A Simplified Fracture Mechanics and Continuum Damage Based Numerical Modelling Approach to Predict Long Term Creep Crack Growth in Components Containing Welds

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by

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Abstract

Components in advanced gas cooled reactor (AGR) operating at elevated temperatures in the range of 500-650°C are typically susceptible to the initiation and growth of cracks due to creep. Type 316H stainless steel steam headers after a long term service are susceptible to reheat cracking in the vicinity of the weld driven by the presence of welding residual stresses. For this reason, this research has focused on developing pragmatic numerical methods for predicting creep crack growth behaviour of welded components containing residual stresses, using a simplified continuum damage and fracture mechanics method.

The work presented had three main aims. The first was to derive a comprehensive set of plastic $\eta$ factors for standard fracture mechanics geometries containing welds. The impetus for this was to improve material crack growth characterisation for welds by improving the creep $C^*$ solutions for these geometries that are presently recommended in standard codes of practice for creep crack growth testing. The second part was to experimentally examine appropriate creep material properties for as-received and service-exposed 316H stainless steels containing welds for use in numerical modelling and predictive methods for creep crack growth in a real component. The third was to develop and validate a simplified method of simulating residual stresses and creep crack growth behaviour in an ex-service AISI 316H weld header with reheat cracking. This approach simulates the presence of residual stresses using appropriate loading and boundary conditions in actual components that undergo reheat cracking without the need to develop full weld simulations to quantify them. The creep crack growth behaviour was studied using two methods based on the theories of fracture mechanics and continuum damage mechanics. Fracture mechanics parameter $C^*$ was firstly used to examine the approximate crack growth rate using the reference stress approach and approximate NSW model. The second method was to predict long term cracking by using a simplified continuum damage mechanics model, with a consideration of stress relaxation. For this purpose, a simplified multi-axial ductility exhaustion model was developed and implemented in an Abaqus user subroutine, taking into account the changes in the ex-service creep properties and the effect of reduction in creep ductility under low loads and long term operation at service temperatures. Resulting from the findings, the task was to identify the geometric and the material reasons of how and why the crack growth follows a
path of least resistance and higher constraint which did not necessarily mean growing through the welds or the heat affected zone region.
Declaration of Originality

This thesis is a presentation of my original research work. The contribution of all else are appropriately referenced. This work was done under the guidance of Professor Kamran Nikbin and Dr. Catrin M Davies at Imperial College London. It has not been submitted for a degree or diploma at any other institution and university.
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Nomenclature

Greek Symbols

\( \alpha \)  Constant in Ramberg-Osgood plasticity law
\( \alpha \)  Coefficient of thermal expansion
\( \Gamma \)  An anticlockwise integration contour around the crack tip
\( \Delta a \)  Minimum crack extension length
\( \Delta f \)  Axial creep displacement during the test
\( \Delta_{\text{load}} \)  Axial displacement due to the instantaneous elastic-plastic material response
\( \Delta_p \)  Plastic displacement
\( \dot{\Delta} \)  Creep displacement rate
\( \dot{\varepsilon} \)  Creep displacement rate
\( \dot{\varepsilon}_c \)  Creep crack mouth opening displacement rate
\( \dot{\varepsilon}_i \)  Instantaneous load line displacement rate
\( \dot{\varepsilon}_{LLD} \)  Creep load line displacement rate
\( \dot{\varepsilon}_{TLD} \)  Total experimental load-line displacement rate
\( \dot{\varepsilon}_{cTLD} \)  Creep crack displacement rate
\( \dot{\varepsilon}_{ei} \)  Elastic instantaneous load line displacement rate
\( \dot{\varepsilon}_{ip} \)  Plastic instantaneous load line displacement rate
\( \eta \)  A non-dimensional geometric factor
\( \eta_{CMOD} \)  A non-dimensional geometric factor obtained from crack mouth opening displacement
\( \eta_{LLD} \)  A non-dimensional geometric factor obtained from load line displacement
\( \eta_{WM}, \eta_{BM} \)  A non-dimensional geometric factor of weld metal and base metal
\( \eta_w \)  A modified experimental calibration factor
\( \theta \)  Angular position
\( \varepsilon \)  Strain
\( \varepsilon^c \)  Creep strain
\( \varepsilon^c(t) \)  Time dependent creep strain
\( \varepsilon^e \)  Elastic strain
\( \varepsilon^p \) Plastic strain

\( \varepsilon_{\text{eng}}^f \) Engineering creep failure strain (creep ductility)

\( \varepsilon_{\text{ROA}}^f \) Engineering reduction of area failure strain

\( \varepsilon_f^* \) Multi-axial creep ductility

\( \varepsilon_{f0} \) Uniaxial failure strain at stress \( \sigma_0 \)

\( \varepsilon_{ij} \) Strain tensors

\( \varepsilon_{ij}^p \) Plastic strain tensor

\( \varepsilon_{\text{load}}^\text{eng} \) Engineering strain during load up

\( \varepsilon_{p0} \) Ramberg-Osgood normalising strain

\( \varepsilon_{p,\text{load}}^\text{eng} \) Engineering axial plastic strain at loading

\( \varepsilon_{\text{true},p,\text{load}} \) True axial plastic strain at loading

\( \varepsilon_{\text{total}} \) Total strain

\( \varepsilon_{\text{0 WM}}, \varepsilon_{\text{0 BM}} \) Normalising strain of weld metal and base metal in the Ramberg-Osgood material model

\( \dot{i} \) Strain rate (in the absence of damage)

\( \dot{i} \) Average creep strain rate

\( \dot{i} \) Creep strain rate tensor

\( \dot{i} \) Reference creep strain rate

\( \dot{i} \) Secondary (or steady state or minimum) creep strain rate

\( \dot{i} \) Normalising strain rate in power-law creep expression and RR field

\( \varepsilon^e \) Equivalent elastic strain

\( \varepsilon^p \) Equivalent (von Mises equivalent) plastic strain

\( \varepsilon^p \) Equivalent plastic strain

\( \dot{i} \) Instantaneous equivalent creep strain rate

\( \dot{i} \) Dimensionless strain in RR field

\( \dot{i} \) Dimensionless strain in HRR field

\( \mu \) Plane stress/strain parameter in expression for \( J_{\text{ref}} \)

\( \sigma \) Applied stress
\( \sigma_e \)   Equivalent stress  
\( \sigma_f \)   Short term flow stress  
\( \sigma_{i=1,2,3} \)   Principal stress components  
\( \sigma_{ij} \)   Stress tensor  
\( \sigma_m \)   Mean (or hydrostatic) stress  
\( \sigma_m / \sigma_e \)   Triaxiality  
\( \sigma_{nom} \)   Engineering nominal net section stress  
\( \sigma_{p0} \)   Ramberg-Osgood normalising stress  
\( \sigma_{ref} \)   Reference stress  
\( \sigma_R \)   Rupture stress for the time and temperature of interest  
\( \sigma_u \)   Material ultimate tensile stress obtained from the engineering stress-strain curve  
\( \sigma_y \)   Yield stress  
\( \sigma_0 \)   Normalising stress in power-law creep expression and RR field  
\( \sigma_{0,WM}, \sigma_{0,BM} \)   Normalising stress of weld metal and base metal in the Ramberg-Osgood material model  
\( \sigma_{0.2} \)   0.2\% proof stress  
\( \sigma^{\ast}_{0.2} \)   The stress corresponding to 0.2\% inelastic strain  
\( \tilde{\sigma} \)   Effective stress  
\( \tilde{\sigma} \)   Dimensionless stress in RR field  
\( \tilde{\sigma} \)   Dimensionless stress in HRR field  
\( \sigma_\perp \)   Stress normal to the crack face  
\( \tau \)   Normalised time defined by \( \tau = t / t_r \).  
\( \mu \)   Plane stress/strain parameter in expression of \( J_{ref} \)  
\( \nu \)   Poisson’s ratio  
\( \nu_r \)   Temperature dependent constant  
\( \phi \)   Material constant to calculate CCG rate using \( C^* \)  
\( \omega \)   Creep damage parameter  
\( \dot{i} \)   The rate of creep damage accumulation  
\( \psi \)   Damage continuity
**English Letters**

