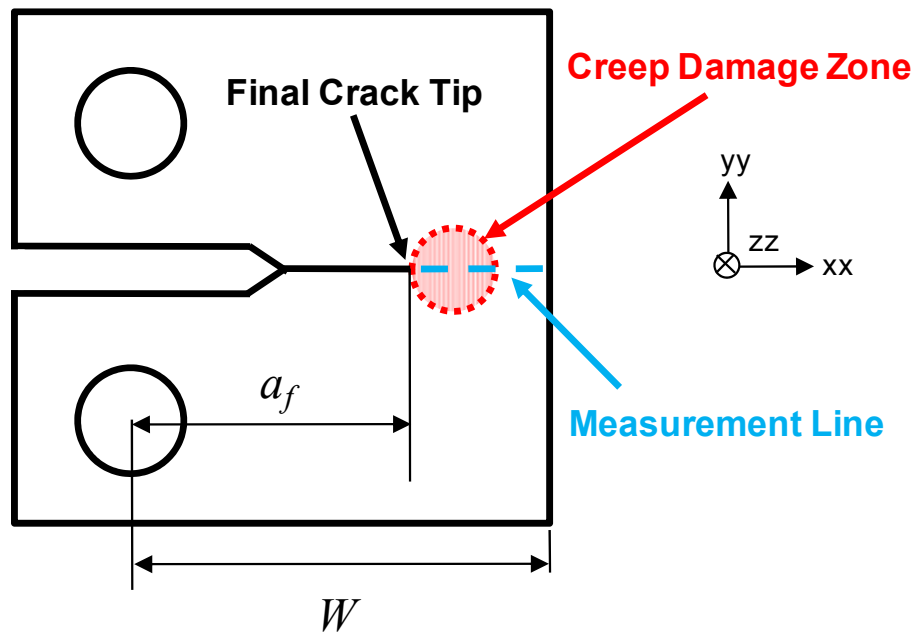


Figure 9.4: Illustration of the ND measurement line for interrupted CCG specimens

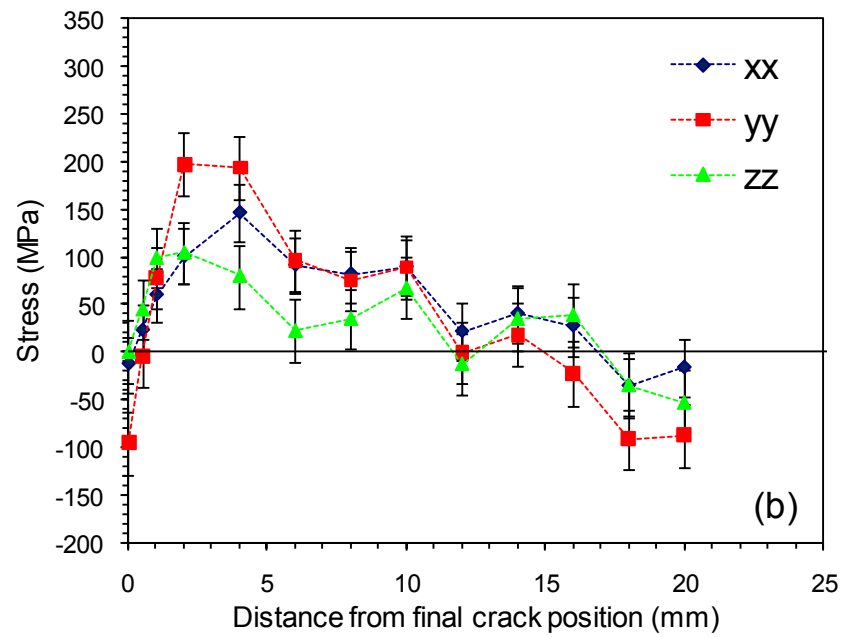
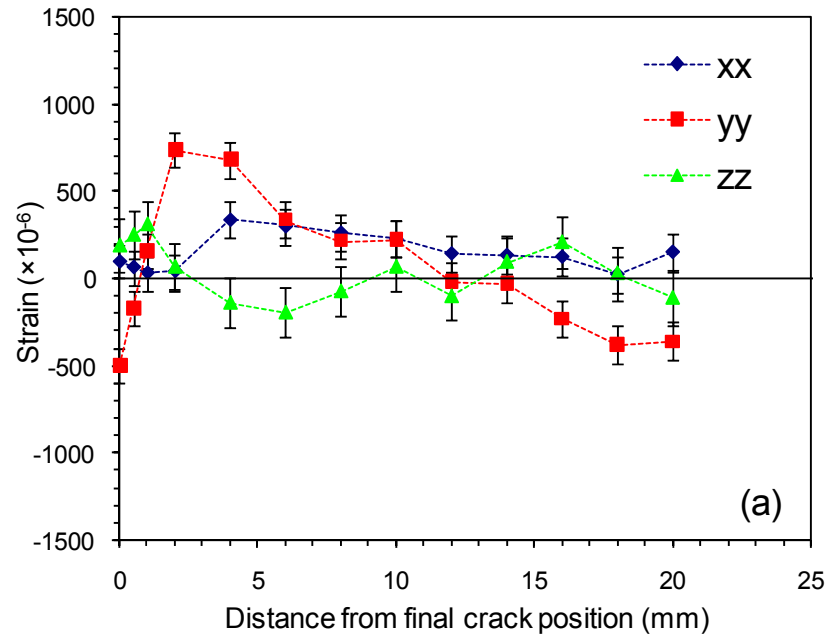


Figure 9.5: Residual (a) strain (b) stress, distribution ahead of the final crack tip in specimen 8PC-A2

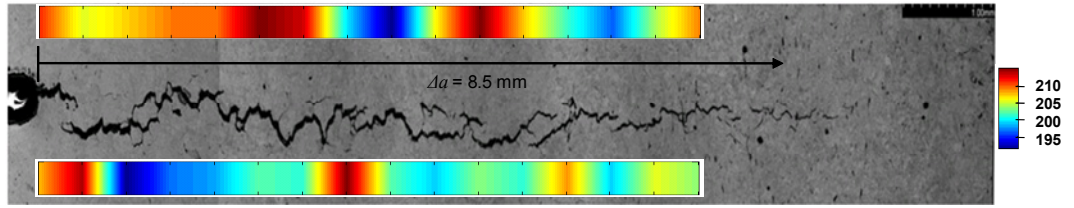


Figure 9.6: Micro hardness linear profiles along the main crack path, 0.5 mm above and below the symmetry line, measured on specimen 8PC-A2

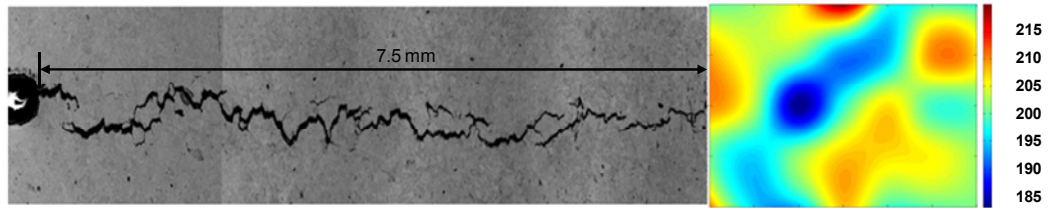


Figure 9.7: Micro hardness map ahead of the final crack tip, measured on specimen 8PC-A2

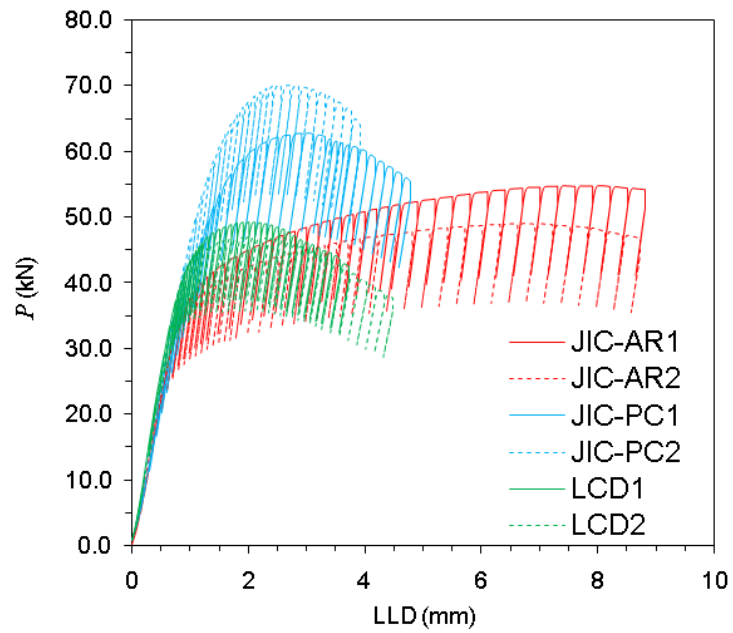


Figure 9.8: Load vs. LLD data for the fracture toughness tests performed on the AR, PC and LCD specimens

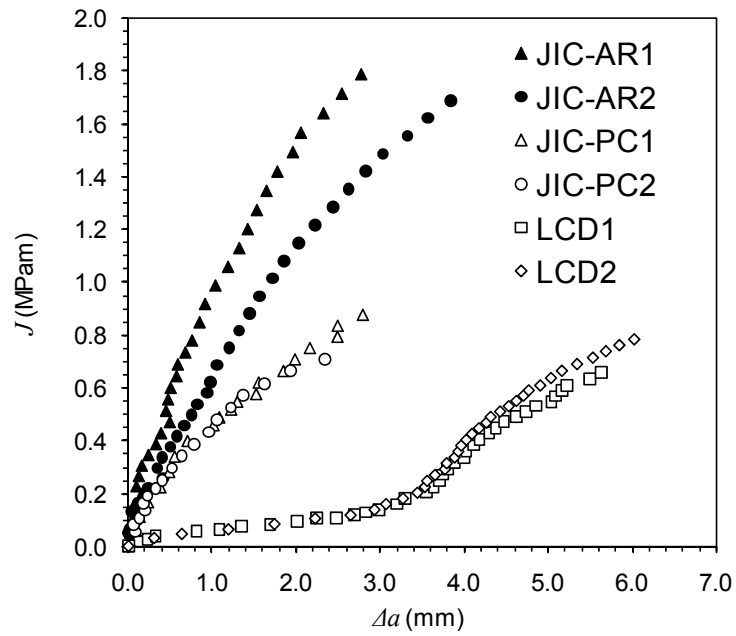


Figure 9.9: The fracture resistance curve for the AR, PC and LCD specimens

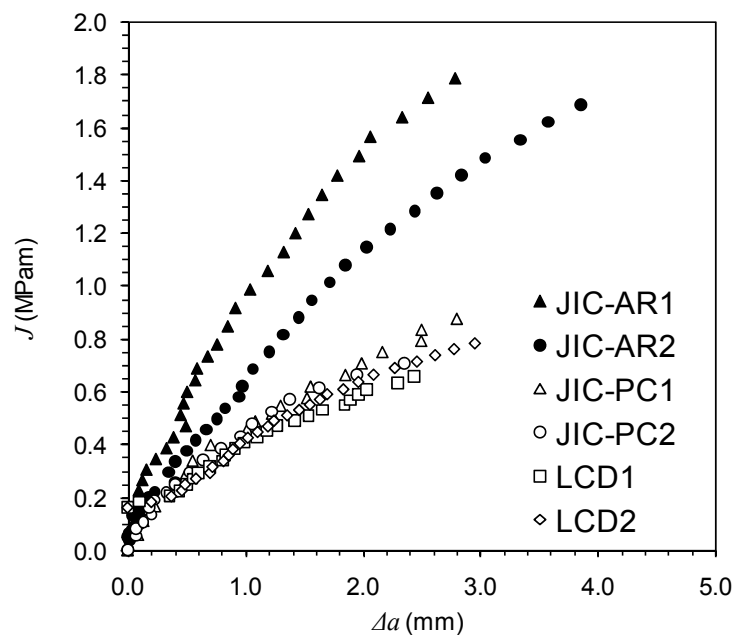


Figure 9.10: The modified fracture resistance curves for the LCD specimens

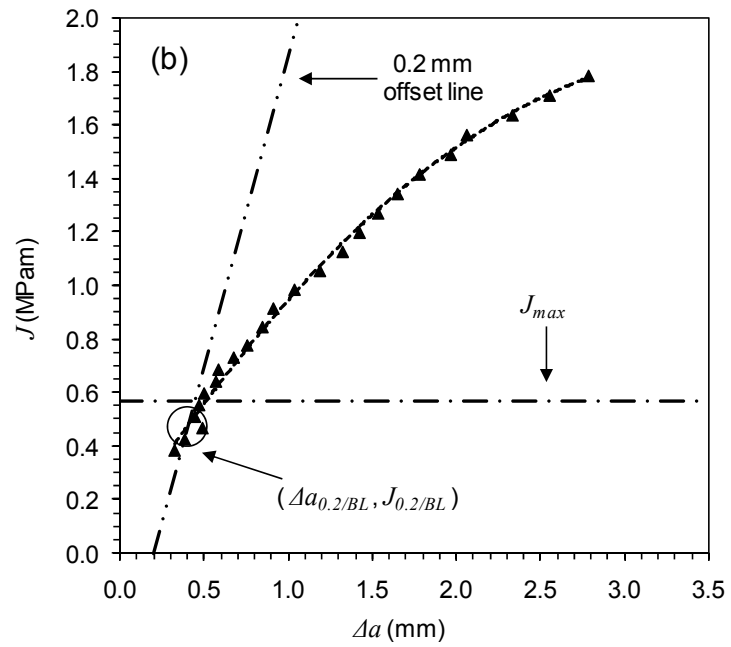
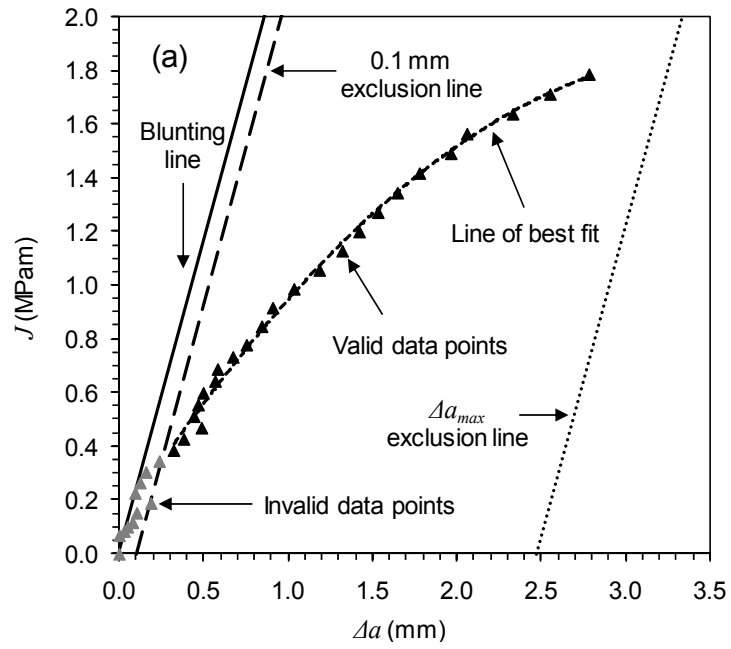


Figure 9.11: Fracture toughness data analysis for JIC-AR1 specimen (a) illustration of the exclusion lines (b) calculation of $J_{0.2/BL}$

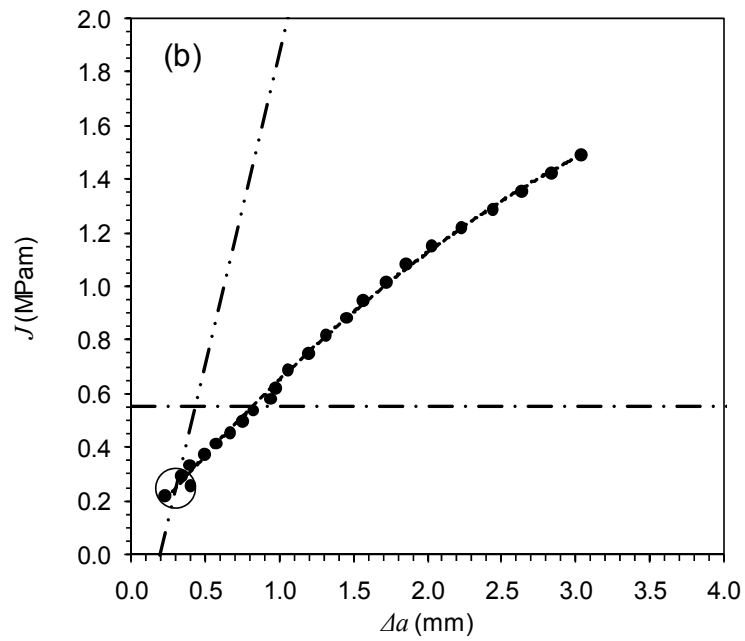
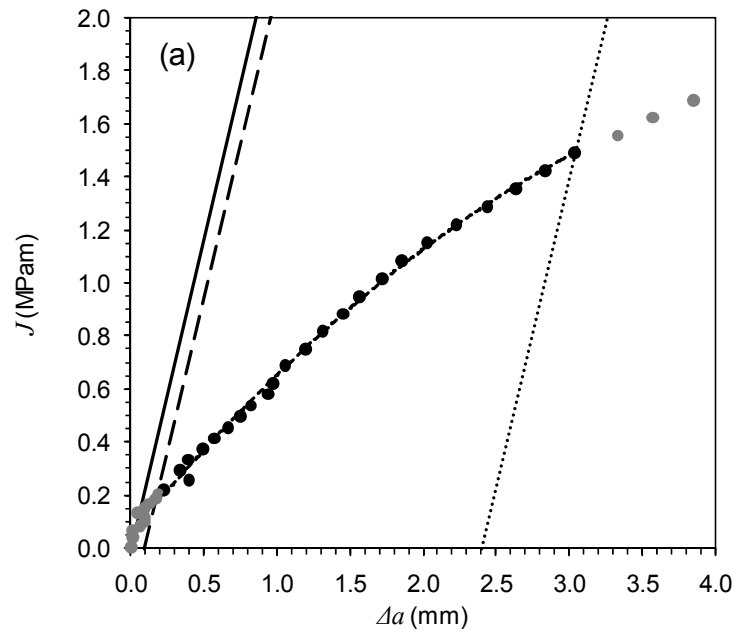


Figure 9.12: Fracture toughness data analysis for JIC-AR2 specimen (a) illustration of the exclusion lines (b) calculation of $J_{0.2/BL}$

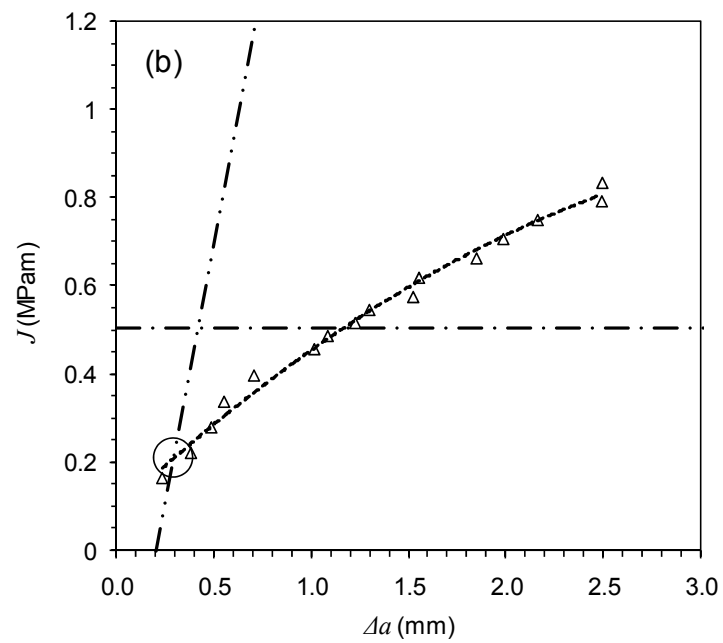
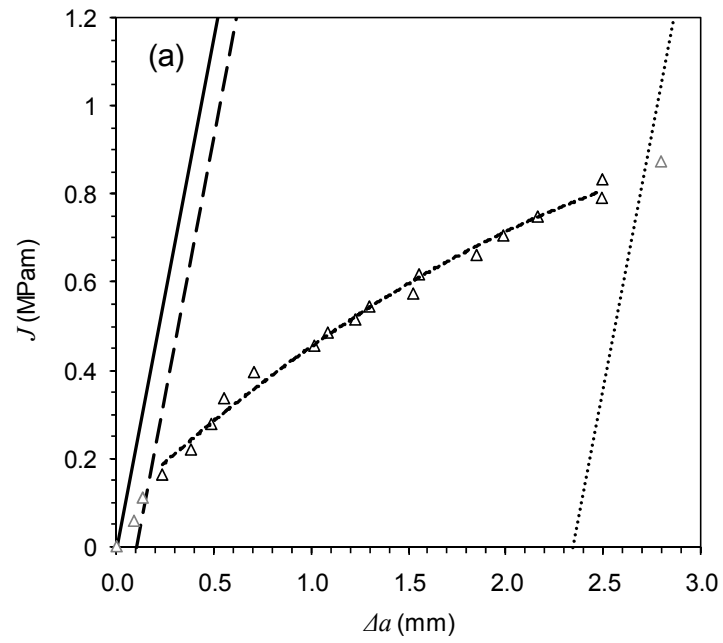


Figure 9.13: Fracture toughness data analysis for JIC-PC1 specimen (a) illustration of the exclusion lines (b) calculation of $J_{0.2/BL}$

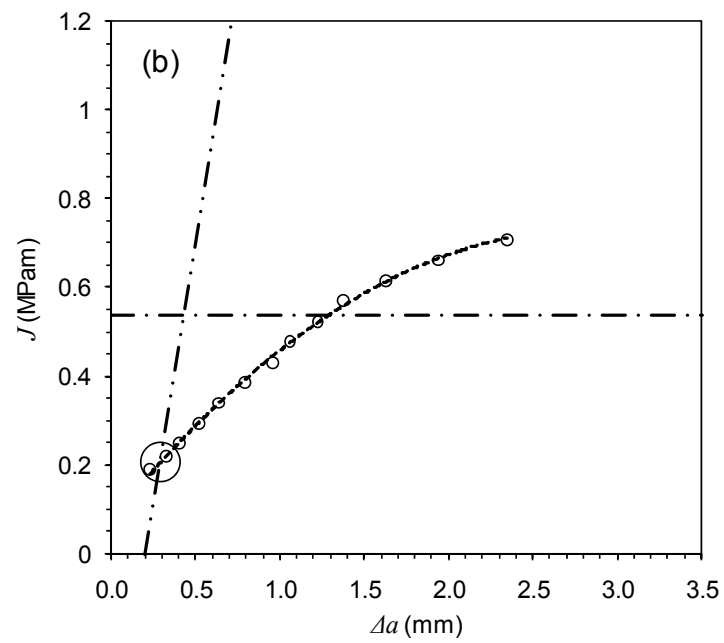
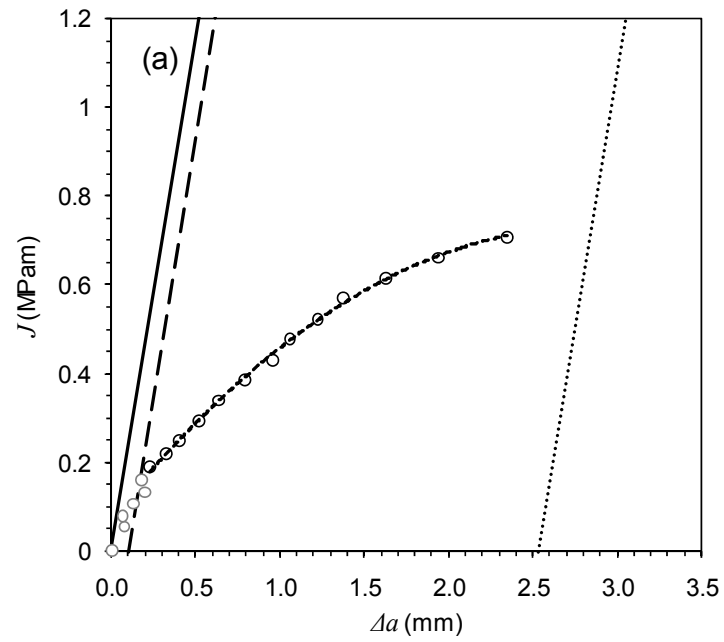


Figure 9.14: Fracture toughness data analysis for JIC-PC2 specimen (a) illustration of the exclusion lines (b) calculation of $J_{0.2/BL}$

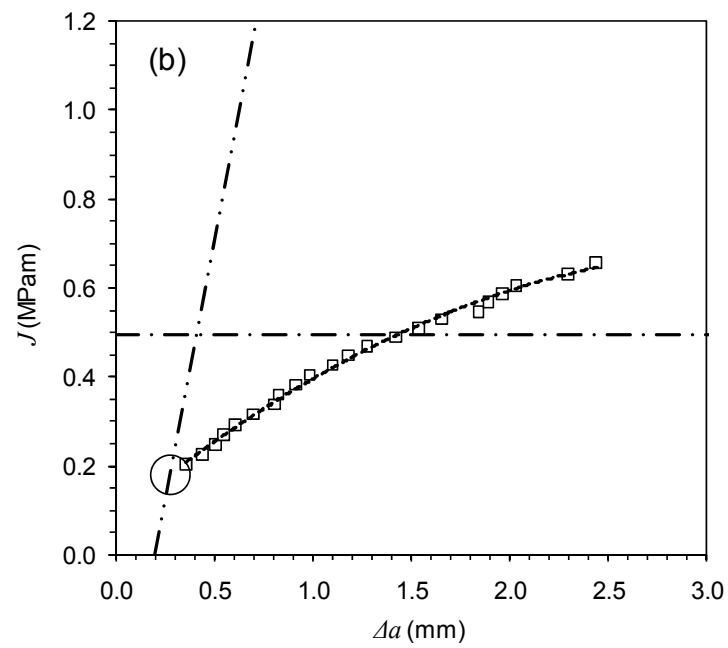
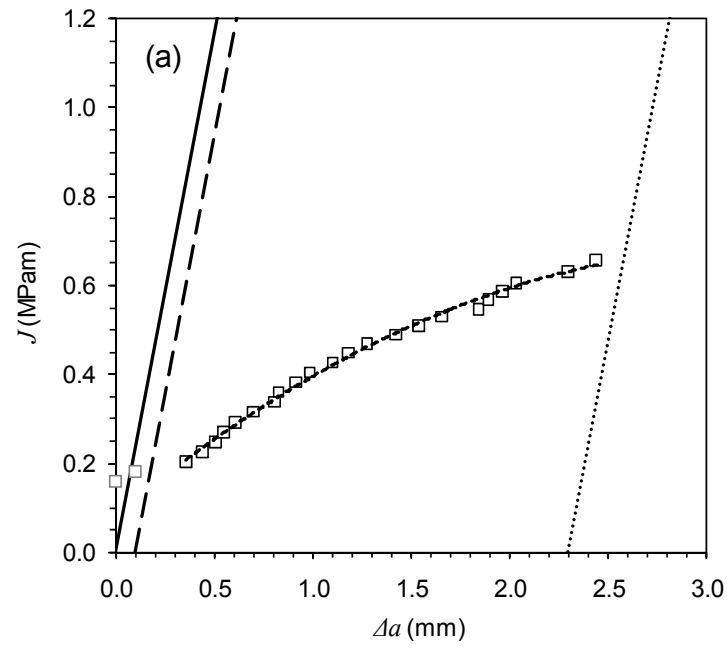


Figure 9.15: Fracture toughness data analysis for LCD1 specimen (a) illustration of the exclusion lines (b) calculation of $J_{0.2/BL}$