- $a$: Crack length or half crack length
- $a_e$: Effective crack length
- $a_f$: Failure crack length
- $a_0$: Initial crack length
- $\dot{a}$: Creep crack growth rate
- $\dot{a}_c$: Creep crack growth rate in NSW model
- $\dot{a}_m$: Creep crack growth rate in approximate NSW model
- $\dot{a}_n$: Creep crack growth rate in Modified NSW model
- $\dot{a}_s$: Creep crack growth rate in NSW model based on creep rupture data
- $\dot{a}_0$: Steady state crack growth rate
- $\dot{a}_0$: Initial crack growth rate (at crack extension $\Delta a = 0$)
- $A$: Area under the load vs. displacement curve
- $A_A$: Norton power-law constant under average creep stage
- $A_e$: Elastic area under the load vs. displacement curve
- $A_f$: Failure cross sectional area
- $A_p$: Coefficient in power-law hardening expression for plasticity
- $A_{net}$: Real carrying area
- $A_p, A_{sec}$: Plastic area and plastic secant area on a load displacement record
- $A_p^{LD}, A_p^{CMOD}$: Plastic area under the load versus load line displacement and crack mouth opening displacement
- $A_T$: Temperature-dependent material constant
- $A_s$: Norton power-law constant under secondary stage
- $A_0$: Initial cross sectional area at loading
- $b$: Uncracked ligament size, $(W - a)$
- $B$: Specimen thickness
- $B_n$: Net thickness between the side grooves
- $B_r$: Material dependent constant, $B_r = \varepsilon_f \sigma_0^{\varepsilon_f}$
- $C^*_EPRI$: Estimate of $C^*$ from EPRI solutions
- $C_{MG}$: Constant in Monkman-Grant relation
$C_{\text{ref}}^*$ Reference stress estimate of $C^*$

d The lengths of the diagonals of the diamond shape

d$_f$ Failure diameter

d$_0$ Initial diameter

c A characteristic length of the specimen

$C_t$ Small scale creep and transition creep parameter

$C(t)$ Transient creep characterising parameter

$(C_t)_{ssc}$ The $C_t$ parameter under small scale creep conditions

$C^*$ The steady state creep fracture mechanics parameter

$D$ Material constant to calculate CCG rate using $C^*$

$D_i$ Material constants related to initiation time

$D_k$ Material constant to calculate CCG rate using $K$

$D_{\text{ref}}$ Material constant to calculate CCG rate using $\sigma_{\text{ref}}$

$E$ Young’s modulus

$E'$ Effective Young’s modulus, $E' = E$ in plane stress and $E' = E / (1 - \nu^2)$ in plane strain condition

$F$ The stress intensity factor function

$h$ Half weld groove width

$h_i(a/W,N)$ A dimensionless function of the normalised crack length and stress exponent

$h_i(n)$ Maximum of $(\epsilon, n)$ in NSW-MOD model

$H$ Half height of specimen

$H'$ Geometric function to calculate $J_p$ or $C^*$

$H_{\text{LD}}, H_{\text{MOD}}$ Geometric function to calculate $J_p$ from load line displacement and crack mouth opening displacement

$I_n$ A dimensionless integration constant in RR filed

$I_n,k$ A dimensionless integration constant in NSW-RUP model

$I_N$ A dimensionless integration constant in HRR filed

$J$ Elastic-Plastic fracture mechanics parameter

$J_e$ The elastic component of $J$

$J_{\text{EPRI}}$ Estimate of $J$ from EPRI solutions
\( J_p \)  The plastic component of \( J \)
\( J_{\text{ref}} \)  Reference stress estimate of \( J \)
\( K \)  Linear elastic stress intensity factor
\( K_1, K_{II}, K_{III} \)  Stress intensity factor for mode I, II, III loading
\( K_{lc} \)  Critical stress intensity factor
\( l_0 \)  Initial gauge length
\( L_r \)  Normalised reference stress \( \sigma_{\text{ref}} / \sigma_{0.2} \)
\( L_{r_{\text{max}}} \)  Maximum value of \( L_r \) at the cut-off point in FAD and TDFAD
\( M \)  Mismatch ratio
\( n \)  Creep stress exponent
\( n_\text{avg} \)  Average creep stress exponent
\( n_y \)  Outward normal vector to the crack tip contour
\( N \)  Ramberg-Osgood power-laws stress exponent
\( p, q \)  Empirical parameters in Spindler equation
\( P \)  Allied load
\( P_{LC} \)  Limit load
\( P_{\text{pr0}} \)  Geometry dependent normalizing load
\( Q \)  Creep activation energy
\( Q \)  Amplitude of the uniform difference between HRR prediction
\( r \)  Radial distance from crack tip
\( r_c \)  Creep process zone size at crack tip
\( r_y \)  Plastic zone near the crack tip
\( R \)  Boltzmann’s constant
\( R_i, R_o \)  Inner or outer radius in CS(T) geometry
\( s \)  Arc length along \( \Gamma \)
\( S \)  Half total length of SEN(B) specimen
\( S_{ij} \)  Deviatoric stress tensor
\( t \)  Time
\( t_c \)  Collapse time
\( t_i \)  Initiation time
\( t_r \)  Time to creep rupture
$t_T$  Transition time
$t_{0.2}$  The time for 0.2 mm crack extension
$t_{0.5}$  The time for 0.5 mm crack extension
$T$  Absolute Temperature
$T_t$  The traction vector
$T_m$  Melting temperature
$T_s$  Softening temperature
$T_0$  Ambient or uniform pre-heat temperature
$u_i$  Displacement vector
$i'$  Displacement rate vector
$V$  PD voltage
$V_f$  Failure PD voltage
$V_0$  Initial PD voltage
$W$  Specimen width or half specimen width
$W'$  Strain energy density of a non-linear elastic material
$Y(a/W)$  A dimensionless shape function that depends on geometry and mode of loading