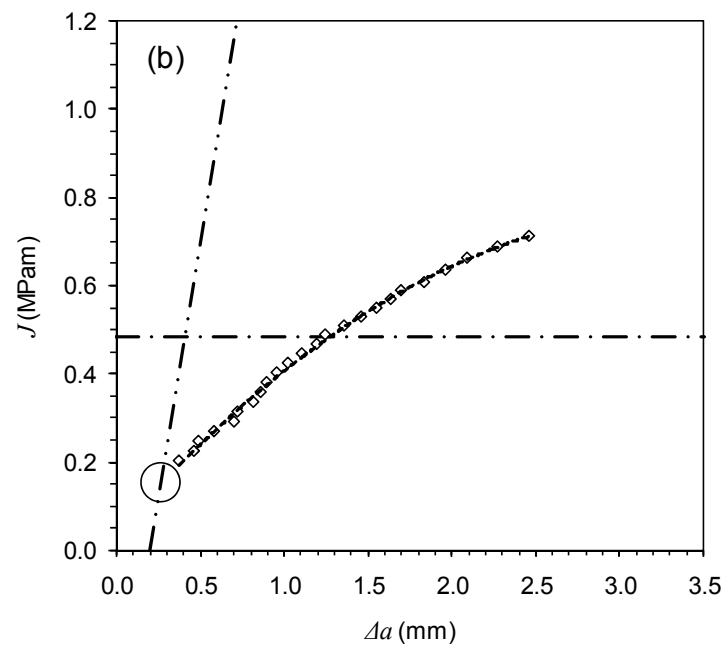
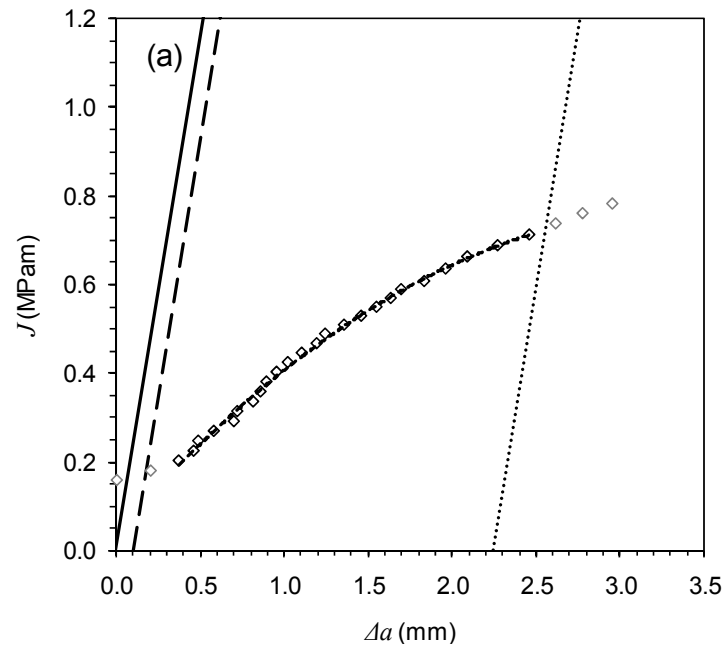


Figure 9.16: Fracture toughness data analysis for LCD2 specimen (a) illustration of the exclusion lines (b) calculation of $J_{0.2/BL}$

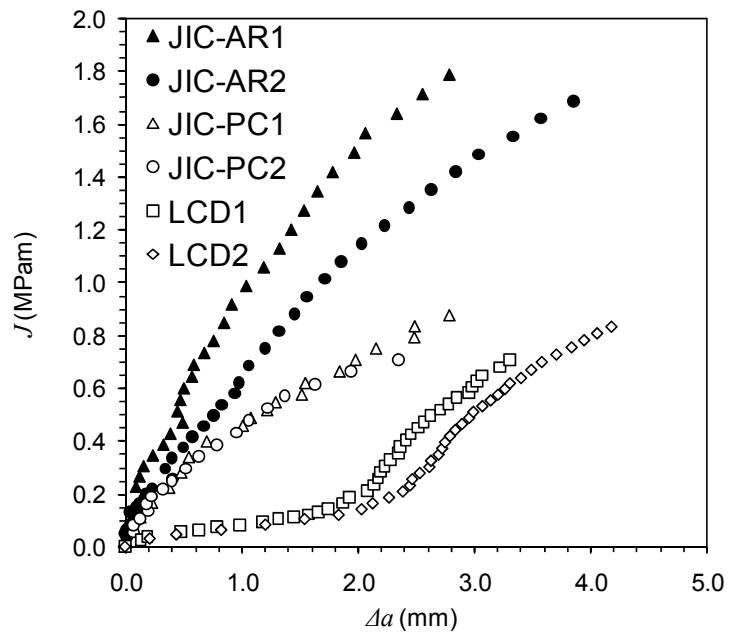


Figure 9.17: Comparison of the fracture resistance curves for AR and PC materials to the LCD specimens, after correcting the crack length measurements

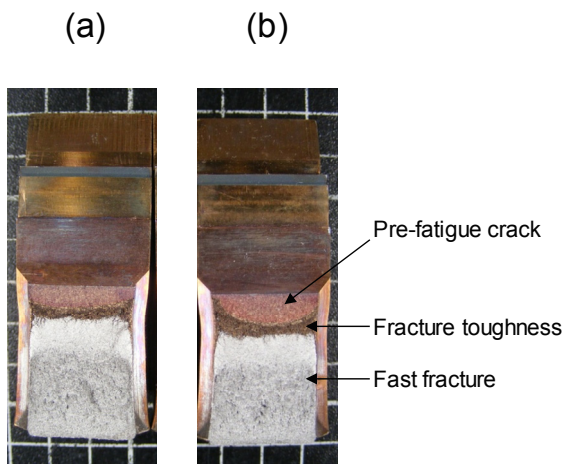


Figure 9.18: Fracture surfaces of (a) JIC-AR1 (b) JIC-AR2

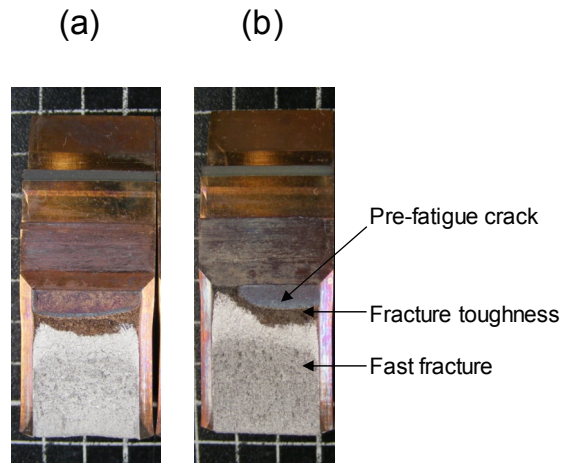


Figure 9.19: Fracture surfaces of (a) JIC-PC1 (b) JIC-PC2

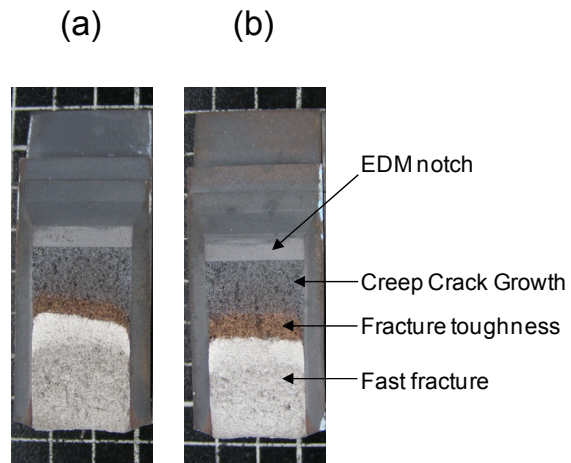


Figure 9.20: Fracture surfaces of (a) LCD1 (b) LCD2

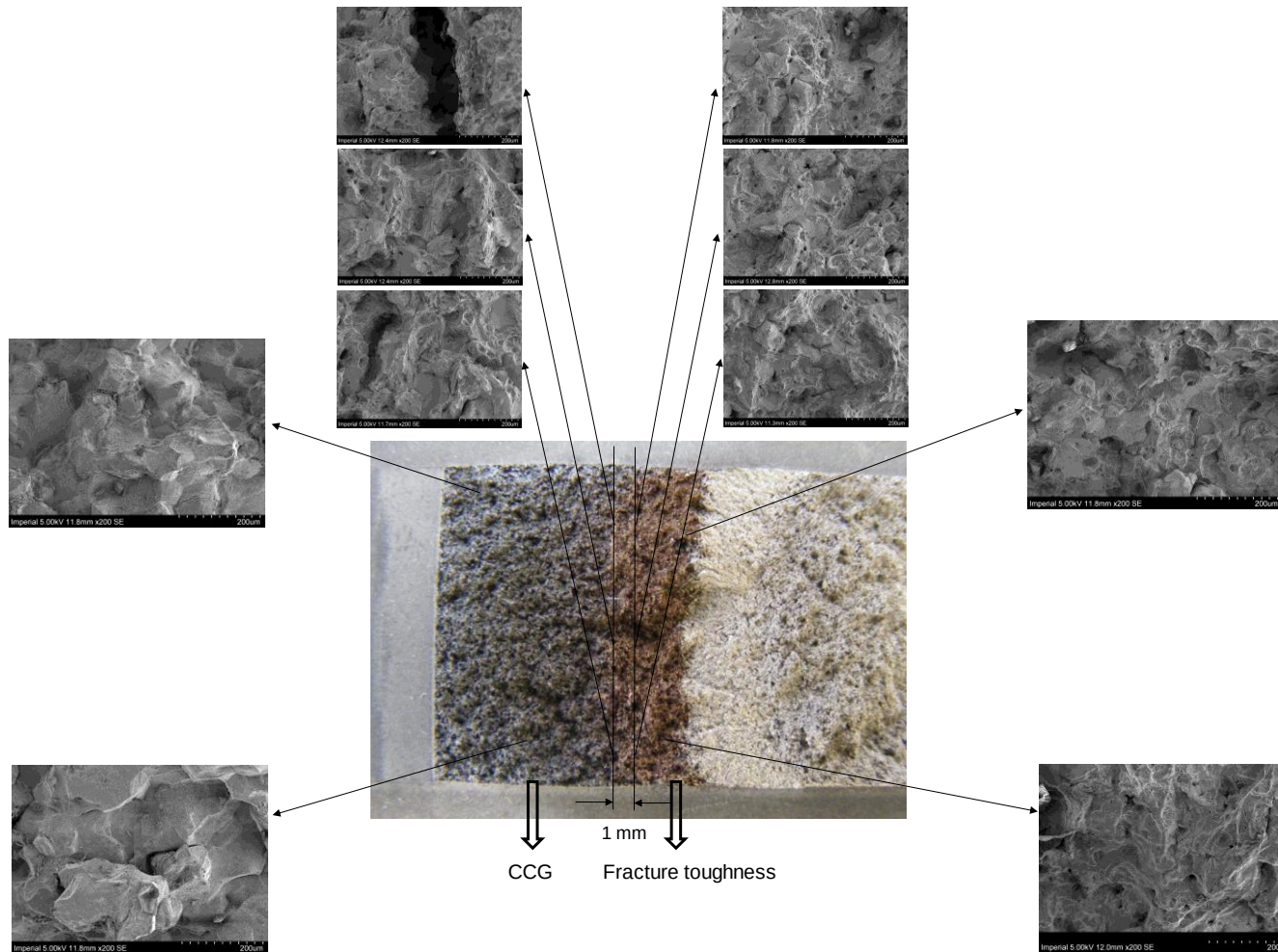


Figure 9.21: SEM study of the failure mechanisms on the fracture surface of the LCD2 specimen

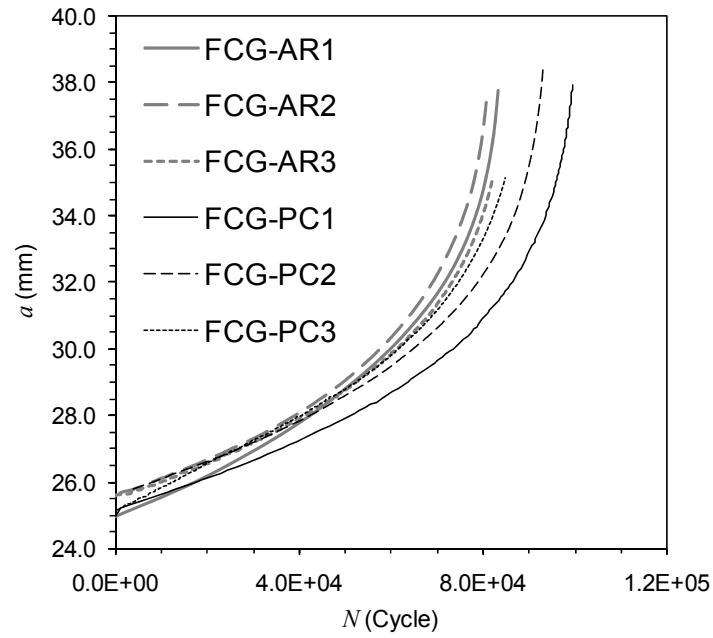


Figure 9.22: Variation of crack length against number of cycles in FCG tests on the AR and PC materials

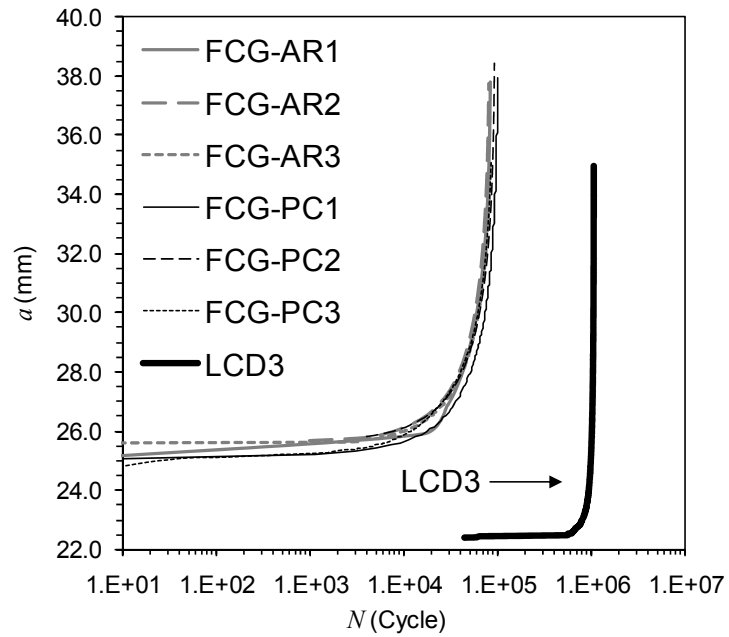


Figure 9.23: Comparison of the crack length vs. number of cycles trends in LCD3 specimen to the AR and PC materials

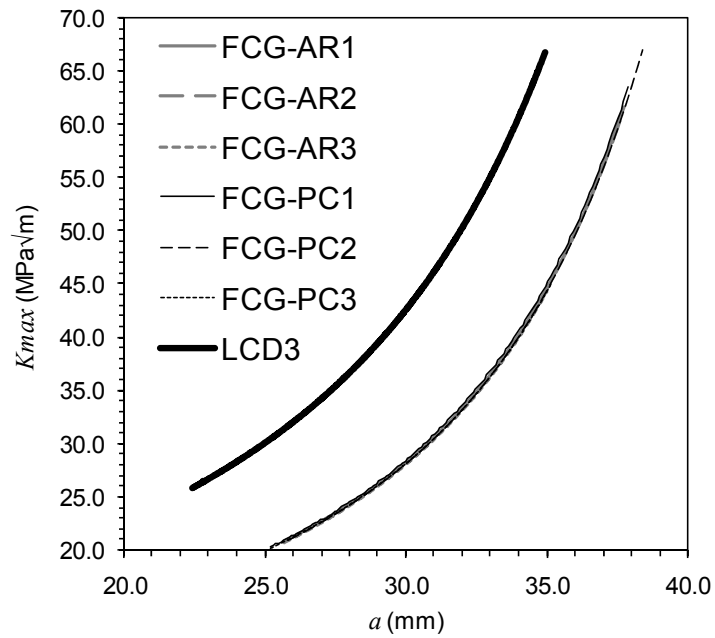
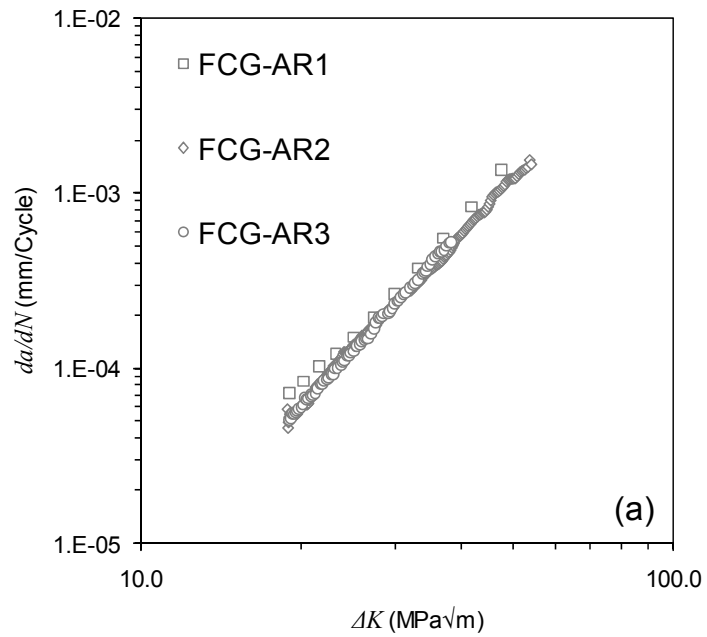


Figure 9.24: Comparison of the maximum stress intensity factor K_{max} vs. crack length trends in LCD3 specimen to the AR and PC materials



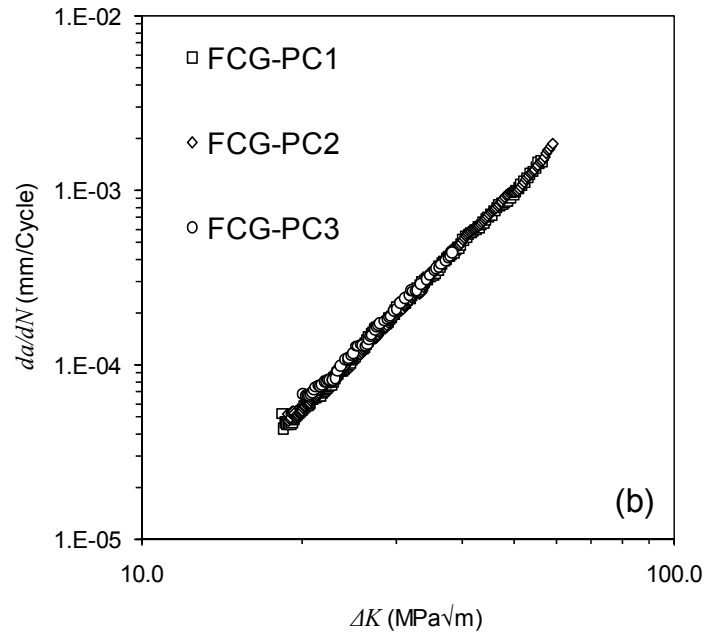


Figure 9.25: Correlation of the crack growth rate per cycle with the stress intensity factor range for the FCG tests performed on the (a) AR and (b) PC specimens

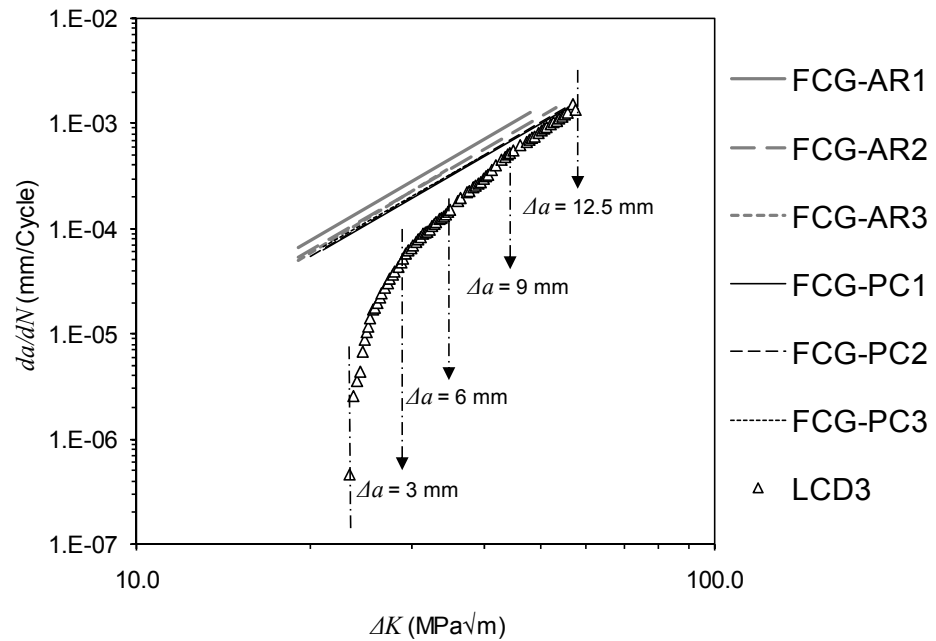


Figure 9.26: Comparison of the FCG trends in AR and PC materials to the LCD3 specimen

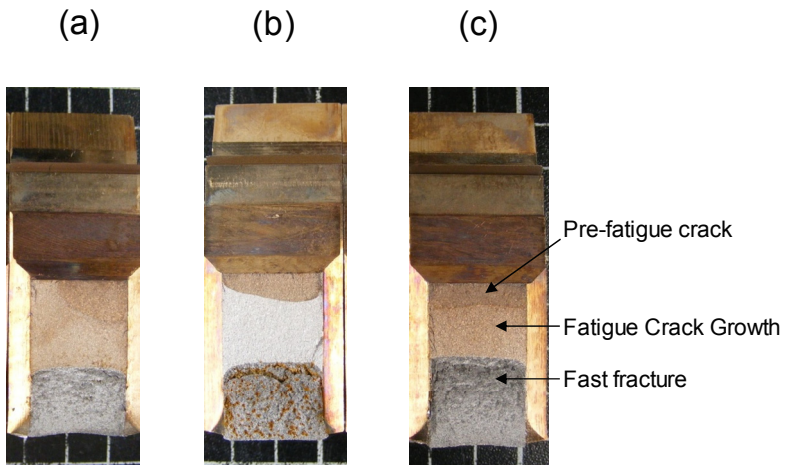


Figure 9.27: Fracture surfaces of (a) FCG-AR1, (b) FCG-AR2 and (c) FCG-AR3

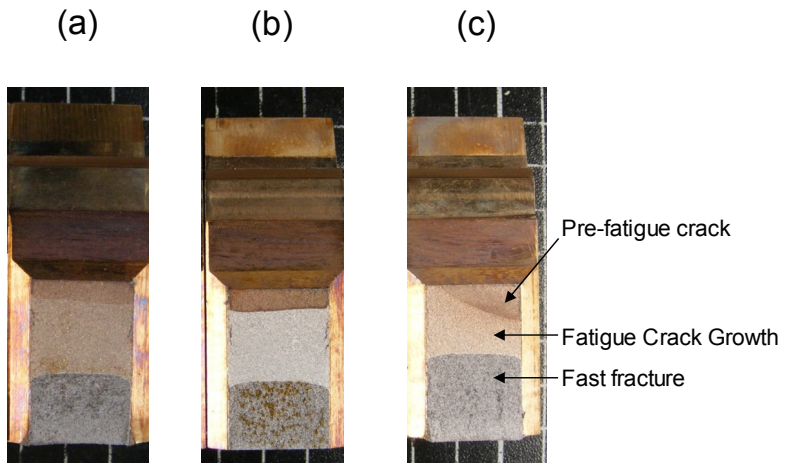


Figure 9.28: Fracture surfaces of (a) FCG-PC1, (b) FCG-PC2 and (c) FCG-PC3

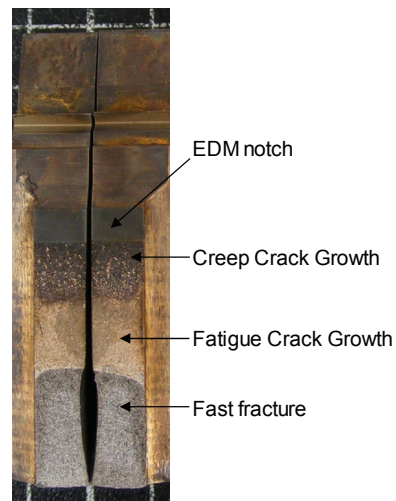


Figure 9.29: Fracture surface of LCD3 specimen

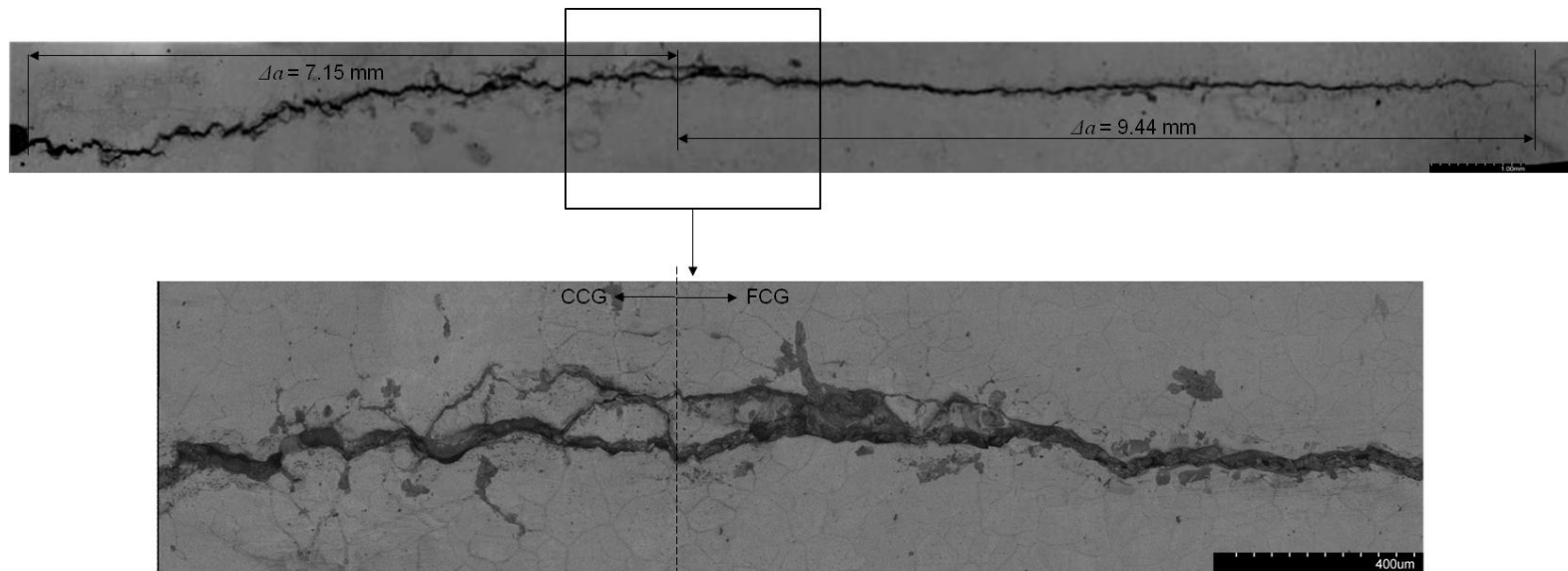


Figure 9.30: The intergranular creep and transgranular fatigue cracking mode in LCD3 specimen

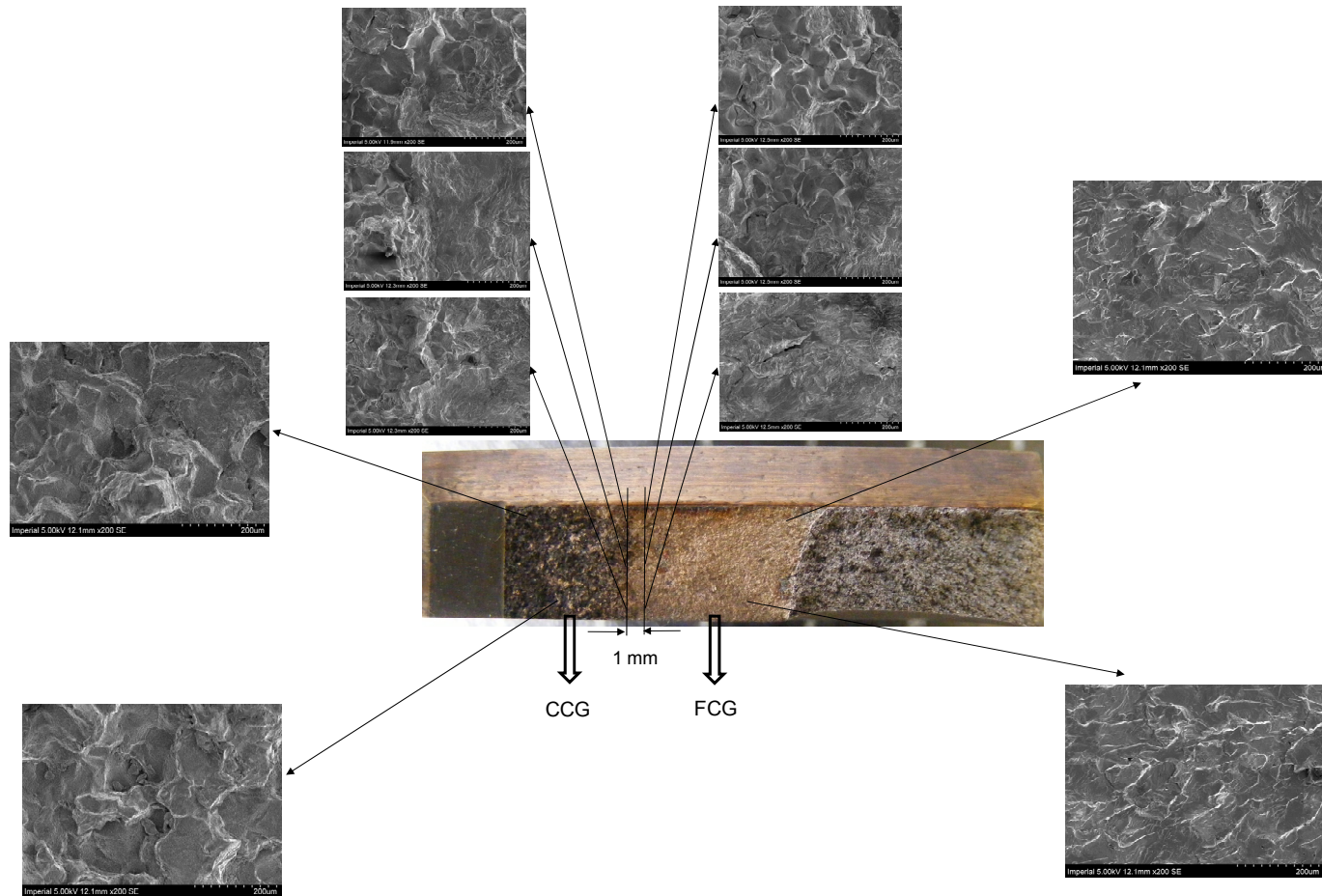


Figure 9.31: SEM study of the failure mechanisms on the fracture surface of the LCD3 specimen