**Abbreviations**

AGR  Advanced Gas cooled Reactor
AISI  American Iron and Steel Institute
AR  As-Received
BC  Boundary Condition
BM  Base Metal
CCI  Creep Crack Initiation
CCG  Creep Crack Growth
CS(T)  C-shaped tension specimen
C(T)  Compact Tension specimen
CTOD  Crack Tip Opening Displacement
CMOD  Crack Mouth Opening Displacement
DCT  Displacement Controlled Tension
<table>
<thead>
<tr>
<th>Abbreviation</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>DEN(T)</td>
<td>Double Edge Notch specimen in Tension</td>
</tr>
<tr>
<td>EDM</td>
<td>Electric Discharge Machining</td>
</tr>
<tr>
<td>EPFM</td>
<td>Elastic-Plastic Fracture Mechanics</td>
</tr>
<tr>
<td>EPRI</td>
<td>Electrical Power and Research Institute</td>
</tr>
<tr>
<td>FE</td>
<td>Finite Element</td>
</tr>
<tr>
<td>FB</td>
<td>Fusion Boundary</td>
</tr>
<tr>
<td>HRR</td>
<td>Hutchinson-Rice-Rosengren elastic-plastic crack tip field solution</td>
</tr>
<tr>
<td>HAZ</td>
<td>Heat Affected Zone</td>
</tr>
<tr>
<td>HT</td>
<td>High Temperature</td>
</tr>
<tr>
<td>LC</td>
<td>Loading Condition</td>
</tr>
<tr>
<td>LCT</td>
<td>Load Controlled Tension</td>
</tr>
<tr>
<td>LVDT</td>
<td>Linear Variable Differential Transformer</td>
</tr>
<tr>
<td>LEFM</td>
<td>Linear Elastic Fracture Mechanics</td>
</tr>
<tr>
<td>MSF</td>
<td>Multiaxial Strain Factor</td>
</tr>
<tr>
<td>NSW</td>
<td>Nikbin-Smith-Webster CCG models</td>
</tr>
<tr>
<td>NSW-A</td>
<td>Approximate NSW model</td>
</tr>
<tr>
<td>NSW-MOD</td>
<td>Modified version of the NSW model</td>
</tr>
<tr>
<td>NSW-RUP</td>
<td>NSW model based on creep rupture data</td>
</tr>
<tr>
<td>NW</td>
<td>New Welded</td>
</tr>
<tr>
<td>$P_e, P_\sigma$</td>
<td>Plane strain, Plane stress</td>
</tr>
<tr>
<td>PC</td>
<td>Pre-Compressed</td>
</tr>
<tr>
<td>PD</td>
<td>Potential Drop</td>
</tr>
<tr>
<td>PWHT</td>
<td>Post-Weld Heat Treatment</td>
</tr>
<tr>
<td>ROA</td>
<td>Reduction Of Area</td>
</tr>
<tr>
<td>RR</td>
<td>Riedel and Rice creep crack tip field solution</td>
</tr>
<tr>
<td>RS</td>
<td>Residual Stress</td>
</tr>
<tr>
<td>RT</td>
<td>Room Temperature</td>
</tr>
<tr>
<td>SEN(B)</td>
<td>Single Edge Notch specimen in Bending</td>
</tr>
<tr>
<td>SEN(T)</td>
<td>Single Edge Notch specimen in Tension</td>
</tr>
<tr>
<td>SIF</td>
<td>Stress Intensity Factor</td>
</tr>
<tr>
<td>SSC</td>
<td>Small Scale Creep</td>
</tr>
<tr>
<td>SSY</td>
<td>Small Scale Yielding</td>
</tr>
<tr>
<td>WM</td>
<td>Weld Metal</td>
</tr>
<tr>
<td>------</td>
<td>------------</td>
</tr>
<tr>
<td>XW</td>
<td>Cross Weld</td>
</tr>
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</table>
Chapter 1

Introduction

1.1 Background

In the power generation industry, especially in the nuclear field, the safe lifetime of structural components operating at high temperatures in the range of 500-650°C is limited by processes such as the initiation and growth of cracks due to creep. During fabrication of some components, such as boiler’s steam headers, steam pipes, chemical reactors and turbine castings, welding is inevitable for joining two components together. The heat-affected zone (HAZ) is generally the region with the lowest creep resistance. Failures often occur due to creep damage accumulation in this region and thus primarily affect the duration of a component’s lifetime. In high-temperature plant, it is essential to assess the crack initiation and growth behaviour in order to optimise the component’s lifetime whilst maintaining safety. The considerations required for an assessment include the operating conditions, the nature of material creep deformation, material properties relevant to the assessed feature (including the base metal, weld metal and heat-affected zone) and the structural constraint[1]. In this work, study of the combined factors for high temperature components were performed in order to predict creep crack initiation and growth in long term components containing welds under given load and boundary conditions, using a simplified fracture mechanics and continuum damage based numerical modelling approach.

There is strong evidence that residual stresses are presented which can be the driving force for creep cracking in the high temperature components containing welds [2]. Residual stresses can be developed due to the different microstructural, thermal and mechanical properties between the neighbouring base metal and weld metal, leading to a mismatch in strains in the cross-weld regions [3]. Evidence establishes that such stresses are often tri-axial and may affect the crack driving forces, alter crack-tip controlled conditions and failure mechanisms [4]. The residual stress contributions on the tensile, creep rupture and creep crack behaviours were analysed in this research. Also, the effects of material properties, geometrical constraints and stress
distributions could also alter crack tip behaviours, further affect the component’s lifetime. Therefore, in order to improve lifetime predictions of the studied welded components, this research has investigated the creep failure behaviour of welded components containing residual stresses based on a Finite Element (FE) model, considering the loading, material and constraint effects. The details are presented in this thesis using comprehensive experimental, analytical and numerical approaches to reach the objectives of the thesis.

1.2 Objectives

The primary aim of this PhD project is to investigate and simulate the effects of residual stresses and crack-tip constraints on the deformation, creep crack initiation and growth for 316H stainless steel components, employing both experimental methods and finite element modelling techniques. The main objectives are:

1. To determine the elastic/plastic fracture mechanics parameter $J$-integral and its analogous creep fracture parameter $C^*$ properties on a range of welded fracture mechanics specimens using the ‘$\eta$ factor’.
2. To perform high-temperature experimental testing to measure tensile, creep deformation and crack growth behaviours of service exposed base and weld materials that can be employed in the continuum damage model.
3. To identify a pragmatic and simplified method of introducing secondary stresses to simulate welding residual stresses in the header under optimised load and boundary conditions.
4. To develop appropriate FE models to predict stress distributions, damage and failure properties of welded components in a service exposed cracked header component, using simplified continuum damage and fracture mechanics based crack growth model.
5. To validate and identify key material and geometric variables from the FE sensitivity studies in order to model and predict the failure behaviour of high-temperature power plant header components in a more conclusive manner.
1.3 Structure of Thesis

This thesis contains eight chapters. Chapter 2 starts with the principles of deformation that perform at high temperatures in metallic materials, including several specific topics, namely elastic-plastic deformation, creep deformation, creep failure mechanics. Two approaches based on the concepts of fracture mechanics and continuum damage mechanics have been introduced, which could be used for correlating and predicting the creep crack growth of an actual component. It is followed by a discussion on the creep reheat cracking behaviours.

Chapter 3 systematically investigates the geometry dependent fracture mechanics parameter ‘\( \eta \)’ factor based on the record of load vs. load line displacement and crack mouth opening displacement. The \( \eta \) factor was evaluated for six fracture mechanics geometries containing thin welds using non-linear FE method for power-law hardening materials and the results were compared with literatures where available. A range of base/weld mismatched material properties were considered and the recommended \( \eta \) for normalised crack lengths were given as a function of mismatch factors. It is shown that the exact value of the \( \eta \) factor has an insignificant part in the accuracy of the correlating parameter regardless of the material variables.

In Chapter 4, materials and experimental techniques used in this study are introduced. The aim was to provide an initial understanding of the experimental testing methods and analysis procedures for hardness, tensile, uniaxial creep rupture and creep crack growth tests which were applied in this work. The material properties appropriate to the fracture mechanics and continuum damage analysis were performed in the case of cross-weld regions, weld metals and base metals over the appropriate temperature and loading conditions. The experimental test matrix adopted for this work is also given.

The experimental tensile and uniaxial creep test results on the studied materials are described in Chapter 5. The creep crack initiation and growth test data are shown in Chapter 6. In this work, samples containing cross-weld regions, existing highly triaxial tensile welding residual stresses were tested. The testing results were compared with as-received ex-service and room temperature pre-compressed to 8% plastic strain base materials in order to reflect possible variations due to the existence of welding residual stresses. The cross-weld and as-received
material properties were used in numerical modelling and predictive methods for creep crack growth.