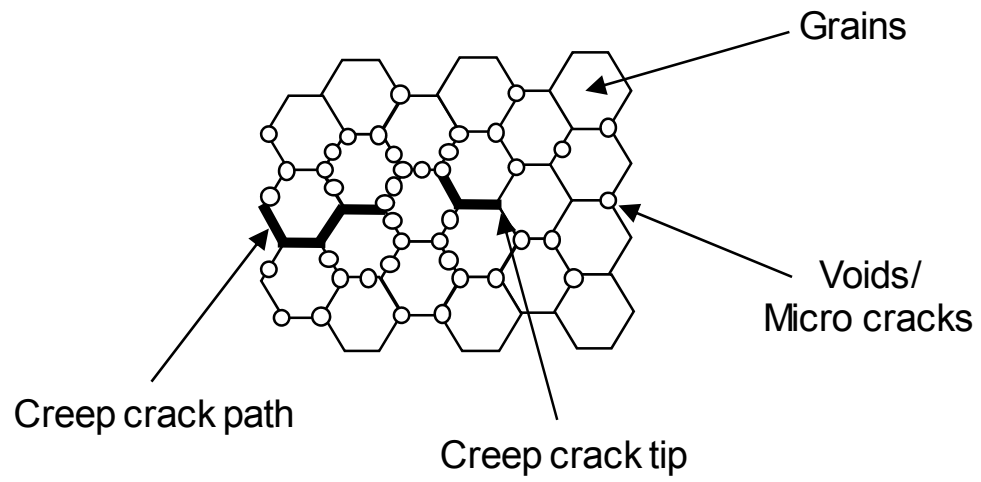


Figure 9.32: Schematic illustration of the intergranular creep damage formation behind and ahead of the final crack tip

Chapter 10

Conclusions and Future Work

10.1 Conclusions

The influence of service exposure, thus creep damage, on subsequent fatigue and fracture toughness behaviour of Type 316H stainless steel has been examined. Creep damage can be introduced into the material by interrupting CCG tests where creep damage will be formed ahead of the crack tip (local creep damage approach) or by globally uniformly creeping a block of material at elevated temperatures (global creep damage approach). To complete this research in the allocated timescales, the local creep damage method has been employed. Creep damage local to the crack tip has been introduced into the material by performing CCG tests on C(T) specimens with the initial normalised crack length of $a_0/W = 0.35$ and interrupting them at a/W of around 0.5. All creep tests were performed at 550 °C, which is the typical operating temperature for this material.

The yield stress of 316H at 550 °C is relatively low ($\sigma_y \approx 170$ MPa). Therefore, some significant plasticity is expected to form at the crack tip in CCG tests. In order to study the influence of creep damage alone on the fatigue and fracture toughness behaviour of the material, blocks of 316H taken from an ex-service steam header were pre-compressed to 8% plastic strain at room temperature to harden the material and limit the plasticity effects during CCG tests. The influence of pre-compression on the tensile and uniaxial creep behaviour of the material has been examined. The comparison of the tensile curves from the AR and PC materials shows that the pre-compression processes increases the yield stress of the material at both room temperature and 550 °C ($\sigma_y \approx 260$ MPa for PC material at 550 °C). As a result of pre-compression and thus hardening effects, insignificant plasticity at the crack tip and also a linear load up response has been observed in CCG tests on PC material.

Pre-compression has been found to have a significant influence on the creep properties of the material. For a given stress, a substantial drop in the rupture time and creep ductility has been observed in the PC material compared to those of attained from AR material. The minimum creep strain rate of the material has been found relatively insensitive to the pre-compression process, however the time to rupture of on

average around 3 times faster has been observed in PC material compared to the AR. The SEM studies on the AR and PC materials shows that the process of pre-compression to 8% plastic strain at room temperature introduces intergranular voids and damage on the grain boundaries. The influence of the intergranular damage induced by pre-compression becomes more significant when the material is tested under creep conditions at elevated temperatures and leads to lower values of creep ductility.

A model has been produced to estimate the creep ductility and strain rate as a function of the plastic strain levels in the material. Following this model, the creep ductility trend for 316H at 550 °C has been estimated using the data available at higher temperatures. The comparison of the creep strain at failure trends for the AR and PC materials shows that the pre-compression process may provide an estimate of the creep ductility of the AR material in the intermediate stress ranges when the data are plotted against the normalised stress in log-log axes. This suggests that pre-compression can be used to estimate the creep ductility of the AR material in much shorter time scales, especially at intermediate and low stress ranges where creep tests on AR material take considerably long time to complete.

In order to investigate the pre-compression effects on the CCG behaviour of the material, and also to obtain PD calibration functions for interrupting LCD tests at the desired crack length, tests have been performed on C(T) specimens made of PC material and the results are compared to the short term and long term test data on AR material. To calibrate the CCG behaviour of the PC material for the purpose of test interruption in LCD specimens, tests have been performed on specimens nominally identical to LCD samples with $a_0/W = 0.35$. The solutions for the shape function $Y(a/W)$ and the η factor available in the literature are valid for specimens with $a/W > 0.45$. Therefore FE simulations were performed to find out the accurate solutions of these parameters for the crack lengths in the range of $0.35 \leq a/W \leq 0.45$. By employing appropriate values of the geometry dependent fracture parameters in the data analysis, similar da/dt vs. C^* trends were found in PC specimens with $a_0/W = 0.35$ and 0.5.

For a given value of C^* , the CCG rate in the PC material is found around an order of magnitude larger than those of short term tests on AR material (due to lower creep ductility in the PC material and also limited crack tip plasticity which leads to higher constraint levels in the specimen) and the data follow the trend of upper bound long term tests on AR material. Similar cracking mode (brittle intergranular) has also been observed in the CCG tests on PC material and the long term tests on AR material. This

suggests that the pre-compression process may be used to generate a model material to study the long term CCG behaviour of the AR material at 550 °C in considerably shorter timescales.

NSW models have been employed to predict the CCI and CCG behaviour of the PC material. The comparison of the prediction lines and the available test data shows that acceptable predictions can be made using NSW models when the constant average creep ductility, based on ROA, is employed in calculations. In addition to NSW models, FE simulations have also been performed to investigate the CCG behaviour of the AR and PC materials at 550 °C. Creep damage has been evaluated using the ductility exhaustion approach where the multiaxial strain factor is calculated using Cocks & Ashby void growth model. The stress dependent creep ductility trends estimated for 316H at 550 °C have been implemented in FE simulations, assuming that the PC material follows the same trend as the AR material when the creep strain at failure is plotted against the applied stress normalised by the yield stress. It has been shown that this model successfully predicts the short term and long term CCG behaviour of the AR and PC materials and more importantly, the transition between high load to load tests on the AR material, which appears as a shift in the da/dt vs. C^* trends, can be modelled by these numerical simulations.

Creep damage local to the crack tip has been introduced into C(T) specimens made of PC material by starting CCG tests at the normalised crack length of $a_0/W = 0.35$ and interrupting them at $a/W \approx 0.5$ where the PD rates are about those of associated with $a = 25$ mm in CCG calibration tests. The ND measurements show that as a result of unloading process in the CCG tests, compressive residual stress of around -100 MPa is expected at the crack tip in a direction parallel to the loading axis. Therefore, the influence of residual stresses has been considered when interpreting the subsequent FCG and fracture toughness test data on LCD specimens.

FCG and fracture toughness tests have been performed on the AR, PC and LCD specimens to firstly examine the pre-compression effects and secondly the local creep damage effects on the material fracture behaviour. To avoid ambiguity in the data, all FCG and fracture toughness tests were conducted at room temperature. An estimate of the instantaneous crack length has been provided using the unloading compliance measurements in all FCG and fracture toughness tests examined.

It has been observed that the fracture toughness value of the material drops as a result of the pre-compression process. Due to the inaccuracies encountered in the unloading compliance measurements in the presence of voids/micro cracks and

unbroken ligaments around the crack tip (thus discontinuous crack) in LCD specimens, the R-curves generated for the LCD specimens were modified by removing the tail regions prior to fracture toughness data analysis. Smaller values of fracture energy have been found in LCD specimens compared to the PC material when the modified R-curves are considered in the analyses. A decrease in the fracture energy of the LCD material is not only observed at small values of crack extension but a considerable drop has also been observed away from the initial crack tip which may be related to the material's failure strain decrease as a result of creep deformation and damage introduced into the specimens. An alternative approach employed to re-generate R-curves for LCD specimens by correcting the initial and final crack length estimates using the values measured on the fracture surface of the specimens showed that a larger reduction in the fracture toughness of the LCD specimens is observed when the tail regions are considered in the fracture toughness analysis. The influence of compressive residual stress on the fracture toughness behaviour of LCD specimens may be considered insignificant, compared to creep damage effects, since the resistant compressive residual stress field is relatively small at the crack front region (around 1 mm) and will be washed out at higher loads.

The FCG results from the room temperature tests performed on AR and PC materials at $R = 0.1$ and $f = 10$ Hz show that the $daldN$ vs. ΔK trends are insensitive to the pre-compression process. It has been observed that the crack initiation is severely delayed in the LCD specimen and the early stage fatigue crack growth rate per cycle is smaller than those of obtained from the AR and PC materials, for a given value of ΔK . Further seen in the FCG behaviour of the LCD specimen is a convergence with the $daldN$ vs. ΔK trends obtained from the AR and PC materials, at larger values of stress intensity factor range which correspond to the regions away from the creep damage zone. The experimental results show that local creep damage has substantial effects on the initiation and early stage FCG behaviour of the material due to a number of reasons such as formation of micro cracks/unbroken ligaments around the crack tip, discontinuity of the crack path in the creep crack tip region, change in the cracking mode from intergranular to transgranular and thus interaction between these failure mechanisms in the creep damage zone, an increase in the yield stress of the material due to creep deformation and finally the compressive residual stress effects at the crack tip which may result in effective R -ratio of smaller than 0.1.

10.2 Future Work

As discussed in Chapter 9, introducing creep damage local to the crack tip leads to formation of micro cracks/voids around the crack tip, appearance of compressive residual stresses at the crack tip and possible discontinuity of the creep crack path in the vicinity of the final creep crack tip. In addition, the change in the cracking mode from intergranular creep to a transgranular mode will influence the subsequent results especially the initiation and early stage FCG. To avoid these inevitable consequences of the LCD approach, creep damage can be globally uniformly introduced into the material.

The first step in the employment of the GCD method is to interrupt uniaxial creep tests on round bar specimens, made of AR and PC materials, at different parts of the creep curve in which various extent of creep damage is expected to exist in the material. These specimens are suggested to be turned down to remove the necking effects and subsequently perform tensile tests at room temperature and 550 °C. Therefore, the global creep damage effects on the tensile behaviour of the material can be examined. Note that the creep deformation and damage in 316H is very sensitive to plasticity. Therefore, to examine creep damage effects only on the tensile response of the material, the stresses lower than the yield stress of the material need to be applied in these tests.

As the second step in GCD testing, it is suggested to perform uniaxial creep tests on large blocks of AR and PC materials. Subsequent to interruption of GCD tests on large blocks of AR and PC materials at different stages of the creep curve, uniaxial and C(T) specimens can be manufactured from the globally creep damaged material in the gauge region to perform tensile, uniaxial creep, CCG, FCG and fracture toughness tests. This way, the creep damage effects on the subsequent fracture behaviour of the material can be examined in the absence of pre-existing cracks, micro cracks and residual stresses. The GCD approach also provides an opportunity to investigate the creep damage effects on deformation behaviour of the uncracked bodies under cyclic loading conditions (*i.e.* high cycle fatigue and low cycle fatigue tests on GCD material).

A significant part of this project was allocated to the investigation of the pre-compression effects on the tensile, creep, fatigue and fracture toughness behaviour of the material. A better understanding of the pre-straining effects can be made by performing in-situ ND tensile tests on AR and PC materials at room/high temperature to study the lattice deformation along different crystallographic planes. Furthermore, fracture toughness, FCG and CCG tests are suggested to be performed on plane

specimens made of AR and PC materials and the local deformation around the crack tip to be measured using the digital image correlation (DIC) equipments. The strain fields obtained by DIC measurements can also be compared to those of predicted by FE simulations.

In order to provide a more confident conclusion about the local creep damage effects on room temperature fatigue and fracture toughness behaviour of 316H, more tests need to be performed on LCD specimens. As discussed in Chapter 9, the FCG test on LCD4 specimen is yet to be done and the results from this test will help examining the repeatability of the observed trends in the test on LCD3 specimen. In addition to experimental testing, advanced FE simulation techniques can be applied to model the intergranular creep damage followed by fatigue and fracture toughness modelling.

As shown and discussed in Chapter 9, the instantaneous crack length estimates using the unloading compliance technique are substantially influenced by the micro cracks/unbroken ligaments around the crack tip in fracture toughness and FCG tests on LCD specimens. It is suggested to monitor the crack growth in future fracture toughness and FCG tests on LCD samples using the PD technique in conjunction with the unloading compliance. This way the accuracy of the PD method in crack length measurements for LCD specimens and also its sensitivity to the link up of micro cracks and discontinuous crack path at the creep crack tip can be examined.

All FCG and fracture toughness tests in this project were performed at room temperature to avoid uncertainties in the measurements. However, the influence of local and also global creep damage, introduced into the material at 550°C, on the subsequent FCG and fracture toughness behaviour of the material can be investigated at 470 °C in future tests.

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Appendices

Appendix A	Test Matrices
Appendix B	Eta Solutions for $0.35 \leq a/W \leq 0.45$
Appendix C	CCG and CCI NSW Predictions for PC Material using Axial Creep Ductility

Appendix A

Test Matrices

A.1 Introduction

The test matrices detailing the specimen geometries, testing temperature and material's pre-straining condition are supplied in this Appendix. The tests performed on the AR and PC material are summarised in Table A.1 and Table A.2, respectively. Furthermore, the specimen details for the fracture toughness and fatigue crack growth tests conducted on the local creep damage (LCD) material are provided in Table A.3.