In Chapter 7, the reheat cracks found in the vicinity of the weld in a header component are studied. Previous studies have investigated a similar header/nozzle component by measuring the residual stress distributions using deep-hole drilling method and predicting the trends based on welding simulation methods. Without considering the full weld simulation, a simplified FE model was presented for residual stress simulation and creep damage analysis. To carry out this study, the examination of load type and boundary conditions were performed to simulate an effective residual stress distribution that relaxed during creep, the optimal results were then compared with previous measured and numerically simulated residual stress profiles. Discussions on creep damage based on fracture mechanics and continuum damage mechanics were also performed. In fracture mechanics, the creep crack growth was approximately predicted based on the simplified NSW model. In continuum damage mechanics, the crack extensions were also presented to identify the effects and sensitivities to geometrical constraint, loading profile, ex-service degradation and material inhomogeneity and creep failure ductility. Finally, the creep crack growth behaviour of the header/nozzle component obtained from continuum damage model were correlated with fracture mechanics parameters and also compared with available data on C(T) specimens.

The conclusions of this study as well as the suggestions for future work are stated in Chapter 8. Available Appendices are also included in the end which detail additional analytical results from Chapter 3.
Chapter 2

Literature Review

2.1 Introduction

Welded components are among the most important construction elements in power generation industry [5]. Over 40,000 to 50,000 weldments are estimated for a 300-MW unit [6]. Typically, those weldments are operated at elevated temperatures. Under such circumstances, consideration of safe and reliable service of welded components should be taken in performance of power and other plants because residual stresses that arise in weldments as a consequence of local heating and cooling during welding processes would give rise to the creep crack behaviours. For this reason, research work on the creep behaviours of cross-weld region is of great importance and has been carried out extensively.

During high temperature service, the existence of welding residual stresses could result in initiation and propagation of creep induced cracks which may initially exist in or near the boundary regions between base and weld materials, therefore, the dissimilar metal joints are prone to a creep failure mode. In high-temperature components, there is a prerequisite to estimate the creep-induced defects in weldments in order to evaluate the component’s lifetime. Two methods are introduced below, namely fracture mechanics and continuum damage concepts. Fracture mechanics concept has been applied to investigate a crack of finite size present in a recognised testing component and then evaluate its propagation in the creeping conditions [7]. Continuum damage concept has been used to estimate the time history on the crack continuum damage accumulation, considering different material and loading conditions.

The research to be described in this thesis deals with the creep properties of weldments in service-aged components. This chapter has reviewed any contributions from previous research, aiming to demonstrate the main problems associated with the creep properties of welds and the ways in which they are tackled at present. The structure of this section has been illustrated in Section 1.3. The concepts of elastic-plastic and creep defamation are reviewed, followed by the
fracture mechanics assessment methods and state of art of continuum damage concepts which is directly relevant to the present work. In addition, the fundamental concepts and effects of reheat cracks which primarily result from welding residual stresses have been discussed.

### 2.2 Elastic-Plastic Deformation

#### 2.2.1 Introduction

The total strain, \( \varepsilon_{\text{total}} \) in high temperatures includes time dependent strain, \( \varepsilon^c(t) \), and the time independent strain which can be subdivided into elastic \( \varepsilon^e \) and plastic \( \varepsilon^p \) components, as

\[
\varepsilon_{\text{total}} = \varepsilon^e + \varepsilon^p + \varepsilon^c(t)
\]

For metallic materials at stress below the yield stress \( (\sigma_y) \), the concept of Hooke’s law is well described a linear elastic deformation under a quasistatic condition, in a form of

\[
\varepsilon^e = \frac{\sigma}{E} \tag{2.2}
\]

where \( \sigma \) is applied stress and \( E \) is Young’s modulus.

When the stress beyond the yield stress, the theory of plasticity is applicable to materials that exhibit time-independent, non-recoverable and non-linear behaviour. Generally, the non-linear (plastic) deformation, \( \varepsilon^p \) is represented by a power-law form as

\[
\varepsilon^p = \alpha \varepsilon^p_0 \left( \frac{\sigma}{\sigma^p_0} \right)^N = A_p \sigma^N \tag{2.3}
\]

where \( N \) is Ramberg-Osgood power-law stress exponent, \( \varepsilon^p_0 \) and \( \sigma^p_0 \) are Ramberg-Osgood normalising strain and stress, \( \alpha \) is constant in Ramberg-Osgood plasticity law and \( A_p \) is coefficient in power-law hardening expression for plasticity, which is equal to \( A_p = \frac{\alpha \varepsilon^p_0}{\sigma^p_0 N} \).

Under uniaxial quasistatic conditions, it could be assumed that the stress-strain behaviour is linear-related below yield stress whilst in a power-law form beyond yield stress, as described below:
\[
\frac{\varepsilon}{\varepsilon_y} = \frac{\sigma}{\sigma_y}, \text{ for } \sigma < \sigma_y
\]
\hspace{1cm} (2.4)

\[
\frac{\varepsilon}{\varepsilon_y} = \left(\frac{\sigma}{\sigma_y}\right)^N, \text{ for } \sigma > \sigma_y
\]
\hspace{1cm} (2.5)

### 2.2.2 Ramberg-Osgood Material Model

During plastic deformation, work hardening or strain hardening would occur due to dislocation generation and movement within the crystal structures of the materials. Ramberg-Osgood material model [8] is commonly used to describe the stress-strain behaviour of strain hardening materials. It provides a smooth continuous curve for the total strain in terms of stress, with no distinct yield point, and it can be defined by [9]

\[
\varepsilon = \frac{\sigma}{E} + A_p\sigma^N, \text{ in a uniaxial form}
\]
\hspace{1cm} (2.6)

\[
\frac{\varepsilon}{\varepsilon_{p0}} = \frac{\sigma}{\sigma_{p0}} + \alpha \left(\frac{\sigma}{\sigma_{p0}}\right)^N, \text{ in a non-dimensional form}
\]
\hspace{1cm} (2.7)

Typical stress-strain curve in Ramberg-Osgood relation is illustrated in Figure 2.1. The hardening behaviour of the material relies on the material constants, \(\alpha\) and \(N\). The normalising strain, \(\varepsilon_{p0}\), can be considered as a yield offset. For most materials, the yield offset is equal to the accepted value of \(\varepsilon_{p0} = 0.2\%\), the corresponding normalising stress is then taken to be the 0.2\% proof stress, written as \(\sigma_{0.2}\). Due to the power-law form in Ramberg-Osgood model, plastic strain is always considered small for very low levels of stress (\(\sigma < \sigma_{0.2}\)). Under this circumstance, the non-linear (plastic) term in Equation (2.7) remains negligible, while the linear term dominates. On the other hand, when \(\sigma > \sigma_{0.2}\), progressively dominated plastic strain results in a magnified power-law distribution, under such condition, the linear component in Equation (2.7) becomes negligible.

### 2.2.3 Multi-axial Deformation

The previous discussion on stress-strain behaviour is based on uniaxial conditions. However, in practice, the body is majorly subjected to a multi-axial stress state, in this case the effects of the
stresses that are applied in different directions cannot be superimposed linearly. Two classical plasticity theories are commonly used to determine whether and when the yield occurred under multi-axial deformations. The interpretations of Tresca criterion is that a critical value of shear stress stored in engineering materials; while von Mises criterion suggests a yield occurs when distortional (shear strain) energy reaches a critical value.

In fracture mechanics, von Mises criterion is generally applied to predict yielding of materials under any loading conditions from the results of uniaxial tensile tests. The equivalent stress, denoted as $\sigma_e$, is in the function of principle stresses as

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

(2.8)

where $\sigma_1$, $\sigma_2$, $\sigma_3$ are principle stress components. Note that under uniaxial condition, $\sigma_2 = \sigma_3 = 0$, von Mises stress is then equal to the principle stress, i.e. $\sigma_e = \sigma_1$.