Table A.1: Test specimen matrix for AR material

Test Type*	Test ID	Pre-Straining Condition	Specimen Geometry	Temperature (°C)	d (mm)	W (mm)	a_0/W
T	RT-A1-1	Uncompressed	Tensile specimen	20	8	N/A	N/A
T	RT-A1-2	Uncompressed	Tensile specimen	20	8	N/A	N/A
T	HT-A1-1	Uncompressed	Tensile specimen	550	8	N/A	N/A
T	HT-A1-2	Uncompressed	Tensile specimen	550	8	N/A	N/A
UC	A1-1	Uncompressed	Uniaxial creep specimen	550	8	N/A	N/A
UC	A1-2	Uncompressed	Uniaxial creep specimen	550	8	N/A	N/A
UC	A1-3	Uncompressed	Uniaxial creep specimen	550	8	N/A	N/A
UC	A1-4	Uncompressed	Uniaxial creep specimen	550	8	N/A	N/A
UC	A1-5	Uncompressed	Uniaxial creep specimen	550	8	N/A	N/A
JIC	JIC-AR1	Uncompressed	C(T)	20	N/A	50	0.5
JIC	JIC-AR2	Uncompressed	C(T)	20	N/A	50	0.5
FCG	FCG-AR1	Uncompressed	C(T)	20	N/A	50	0.5
FCG	FCG-AR2	Uncompressed	C(T)	20	N/A	50	0.5
FCG	FCG-AR3	Uncompressed	C(T)	20	N/A	50	0.5

T = Tensile

UC = Uniaxial Creep

JIC = Fracture Toughness

FCG = Fatigue Crack Growth

Table A.2: Test specimen matrix for PC material

Test Type*	Test ID	Pre-Straining Condition	Specimen Geometry	Temperature (°C)	d (mm)	W (mm)	a_0/W
T	RT-A2-1	8% Pre-compressed	Tensile specimen	20	8	N/A	N/A
T	RT-A2-2	8% Pre-compressed	Tensile specimen	20	8	N/A	N/A
T	HT-A2-1	8% Pre-compressed	Tensile specimen	550	8	N/A	N/A
T	HT-A2-2	8% Pre-compressed	Tensile specimen	550	8	N/A	N/A
UC	A2-1	8% Pre-compressed	Uniaxial creep specimen	550	8	N/A	N/A
UC	A2-2	8% Pre-compressed	Uniaxial creep specimen	550	8	N/A	N/A
UC	A2-3	8% Pre-compressed	Uniaxial creep specimen	550	8	N/A	N/A
UC	A2-4	8% Pre-compressed	Uniaxial creep specimen	550	8	N/A	N/A
JIC	JIC-PC1	8% Pre-compressed	C(T)	20	N/A	50	0.5
JIC	JIC-PC2	8% Pre-compressed	C(T)	20	N/A	50	0.5
FCG	FCG-PC1	8% Pre-compressed	C(T)	20	N/A	50	0.5
FCG	FCG-PC2	8% Pre-compressed	C(T)	20	N/A	50	0.5
FCG	FCG-PC3	8% Pre-compressed	C(T)	20	N/A	50	0.5
CCG	8PC-A1	8% Pre-compressed	C(T)	550	N/A	50	0.35
CCG	8PC-A2	8% Pre-compressed	C(T)	550	N/A	50	0.35
CCG	8PC-A3	8% Pre-compressed	C(T)	550	N/A	50	0.35
CCG	8PC-A4	8% Pre-compressed	C(T)	550	N/A	50	0.5

CCG = Creep Crack Growth

Table A.3: Test specimen matrix for local creep damage material

Test Type	Test ID	Pre-Straining Condition	Specimen Geometry	Temperature (°C)	W (mm)	a_0/W
JIC	LCD1	8% Pre-compressed	C(T)	20	50	≈ 0.5
JIC	LCD2	8% Pre-compressed	C(T)	20	50	≈ 0.5
FCG	LCD3	8% Pre-compressed	C(T)	20	50	≈ 0.5
FCG	LCD4	8% Pre-compressed	C(T)	20	50	≈ 0.5

Appendix B

Eta Solutions for $0.35 \leq a/W \leq 0.45$

B.1 Comparison of Eta Parameter Calculated Using H and A_p

The η parameter can be calculated using the H function and the plastic area under the load *vs.* displacement curve, A_p . It has been shown in [1] that the H parameter is derived based on the plastic area under the curve, therefore the solutions calculated using both approaches are expected to be identical. Comparison of the η solutions from both methods has confirmed this equality. An example is shown in Table B.1 for $a/W = 0.35$. Note that for a C(T) specimen $H = N(N+1)$ and A_p has been calculated using trapezoidal method. Also observed in Table B.1 is that under both plane stress (PS) and plane strain (PE) conditions, $\eta^{LLD} = \eta^{CMOD}$ for large values of N (*i.e.* $N = 10$ and 20) at this crack length, but some differences between LLD and CMOD solutions are observed for smaller values of N (*i.e.* $N = 3, 5$ and 6)

Table B.1: Comparison of Eta solutions evaluated using A_p and H at $a/W = 0.35$ for a range of N values

$a/W = 0.35$		η_{Ap}	η_H
η^{LLD} -PS	N = 3	2.17	2.17
	N = 5	2.10	2.10
	N = 6	2.10	2.10
	N = 10	2.14	2.14
	N = 20	2.21	2.21
η^{LLD} -PE	N = 3	2.03	2.03
	N = 5	1.85	1.85
	N = 6	1.81	1.81
	N = 10	1.75	1.75
	N = 20	1.70	1.70
η^{CMOD} -PS	N = 3	2.21	2.21
	N = 5	2.12	2.12
	N = 6	2.11	2.11
	N = 10	2.14	2.14
	N = 20	2.21	2.21
η^{CMOD} -PE	N = 3	2.07	2.07
	N = 5	1.87	1.87
	N = 6	1.82	1.82
	N = 10	1.75	1.75
	N = 20	1.70	1.70

B.2 N Dependency of Eta Solution for $0.35 < a/W < 0.45$

The dependency of η solution on the power law stress exponent, N , stress state conditions and LLD/CMOD has been examined at a few different crack lengths in the range of $0.35 \leq a/W \leq 0.45$ and the results are shown in Figure B.1–Figure B.5. It can be seen that for shallow cracks (*i.e.* $a/W = 0.35$) under both PS and PE conditions the biggest and the smallest differences between η^{LLD} and η^{CMOD} are observed at $N = 3$ and $N = 20$, respectively. However for larger cracks (*i.e.* $a/W = 0.45$), the dependency of the Eta solution on LLD/CMOD displacement is almost eliminated and also the values calculated under PS and PE conditions are found very similar. Furthermore, the η solutions at larger crack lengths, calculated under different stress state conditions using both LLD and CMOD are found consistent with that of recommended in ASTM E1457 [2] for $a/W > 0.45$ (*i.e.* $\eta = 2.2 \pm 0.1$).

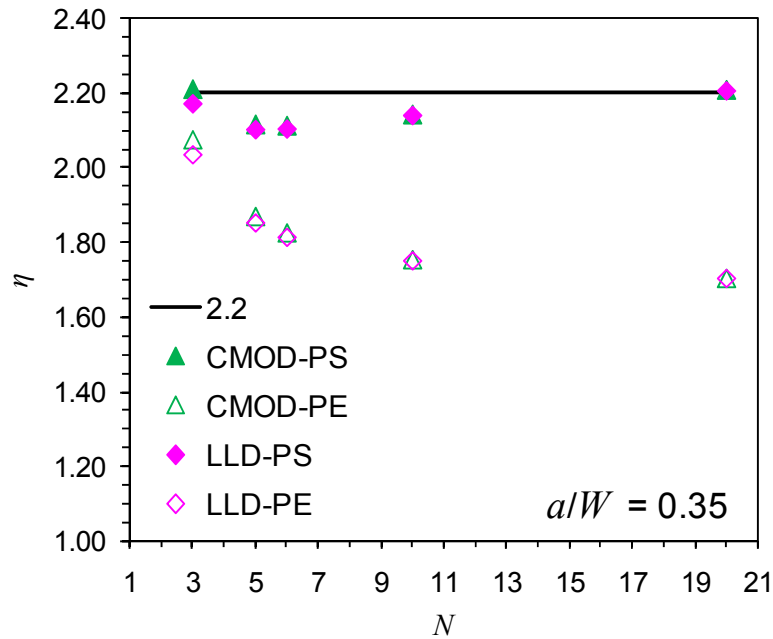


Figure B.1: Dependency of Eta solution on N , stress state conditions and LLD/CMOD at $a/W = 0.35$

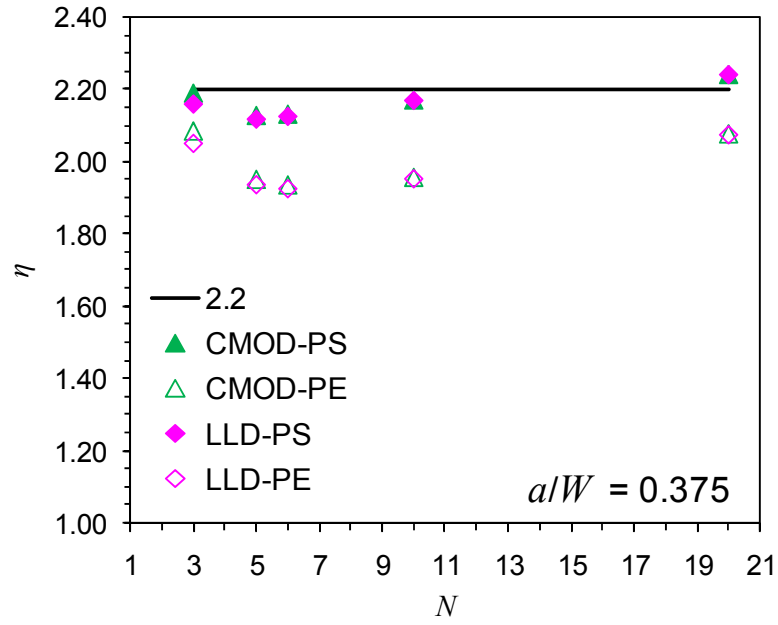


Figure B.2: Dependency of Eta solution on N , stress state conditions and LLD/CMOD
at $a/W = 0.375$

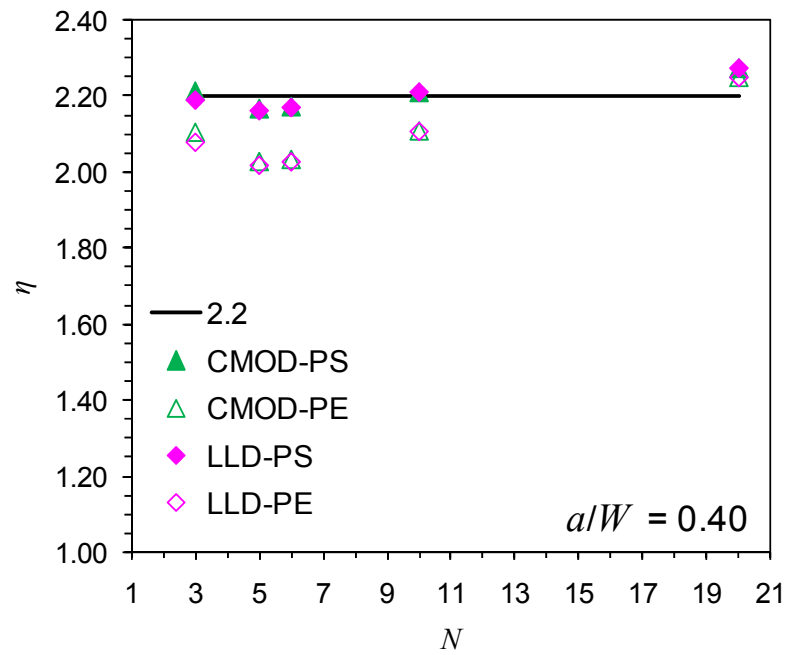


Figure B.3: Dependency of Eta solution on N , stress state conditions and LLD/CMOD
at $a/W = 0.40$

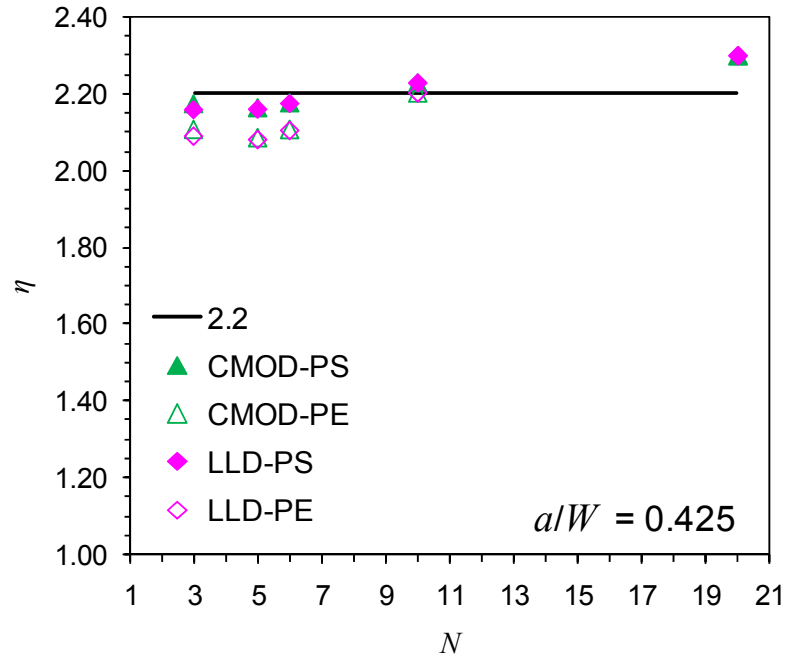


Figure B.4: Dependency of Eta solution on N , stress state conditions and LLD/CMOD at $a/W = 0.425$

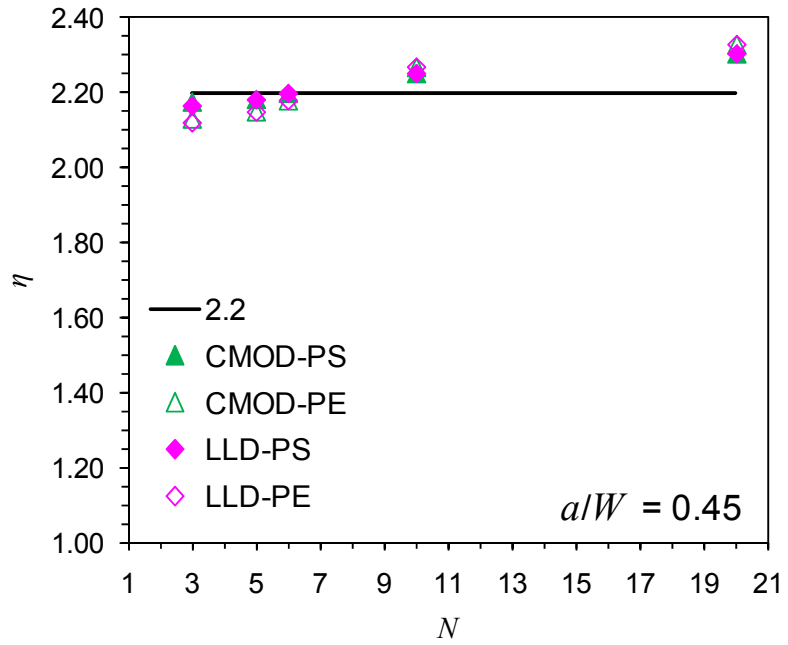


Figure B.5: Dependency of Eta solution on N , stress state conditions and LLD/CMOD at $a/W = 0.45$

B.3 η Solutions for $0.35 \leq a/W \leq 0.7$

The η solutions evaluated for $0.35 \leq a/W \leq 0.45$ under PS and PE conditions using LLD and CMOD are compared to those of presented in [1] for $0.45 \leq a/W \leq 0.7$ and also the general solution suggested in ASTM E1457 standard [2]. As seen in Figure B.6, the standard solution provided in ASTM E1457 ($\eta = 2.2 \pm 0.1$) has relatively good agreement with the numerical values of η evaluated for $0.35 \leq a/W \leq 0.45$ under plane stress conditions, however under plane strain conditions the solutions are different and a strong dependency on the normalised crack length a/W and the power law stress exponent N is observed. Further seen in Figure B.6 is that the numerical values of η for $0.35 < a/W < 0.45$ follow the trend of those values previously evaluated in [1] for larger crack lengths.

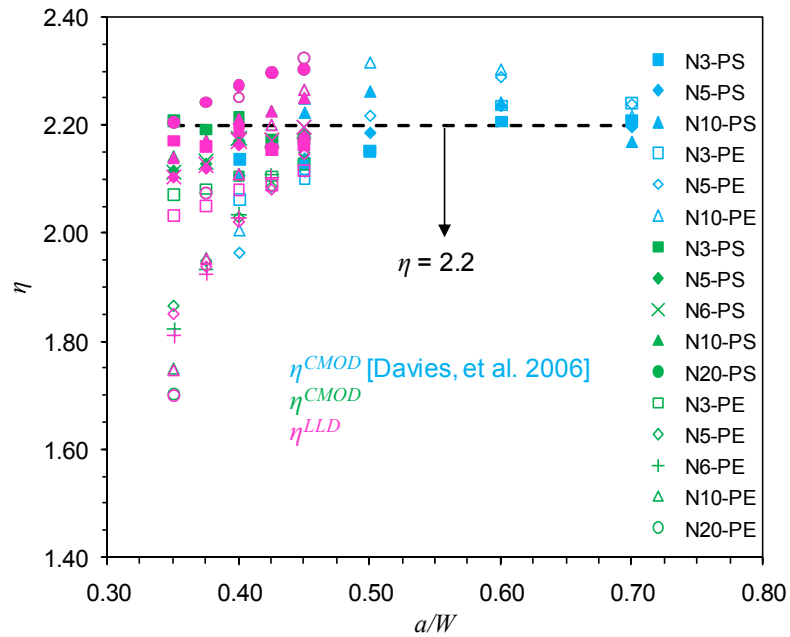


Figure B.6: Numerical solutions of Eta parameter evaluated under PS and PE conditions for a wide range of a/W and N using LLD and CMOD

B.4 Comparison to EPRI Solutions

The η solutions evaluated under PS and PE conditions for $0.35 \leq a/W \leq 0.45$ are compared to the EPRI solutions, which can be described for a fully plastic material as

$$\eta_{EPRI}^{LLD} = \frac{1}{H} \frac{b^2}{a} \frac{B}{P_0} \frac{g_1(a/W, N)}{g_3(a/W, N)} \frac{h_1(a/W, N)}{h_3(a/W, N)} \quad (B.1)$$

where $H = N(N+1)$ for C(T) specimens, B is the specimen thickness, P_0 is a normalising load, h_1 , h_3 , g_1 , and g_3 are dimensionless functions, values of which are tabulated in [3], and b is the uncracked ligament size ahead of the crack tip ($W-a$). This comparison is shown in Figure B.7. Also included in this figure are the solutions available in [1] for $0.45 \leq a/W \leq 0.7$ and the general solution provided in ASTM E1457 standard [2].

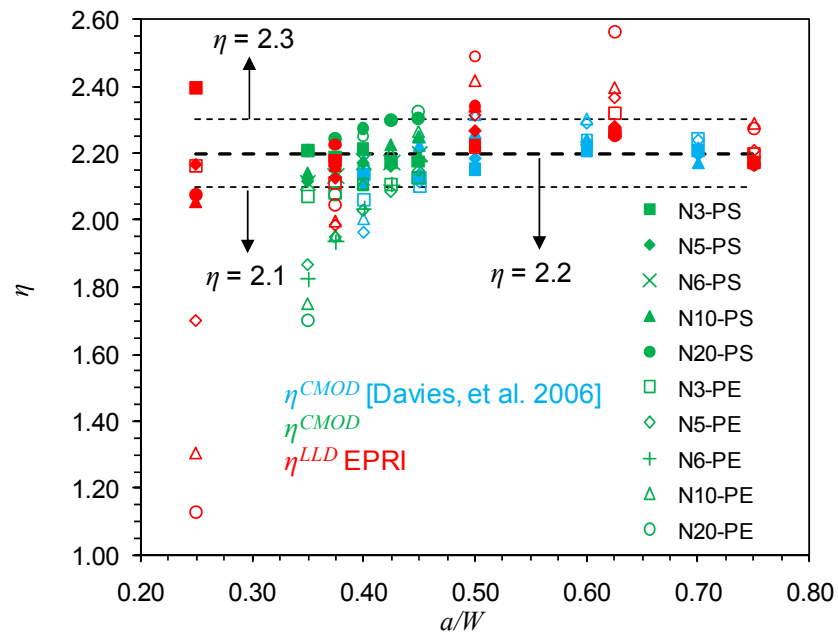


Figure B.7: Comparison of the numerical η values to the EPRI solutions evaluated for a range of a/W and N under plane stress and plane strain conditions

It can be seen in Figure B.7 that the observed trends in the numerical η solutions for $0.35 \leq a/W \leq 0.45$ are consistent with those of provided by EPRI solutions for $0.25 \leq a/W \leq 0.75$. The linear fit to the numerical plane strain solutions of the η factor for shallow cracks is compared to that of made to the EPRI data points for $N = 10$ in Figure B.8. As seen in this figure, the linear fits fall very close to each other. This implies that the numerical values of η evaluated under plane strain conditions for crack lengths in the range of $0.35 \leq a/W \leq 0.45$ are in good agreement with the values provided by the EPRI solution.

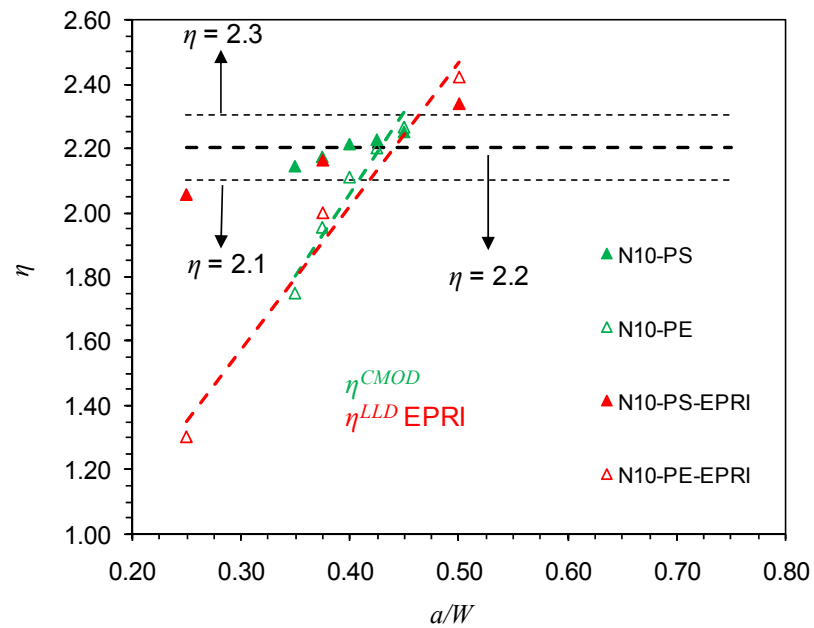


Figure B.8: Comparison of the numerical values of Eta evaluated for $N = 10$ to the EPRI solutions

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Appendix C

CCG and CCI Predictions using Axial Creep Ductility

C.1 CCG Predictions

The average of the axial creep ductilities for the PC material, under the range of stresses examined, is found around 5 times smaller than the mean creep strain at failure found using the reduction of area (ROA). Therefore, the creep crack growth (CCG) rates predicted by NSW models using axial creep ductility are expected to be approximately 5 times larger than those of calculated using ROA. The NSW, NSW-MOD and NSWA predictions using axial creep ductility and different values of the creep power law stress exponents ($n = 10.9$ and $n_A = 14$) are shown in the figures below and compared to the experimental data available on PC material. As seen in these figures, employment of the axial creep ductility in NSW models generally overestimates the CCG rate for a given value of C^* .

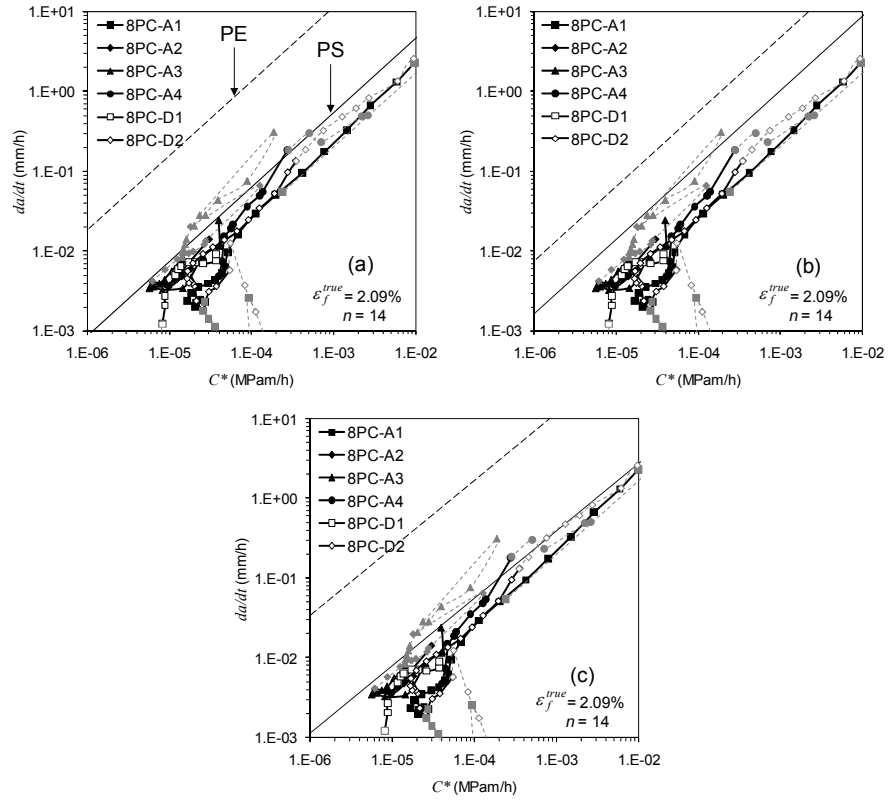
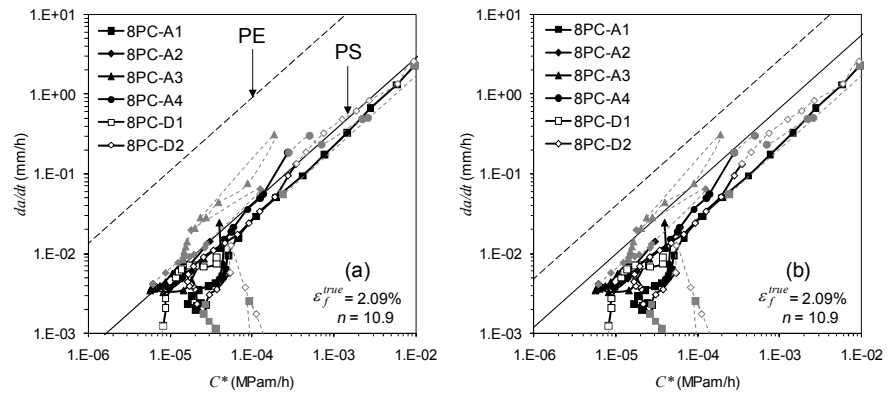


Figure C.1: CCG predictions using the average creep stress exponent and the axial creep ductility by (a) NSW (b) NSW-MOD (c) NSWa models



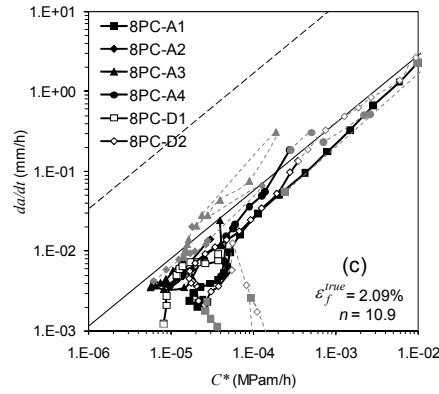


Figure C.2: CCG predictions using the minimum creep stress exponent and the axial creep ductility by (a) NSW (b) NSW-MOD (c) NSWA models

C.2 CCI Predictions

The creep crack initiation (CCI) trends were predicted using NSW models in Chapter 6 by employing the ROA and the average power law creep stress exponent ($n_A = 14$). Further predictions made by NSW, NSW-MOD and NSWA models using the axial creep ductility and different power law stress exponents are presented below for the CCI time corresponding to 0.2 and 0.5 mm crack extension. The sensitivity of the NSW CCI predictions to the power law stress exponent has been further examined by calculating the NSW prediction lines using ROA and the minimum creep stress exponent ($n = 10.9$).

C.2.1 $t_{0.2}$

The CCI trends corresponding to 0.2 mm crack extensivities, $t_{0.2}$, are predicted by NSW, NSW-MOD and NSWA models and the results are shown in the figures below. Figure C.3 shows that good predictions can be made by employing the ROA in the NSW models and the change in n has small effects on NSW and NSW-MOD predictions but doesn't influence NSWA (compared to the predictions made using ROA and n_A in Chapter 6.)