The mean (or hydrostatic) stress, denoted as $\sigma_m$, is required consideration. It is defined as the average of the three normal stress components of any stress tensor as

$$
\sigma_m = \frac{\sigma_1 + \sigma_2 + \sigma_3}{3}
$$

(2.9)

### 2.2.4 Limit Load and Reference Stress Concepts

The concepts of limit load and reference stress have been widely applied in structural engineering design and components’ integrity assessments. The limit (or plastic collapse) load describes a maximum sustainable load of the uncracked or cracked body made of elastic-perfect plastic materials ($N \to \infty$) and is often used to determine the carrying capacity of structures. Knowledge of limit load can aid designing mechanical properties of components and structures, since limit load provides information of the modes of failure associated with load-controlled effects [10].

The reference stress method has been employed for estimating structural integrity of components with and without defects in either below or within the creeping range of temperatures. The reference stress is defined by
In Equation (2.10), the normalised reference stress \( \sigma_{\text{ref}} / \sigma_{0.2} \) denotes the level of plasticity in structural components. If the ratio is less than unity, components may store a limit level of plasticity; if the ratio is greater than unity, amount of plasticity in the components is expected.

The limit load, \( P_{\text{LC}} \) in Equation (2.10) is the function of normalised crack size and specimen thickness. \( P_{\text{LC}} \) is related to the yield stress, \( \sigma_y \) (or \( \sigma_{0.2} \)) whilst \( \sigma_{\text{ref}} \) is independent of \( \sigma_y \). The value of \( P_{\text{LC}} \) under plane strain condition is generally greater than that under plane stress condition. The information of \( P_{\text{LC}} \) in each fracture mechanic specimen is tabulated in Ref [11].

2.3 Creep Deformation

2.3.1 Introduction to Creep

Components that are subjected to stresses at temperature greater than 30% of their absolute melting temperature may fail by slow and non-reversible deformation referred as creep [12]. The material creep behaviours are a deformation pattern that changes continuously as a function of time under constant loads [13].

Creep in polycrystalline materials is generally the results of diffusion and dislocations processes. Since the mobility of atoms increases rapidly with temperatures, diffusion-controlled processes can significantly affect high temperature mechanical properties. Also, high temperature would accelerate the movement of dislocations by the mechanisms of climb and glide [14].

2.3.2 Deformation Mechanics (or Ashby) Map

A particular creep mechanism may predominate at a given temperature and stress, therefore, the mechanics of creep deformation can be described by a deformation mechanics (or Ashby) map, where the dominated mechanism is mainly related to normalised tensile stress (or stress normalised by Young’s modulus) and homologous temperature (or the absolute temperature, \( T \) normalised by melting temperature \( T_m \)). An example of deformation mechanics map for 316 stainless steel has been illustrated in Figure 2.2 [15].
Previously, numerous uniaxial creep testing for 316 stainless steel have been performed for a wide range of constant stresses from 50 MPa to 350 MPa and for a temperature range of 500-750°C. The applied region in deformation mechanics map is constructed, as shown in Figure 2.3 [16, 17]. Consideration should be taken on the main creep mechanisms while formulating the material model: power-law creep and viscous (linear) creep. In laboratory testing with higher applied stresses, the creep regions lay mostly in the power-law region. But in industrial application, the creep in reality may occur in the ‘viscous creep’ region corresponding to medium or low applied stresses (<45MPa) at a temperature range of 490-550°C [18]. Taking into account the different creep behaviours depending on stresses, the extrapolation methods relying mainly on the power-law region used for the evaluation of the creep properties require further considerations.

### 2.3.3 The Creep Curve

A creep curve is a plot of creep strain, $\varepsilon_c$ against time, $t$. To determine the engineering creep curve of a metallic material, a constant stress is applied to a uniaxial creep specimen during a creep test where the strain of the specimen is monitored as a function of time. The general procedures for creep testing are standardised in ASTM E139-11 [19]. A typical curve is illustrated in Figure 2.4 where $\varepsilon_f$ is the uniaxial creep ductility and $t_r$ is the time to creep rupture.

During loading-up, there is an initial rapid elongation of the specimen. It is considered as an incubation period due to time independent elastic and plastic strains (also referred as loading strains). After the loading-up period, time dependent (creep) strain is naturally separated into primary, secondary and tertiary creep.

In the primary creep, creep rate would decrease with time. This is solely due to the dislocation movement. Strain hardening dominates the primary region and occurs rapidly than the corresponding thermal recovery procedure.

Subsequently, the steady-state creep state is observed in the secondary region of creep. At high temperature, recovery processes, such as climbing and cross slipping allow annihilation of dislocation movement and make a balance between work hardening and thermal softening.
effects. The constant slope in the secondary creep is referred to secondary (or steady state or minimum) creep strain rate,  \( \dot{\epsilon} \) [20].

The next increasing strain rate stage is referred as tertiary creep stage. Since the tertiary creep stage is the beginning of a failure process, it is not considered in the evaluation of components’ service life.

From the creep curve, a number of material’s creep properties can be deduced. The most important factors are duration of each of creep stages, rupture time  \( t_r \), uniaxial creep ductility  \( \varepsilon_f \) and creep strain rate, particularly steady state rate  \( \dot{\epsilon}_{ss} \) and average rate  \( \dot{\epsilon}_A \).

The instantaneous creep strain rate,  \( \dot{\epsilon} \) depends on stress, temperature and ‘state’ of the material. Usually, the ‘state’ can be described by either the creep strain given by a strain hardening law or the current time defined by a time hardening law, expressed as

\[
\begin{align*}
\text{Strain hardening law: } & \dot{\epsilon} = f(T, \sigma, \varepsilon) \quad (2.11) \\
\text{Time hardening law: } & \dot{\epsilon} = f(T, t) \quad (2.12)
\end{align*}
\]

The results obtained from those two laws are same in the secondary stage, but different in the primary and tertiary stages. Strain hardening law is more suitable if the behaviour of work hardening dominates the creep during the primary stage, while time hardening law is preferable if aging process plays a leading role during the tertiary stage.

### 2.3.4 Power-law Representation of Creep

The relative length of the primary, secondary and tertiary stages in Figure 2.4 will depend on the material and testing conditions. The secondary creep is the most important creep region for most engineering materials and usually takes the longest portion in the life of creep. The secondary creep strain rate should obey the Arrhenius law since creep is a thermally activated process. Dorn [1961] established a relationship between steady creep strain rate,  \( \dot{\epsilon} \) and steady state creep stress,  \( \sigma \), as

\[
\dot{\epsilon}_{ss} = \exp \left( -\frac{Q}{RT} \right) \quad (2.13)
\]
where \( A_T \) is temperature-dependent material constant, \( Q \) is creep activation energy, \( R \) is the universal gas constant.

Assuming that materials are creeping at a constant temperature under uniaxial stress conditions, the steady state creep strain rate for the power-law creeping materials can be expressed as

\[
\dot{\varepsilon} = \frac{\dot{\varepsilon}_0}{(\sigma/\sigma_0)^n} = A_s \sigma^n
\]

where \( \dot{\varepsilon}_0 \) and \( \sigma_0 \) are normalising strain rate and stress in power-law creep expression, \( A_s \) is Norton power-law constant under the secondary stage.

The above equation is known as Norton’s law. The reason for its popularity is its simplicity in application to stress analysis and its analogy to the Ramberg-Osgood law as seen in Equation (2.2). If plotting minimum creep strain rate, \( \dot{\varepsilon} \), versus \( \sigma \) on log-log form, the values of the steady state power-law constants \( A_s \) and \( n \) can be extrapolated.

### 2.3.5 Creep Fracture

The next important engineering parameter, after \( \dot{\varepsilon} \), is the time to rupture (or fracture time), \( t_f \).

Trans-granular and inter-granular microstructural failure mechanisms may exist in a creep fracture which is affected by diffusional flow, dislocation creep and rupture [21]. Fracture mechanics map is used for identifying modes of failures under given temperature and stress. An example of fracture mechanics failure map for type 316 stainless steel is shown in Figure 2.5 [22]. As seen in Figure 2.5, the trans-granular creep fracture occurs at short creep lifetime and relatively low temperature or high stress, while the inter-granular failure occurs at long lifetime and higher temperature or lower stress; rupture is another failure mode that occurs at high temperature due to dissolution of the carbides, local high temperature recovery and dynamic recrystallization [23].

As long-term creep properties of structural materials used in high-temperature plants are evaluated from short-term experimental data, the lifetime assessment in engineering components requires interpolating or extrapolating other stress and temperature conditions in an appropriate form. The conventional approach to predict the lifetime of long-term tests from short-term tests
are presented using Monkman-Grant relation [24], it states that the strain accumulated during secondary creep is approximately same at failure so that the product of the secondary creep rate and the rupture time can be considered as a constant, i.e.

\[ \dot{\varepsilon} = \alpha \cdot \varepsilon \]  \hspace{1cm} (2.15)

As \( \dot{\varepsilon} \) can be described in a power-law form, the rupture time \( t_r \) remains a power-law form following Equation (2.15), written as

\[ t_r = \frac{\varepsilon_f}{\dot{\varepsilon}} \left( \frac{\sigma}{\sigma_0} \right)^{-\nu_r} = B_r \sigma^{-\nu_r} \]  \hspace{1cm} (2.16)

where \( \varepsilon_f \) is the uniaxial failure strain at the stress \( \sigma_0 \), \( B_r = \varepsilon_f \sigma_0^{-\nu_r} / \dot{\varepsilon} \) which is a material dependent constant, \( \nu_r \) is a temperature dependent constant, ideally, \( \nu_r \approx n \) according to Equation (2.15). The constant \( B_r \) and \( \nu_r \) can be obtained during creep test.

### 2.3.6 Average Creep Strain Rate

Following Norton’s creep law as shown in Equation (2.14), the steady creep strain rate can be obtained at the given stress during the secondary stage. Account for all three creep stages, an approximate method is generally used. The effects of primary, secondary and tertiary creep are usually described by an average creep strain rate, \( \dot{\varepsilon}_A \). In this case, the average creep strain rate is usually defined by the ratio of the creep ductility to the time to rupture as

\[ \dot{\varepsilon}_A = \frac{\varepsilon_A}{t_r} \]  \hspace{1cm} (2.17)

The definition of steady and average creep strain rates is shown in Figure 2.6. The average creep strain rate is always greater than the steady creep strain rate. Similar to the behaviour of \( \dot{\varepsilon} \), for the given stress and temperature, \( \dot{\varepsilon}_A \) can be also expressed as a power-law form as

\[ \dot{\varepsilon}_A = A_d \sigma^{-n_d} \left( \frac{\sigma}{\sigma_0} \right)^{-\nu_d} \]  \hspace{1cm} (2.18)
where \( n_A \) and \( A_A \) are material constants which can be obtained from rupture data. The dimension of \( \dot{\varepsilon} \) is often taken as [h\(^{-1}\)]. \( \sigma_0 \) is obtained from \( A_A = (\dot{\varepsilon} \sigma_0) \). As seen in Figure 2.6, during the early stages of the creep test, the average creep strain may under-predict the creep strain accumulation (mainly during the primary creep) but over-predict the creep strains during the late stage of creep. The extent of under or over predictions of creep strains in the different regions will depend on the magnitude of the strains in each stage and the test conditions.

According to Equation (2.16), (2.17) and (2.18), the creep strain at failure (creep ductility) for the given stress and temperature can be obtained as

\[
\varepsilon_f = \dot{\varepsilon} \left( \frac{\sigma}{\sigma_0} \right)^{n_A \varepsilon_f} = A_A B_r \sigma^{n_r \varepsilon_f} \tag{2.19}
\]

This equation predicts the sensitivity of the failure strain in terms of the applied stresses. The creep ductility increases with the increasing stress as \( n_A > \nu_r \), but becomes independent on the stresses as \( n_A = \nu_r \). According to the above statement, the uniaxial creep failure strain commonly has two types of behaviour which are associated with particular cavity growth behaviour, which are [25]:

1. Ductility increases with the increasing net section stresses, which usually forms a lower and upper shelf (Figure 2.7 (a) : P91 base metal at 600°C [25]).
2. No clear trends of ductility variations with stresses (Figure 2.7 (b): 316L(N) materials at 550°C [25]).

The behaviour of first type in Figure 2.7 (a) is briefly described here. In lower shelf region, creep failure behaviour is controlled by constrained cavity growth [26] while in upper plateau, the failure is controlled by plastic hole growth behaviour [27]. Those two effects are independent on the creep strain rate or the net section stress. Between the lower and upper shelf, there is a transition region, in this region, the creep failure may due to the diffusional control cavity growth as proposed by Hull and Rimmer [28] where the creep ductility increases with increasing the creep strain rate or the net section stress. The relationship between ductility and net section stress of the studied AISI type 316H stainless steel at 550°C follows the first type.
2.4 Fracture Mechanics Concepts

2.4.1 Introduction

Fracture mechanics concepts are concerned with deriving a stress, material and geometry dependent singularity which can characterise magnitude of the stresses and strains at vicinity of the crack, and also describe a criterion for crack initiation, growth and subsequent failure, in order to determine the properties of the materials to be used.

This section firstly presents fundamental concepts of linear elastic and elastic-plastic fracture mechanic. The methods applied to calculate the singular parameter are briefly reviewed using $K$ and $J$ parameters respectively. Subsequently, a more coverage is given on the time-dependent fracture mechanics which is used under creeping conditions. The $C^*$, $C(t)$ and $C_t$ parameters for characterising creep crack growth are introduced, together with its application on crack initiation and propagation.

2.4.2 Linear Elastic Fracture Mechanics (LEFM)

In a cracked or notched body under linear elastic deformation, the stresses near the crack tip are higher than the remote applied stress, and varies with $1/\sqrt{r}$. The stress in the vicinity of the crack tip may be described by the stress intensity factor (SIF), $K$, which was firstly introduced by Irwin in 1957 [29]. The SIF factor is usually given a subscript to denote the mode of loading, i.e. $K_I$, $K_{II}$ or $K_{III}$. Since Mode I loading which means the applied stress normal to the crack plane is performed in the majority of fracture mechanics test, the subscript in this research is omitted.

2.4.2.1 The Stress Intensity Factor (SIF), $K$

In isotropic linear-elastic material model, the stress distribution ahead of the crack tip can be obtained if the stress intensity factor is known. $K$ can describe the singular stress field ahead of the crack tip and predict fracture of the specimen when $K$ reaches the critical value, the corresponding $K$ is denoted as critical stress intensity factor ($K_{ic}$). The near-tip stress field for a mode-I crack can be expressed as [30]
\[
\sigma_{ij} = \frac{K}{\sqrt{2\pi r}} f_\theta(\theta)
\]

(2.20)

where \(\sigma_{ij}\) is stress tensor; \(r\) is distance away from crack tip.