Figure C.4 and Figure C.5 show that using the axial creep ductility, relatively good predictions are made by NSW plane stress lines for the experimental CCI data points obtained from the tests performed on PC material at lower loads.

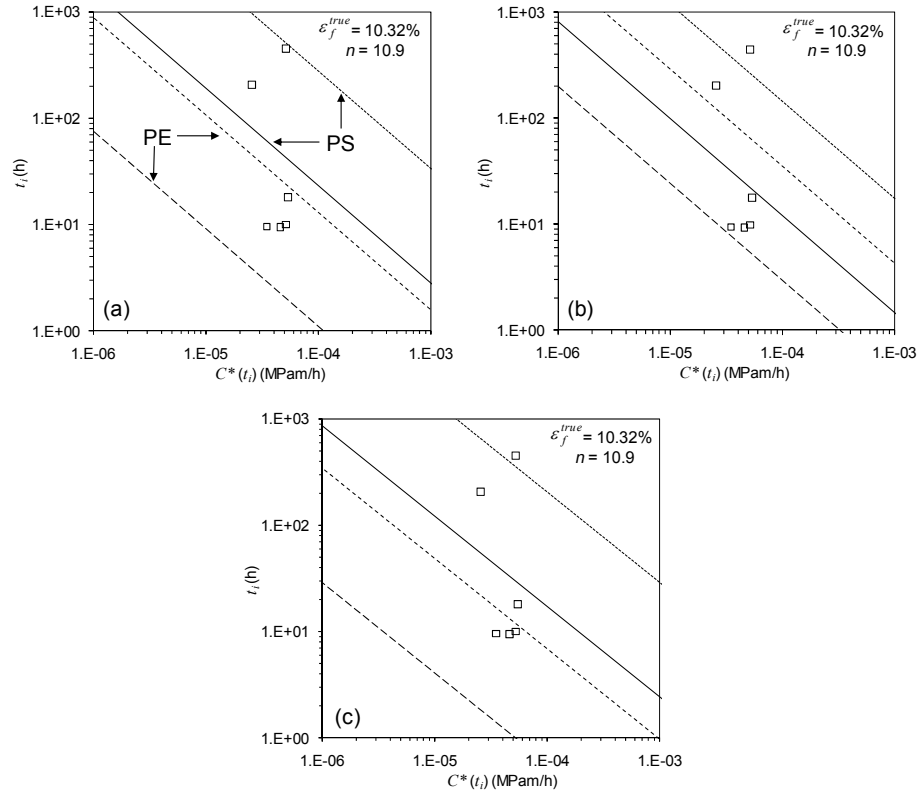
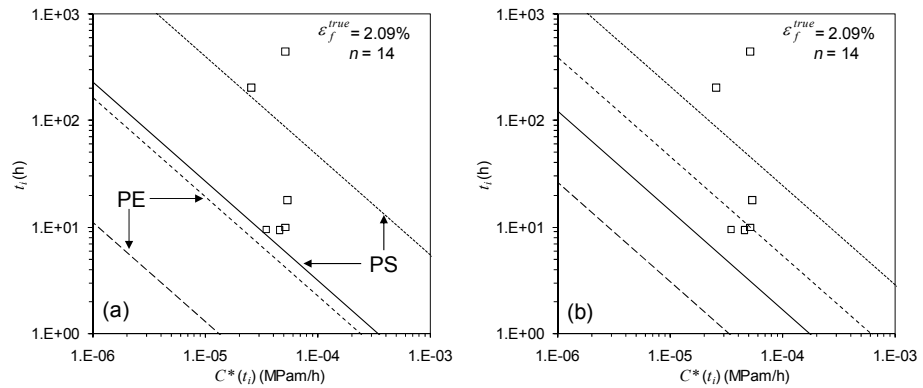


Figure C.3: Prediction of $t_{0.2}$ trends using the minimum creep properties and ROA by (a) NSW (b) NSW-MOD (c) NSW models



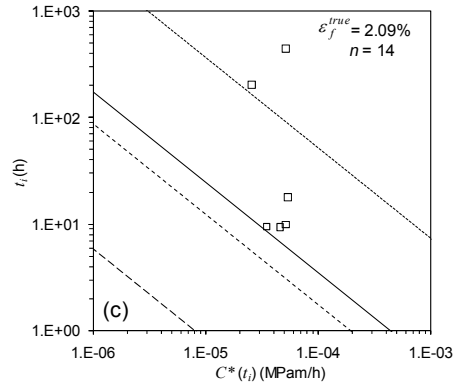


Figure C.4: Prediction of $t_{0.2}$ trends using the average creep stress exponent and axial creep ductility by (a) NSW (b) NSW-MOD (c) NSWA models

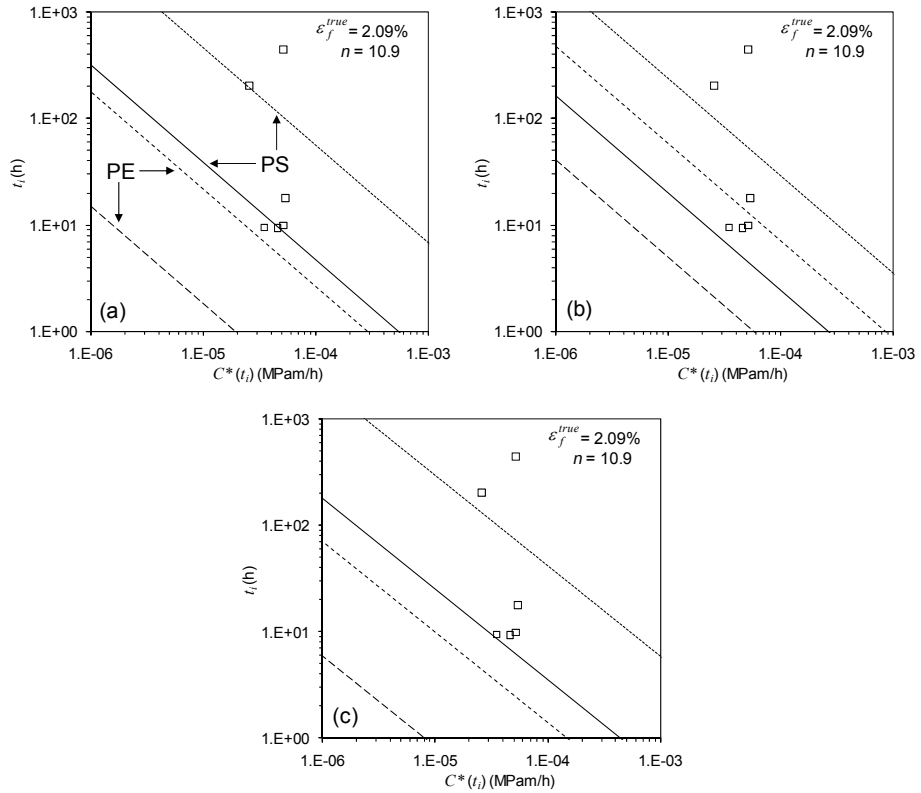


Figure C.5: Prediction of $t_{0.2}$ trends using the minimum creep stress exponent and axial creep ductility by (a) NSW (b) NSW-MOD (c) NSWA models

C.2.2 $t_{0.5}$

Similar to $t_{0.2}$, Figure C.6 shows that the CCI corresponding to 0.5 mm crack extension, $t_{0.5}$, can be successfully predicted by NSW, NSW-MOD and NSW models when the ROA and minimum creep stress exponent ($n = 10.9$) are employed in calculations. Furthermore, Figure C.7 and Figure C.8 show that acceptable predictions for the majority of the test data can be made by employing the axial creep ductility and different values of the power law stress exponents ($n = 10.9$ and $n_A = 14$) in NSW models calculated for PS conditions.

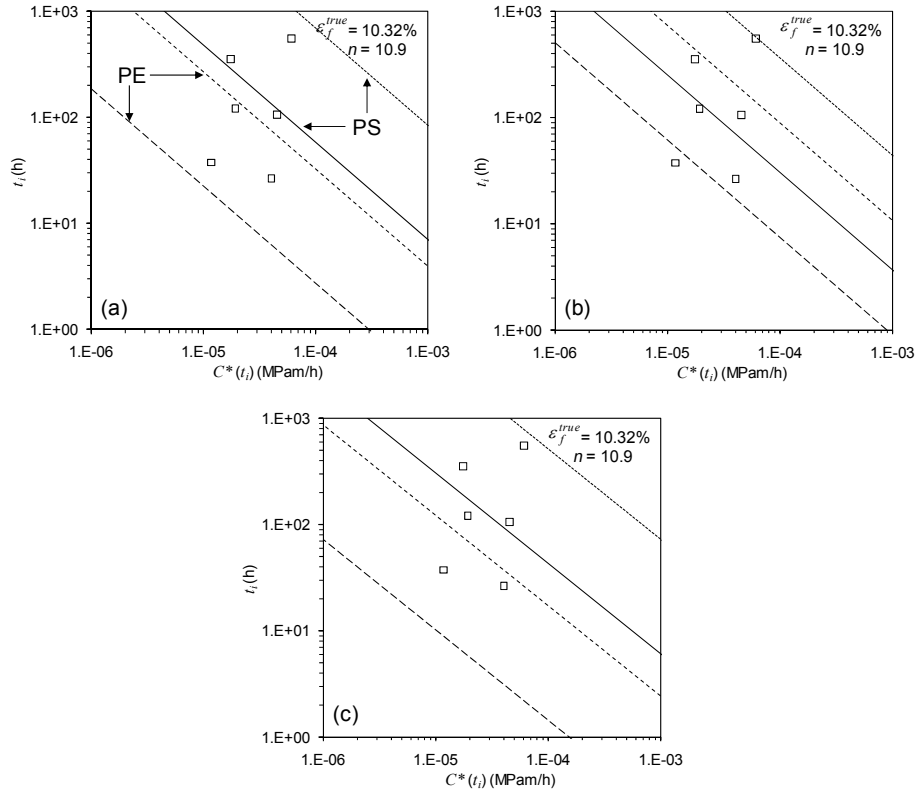


Figure C.6: Prediction of $t_{0.5}$ trends using the minimum creep properties and ROA by (a) NSW (b) NSW-MOD (c) NSW models

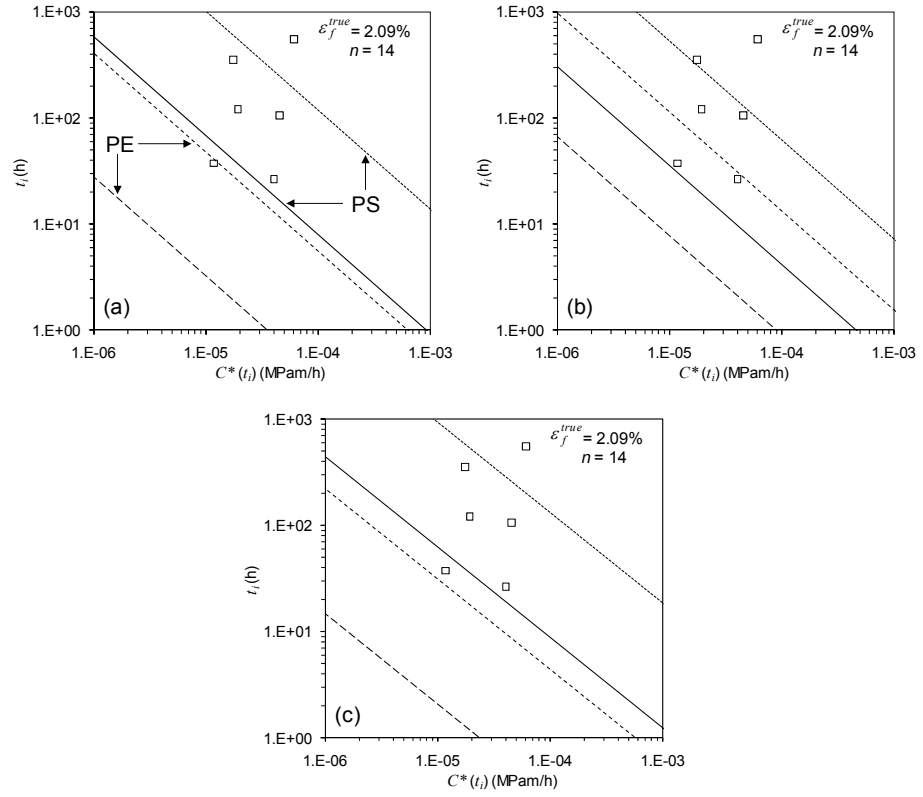
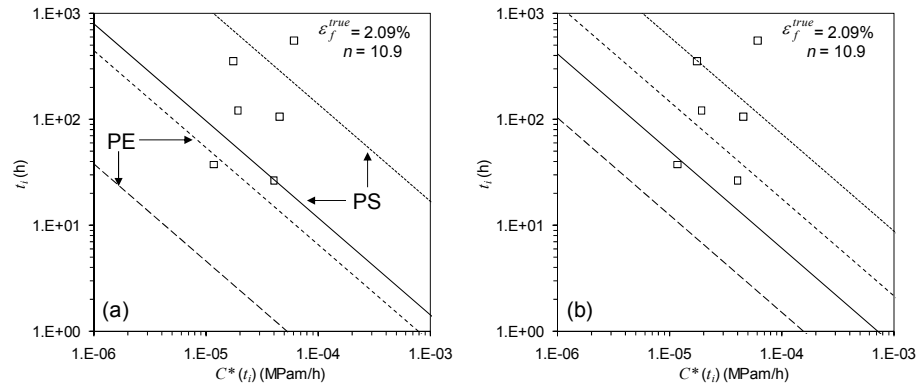


Figure C.7: Prediction of $t_{0.5}$ trends using the average creep stress exponent and axial creep ductility by (a) NSW (b) NSW-MOD (c) NSWA models



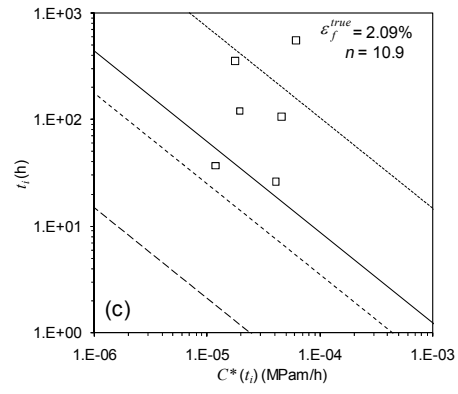


Figure C.8: Prediction of $t_{0.5}$ trends using the minimum creep stress exponent and axial creep ductility by (a) NSW (b) NSW-MOD (c) NSWA models