According to Equation (2.20), stress near the crack tip increases in proportion to \(K\). The solution for a crack in a finite body is written with an appropriate correction factor \(Y(a/W)\) by

\[
K = Y(a/W)\sigma \sqrt{\pi a}
\]

(2.21)

where \(\sigma\) is the applied stress, \(a/W\) is the normalised crack length and \(Y(a/W)\) is a dimensionless shape function that depends on geometry and mode of loading. The solution of \(Y(a/W)\) for conventional fracture mechanics geometries can be found in fracture mechanics textbook, such as [11].

### 2.4.2.2 Small Scale Yielding (SSY)

Linear elastic stress analysis of sharp cracks predicts infinite stresses at the crack tip according to Equation (2.20). In real materials, however, stresses at the crack tip are finite due to the existence of plastic zone. Simple corrections to linear elastic fracture mechanics are available under small scale yielding (SSY) conditions which means the plastic zone is small compared to relevant dimensions of geometries.

Stresses are redistributed beyond the yield stress, thus leading to a higher effective stress intensity factor. Irwin [29] accounted for this increase in \(K\) by defining an effective crack length that is slightly longer than the actual crack length. Under SSY conditions, plasticity effects can be incorporated in LEFM analyses by replacing the crack length \(a\) in Equation (2.21) with effective crack length, \(a_e\). The correction of \(K\) in SSY condition can be written as [31]

\[
K = Y(a/W)\sigma \sqrt{\pi (a + r_y)}
\]

(2.22)

where \(r_y\) is the plastic zone size near the crack tip.
2.4.3 Elastic-Plastic Fracture Mechanics (EPFM)

Linear elastic fracture mechanics is only applicable as long as nonlinear material deformation is confined to a small scale region surrounding the crack tip. As sufficient plastic deformations occur in many materials, an alternative elastic-plastic fracture mechanics (EPFM) is required. EPFM applied to materials that exhibit time-independent nonlinear behaviour (i.e. plastic deformation). Although the stress and strain behaviour can be characterised by $J$ integral and crack tip opening displacement (CTOD), only $J$ method is introduced here.

2.4.3.1 $J$-Integral

Rice [32, 33] described that for non-linear elastic materials, the crack tip stress fields could be correlated with $J$ -integral. The use of $J$ provides a means of directly extending LEFM behaviour to fully plastic behaviour. $J$ can be defined as

$$J = \int_{\Gamma} \left( W d\gamma - T_i \frac{\partial u_i}{\partial x} ds \right)$$

(2.23)

where $\Gamma$ is an anti-clockwise integration contour around the crack tip as seen in Figure 2.8, $s$ is the arc length along $\Gamma$, $W$ is the strain energy density of a non-linear elastic material, expressed as

$$W = W(\varepsilon_y) = \int_0^{\varepsilon_y} \sigma_{\gamma} d\varepsilon_{\gamma}$$

(2.24)

where $\sigma_{\gamma}$ and $\varepsilon_{\gamma}$ are the stress and strain tensors. $T_i$ is the traction vector defined by the outward normal vector $n_{\gamma}$ along $\Gamma$

$$T_i = \sigma_{\gamma} n_{\gamma}$$

(2.25)

$J$ -integral can be also used in linear elastic conditions. It proves that Equation (2.23) is valid for any contour around a crack tip due to its path-independency.

2.4.3.2 HRR Fields

As a criterion of fracture mechanics, Hutchison [34, 35], Rosengren and Rice [33] independently determined that $J$ -integral could be a crack tip stress or strain singularity for large-scale yielding for non-linear power-law hardening materials. Thus $J$ is a stress intensity parameter,
which describes the amplitude of the stress and strain field, known as HRR filed. The HRR solutions can be expressed as

\[
\frac{\sigma_{ij}}{\sigma_{p0}} = \left[ \frac{J}{\alpha \sigma_{p0} \varepsilon_{p0} I_{N} r} \right]^{\frac{1}{N+1}} \tilde{c}
\]

(2.26)

\[
\frac{\varepsilon_{ij}}{\varepsilon_{p0}} = \alpha \left[ \frac{J}{\alpha \sigma_{p0} \varepsilon_{p0} I_{N} r} \right]^{\frac{N}{N+1}} \tilde{\varepsilon}
\]

(2.27)

where \(\tilde{c}\) and \(\tilde{\varepsilon}\) are the dimensionless functions of \(N\) and \(\theta\) under power-law conditions. \(\alpha\), \(\sigma_{p0}\) and \(\varepsilon_{p0}\) are the Ramberg-Osgood material model constants (as seen in Equation (2.7)), \(I_{N}\) is the dimensionless integration constant that depends on \(N\), the values of \(I_{N}\) can be estimated from the following equations [21]

Plane stress: \(I_{N} = 7.2 \sqrt{0.12 + \frac{1}{N} - \frac{2.9}{N}}\) (2.28)

Plane strain: \(I_{N} = 10.3 \sqrt{0.13 + \frac{1}{N} - \frac{4.6}{N}}\) (2.29)

**2.4.3.3 J-Integral Estimation Methods**

For a non-linear elastic material, \(J\) can be separated into elastic \(J_{e}\) and plastic \(J_{p}\) parts, written by

\[
J = J_{e} + J_{p}
\]

(2.30)

where the elastic part, \(J_{e}\) is related to the stress intensity factor \(K\) and the effective elastic modulus of the material \(E^{'}\) which is equal to \(E\) and \(E/(1-\nu^{2})\) under plane stress and plane strain condition respectively

\[
J_{e} = \frac{K^{2}}{E^{'}}
\]

(2.31)
\( J_p \) can be obtained from the plastic area under the load-displacement curve (Figure 2.9) based on the respond of load-line displacement (LLD) or crack mouth opening displacement (CMOD), expressed as

\[
J_p = \frac{A_p^{LLD}}{B_n(W-a)} \eta^{LLD} \tag{2.32}
\]

\[
J_p = \frac{A_p^{CMOD}}{B_n(W-a)} \eta^{CMOD} \tag{2.33}
\]

where \( B_n \) is the net thickness between the side grooves, \( A_p^{LLD}, A_p^{CMOD} \) is plastic area under the load versus load line displacement or crack mouth opening displacement, \( \eta^{LLD}, \eta^{CMOD} \) is a non-dimensional geometric factor obtained from load line displacement and crack mouth opening displacement respectively. The evaluations of \( \eta \) factor in different fracture mechanics specimens have been comprehensively discussed in Chapter 3.

Both \( A_p^{LLD} \) and \( A_p^{CMOD} \) can be obtained by integrating an appropriate load-displacement area for each cracked configuration. Also, \( \eta \) is related to an appropriately chosen area of the load-displacement curve for each specimen. There are two areas which could be selected, namely \( A_p \) and \( A_{sec} \). The definition of \( A_p \) and \( A_{sec} \) is shown in Figure 2.9 (a) and (b) separately. These areas may be expressed as

\[
A_p = \int_0^{\Delta_p} Pd\Delta_p \tag{2.34}
\]

\[
A_{sec} = \int_0^{\Delta_p} Pd\Delta_p - \frac{1}{2} P\Delta_p \tag{2.35}
\]

where \( P \) is the applied load and \( \Delta_p \) is the plastic displacement.

For power-law hardening materials based on Ramberg-Osgood material model, the plastic areas can be simplified by [36]

\[
A_p = \frac{N}{N+1} P\Delta_p \tag{2.36}
\]

\[
A_{sec} = \frac{1}{2} \frac{N-1}{N+1} P\Delta_p \tag{2.37}
\]
Replacing the plastic areas in Equation (2.32) and (2.33) with Equation (2.36) or (2.37), the solution of $J_p$ can be rewritten as

$$J_p = \frac{P\Delta_p}{B(W-a)}H'\eta$$

(2.38)

A parameter, $H'$ is used to simplify the definition of $\eta$ for different geometries, relies on LLD and CMOD records and is dependent on $N$. Six fracture mechanics geometries are studied in Chapter 3, those are compact tension specimen C(T), and C-shaped tension specimen, CS(T), single edge notch specimen in tension SEN(T) and bending SEN(B), double edge notch specimen in tension DEN(T), middle crack specimen in tension M(T). The values of $H^{LLD}$ and $H^{CMOD}$ for each fracture mechanics specimen are provided in Table 2.1, associated with the appropriate area to use.

The theoretical estimation of $J_p$ can be used EPRI (Electrical Power and Research Institute) solution, written as

$$J_p = \alpha\varepsilon_{p0}\sigma_p(W-a)h_l\left(\frac{P}{P_{LC}}\right)^{N+1}$$

(2.39)

where $h_l(a/W,N)$ is the dimensionless function of the normalised crack length and stress exponent, $P_{LC}$ is the limit load. The value of $h_l(a/W,N)$ and $P_{LC}$ can be found in textbook [11].

The EPRI estimate for the total value of $J$ can be obtained by the sum of elastic components based on small scale yielding (SSY) solution and plastic components by

$$J_{EPRI} = \frac{K^2(a_c)}{E} + \alpha\varepsilon_{p0}\sigma_p(W-a)h_l\left(\frac{P}{P_{LC}}\right)^{N+1}$$

(2.40)

### 2.4.3.4 Reference Stress Approach

The estimation of $J$ using reference stress approach [37] can be described by

$$J_{ref} = \frac{K^2(a_c)}{E} + \mu\varepsilon ref \left(\frac{E}{\sigma ref}\right)\left(\frac{K^2(a)}{\sigma ref}\right)^2$$

(2.41)
where \( \varepsilon_{\text{ref}} \) is the total strain at the reference stress \( \sigma_{\text{ref}} \), which can be obtained by Ramberg-Osgood equation. The value of \( \mu \) depends on plane stress/strain condition where \( \mu = 1 \) for plane stress and \( \mu = 1 - \nu^2 \) for plane strain respectively.

### 2.4.4 Time Dependent (Creep) Fracture Mechanics

Considering a remote load applied to a cracked body at elevated temperatures, initially, the material responded immediately with an elastic strain distribution, the stress exhibited a \( 1/\sqrt{r} \) singularity near the crack tip, which could be defined by \( K \). When the cracked body was subject to creep deformation, under a small scale creep (SSC) condition, that creep zone was relatively small compared to specimen size and crack length, \( K \) and \( J \) could continue to characterise the crack tip stress distribution outside the creeping area. However, as time proceeded, the creep zone grew until it was widespread. Under widespread creep conditions, steady state creep condition could prevail under constant loading as described in section 2.3. Under this circumstance, \( C^* \) could be used to characterise the steady state crack tip stress distribution. Between SSC and widespread creep, there was a transient field at a short time, creep stress under non-steady conditions could be characterised by \( C(t) \) or \( C_t \) [11].

#### 2.4.4.1 Steady State Creep Crack Growth

For creeping materials in a steady state creep crack growth region, a fracture mechanics approach has been developed after the establishment of \( J \) integral as the elastic-plastic fracture parameter. Goldman and Hutchinson [38] suggested a path-independent \( C^* \) to characterise crack growth in materials undergoing steady-state creep, which could characterise the stress and strain rate around the crack region. \( C^* \) was defined by replacing strain with strain rate, and displacement with displacement rate in the definition of \( J \) integral, as:

\[
C^* = \int_{r} \left( \dot{V} \right)_{\text{cr}} dx \quad \text{(2.42)}
\]

where \( \dot{\vec{i}} \) is the displacement rate vector and \( \dot{V} \) is the strain energy rate density, given by
\[ \dot{\epsilon}_{ij} = \int_{0}^{t} \tau_{ij} dt. \]  

(2.43)

where \( \dot{\epsilon} \) is the creep strain rate tensor. Considering a steady state creep occurs following a Norton’s creep law as described in Equation (2.14), the stress and strain rate can be defined using \( C^* \) integral, the equation is known as RR (Riedel and Rice [39]) equation, showing as:

\[ \frac{\sigma_{ij}}{\sigma_0} = \left[ \frac{C^*}{\sigma_0 \dot{\epsilon}^*} \right]^{\frac{1}{n+1}}. \]  

(2.44)

\[ \dot{\epsilon}_{ij} = \left[ \frac{C^*}{\sigma_0 \dot{\epsilon}^*} \right]^{\frac{n}{n+1}}. \]  

(2.45)

\( \sigma_{ij} \) and \( \dot{\epsilon} \) are the stress and strain rate tensors respectively; \( \sigma_0 \) is the normalising stress; \( \dot{\epsilon}^* \) is the normalising strain rate and \( r \) is the radial distance from the crack tip. The dimensionless constant \( I_n \) is evaluated by replacing strain-hardening exponent, \( N \) in Equation (2.28) and (2.29) with creep exponent, \( n \).

As derived from Equation (2.45), the time-dependent crack growth rate in viscous materials depends on the value of \( C^* \). Experimental studies prove that the CCG rate, \( \dot{a} \) (or \( da/dt \) ) correlates well with \( C^* \) in a power-law relationship, expressed as

\[ \dot{a} = D \phi. \]  

(2.46)

where \( D \) and \( \phi \) are the material constants and can be measured experimentally or determined from a model of the cracking process based on ductility exhaustion in a damage zone at the crack tip. In many materials, \( \phi \approx n / (n+1) \), which is based on a prediction under grain boundary cavitation models [11].

### 2.4.4.2 \( C^* \)-Integral Estimation Methods

For creeping materials, the analogy between steady state creep and plastically deforming described in Section 2.4.3 allows similar estimation methods for measuring \( C^* \) integral.
Experimental measurement of $C'$ also takes advantage of analogies with the $J$ integral, the $C'$ integral is related to the creep displacement rate, $\dot{\epsilon}$. In a power-law creeping material, $C'$ is given by

$$C' = \frac{P' H'}{B(W-a) \eta}$$  \hspace{1cm} (2.47)

where $H'$ and $\eta$ are geometry dependent factors which is similar to the parameters described in Equation (2.38).

The theoretical estimation of $C'$ in EPRI solution was similar with that of plastic component of the $J$ solution (as seen in Equation (2.39)) by replacing the variables $\varepsilon_{po}, \sigma_{po}, N$ and $P_{LC}$ with $\dot{\epsilon}, \sigma_0, n$ and $P_0$ respectively, in such a way, $C'_{EPRI}$ is defined by

$$C'_{EPRI} = \dot{\epsilon} \left( \frac{P}{P_0} \right)^{N+1}$$  \hspace{1cm} (2.48)

where $P_0$ is the normalising load related to the normalising stress $\sigma_0$, $\dot{\epsilon}$ is the normalising creep strain rate, $h_1(a/W,n)$ is the non-dimensional factor, dependent on the normalised crack length, $a/W$ and creep exponent, $n$.

An alternative reference stress method was also developed from analogous $J$-solution, which is consistent with the definition of $C'$, given by

$$C'_{ref} = \dot{\epsilon} \left( \frac{\varepsilon'(a)}{\varepsilon_{ref}} \right)^2$$  \hspace{1cm} (2.49)

where $\dot{\epsilon}$ is the reference creep strain rate which can be obtained by Norton’s law. This method was used to investigate $C'$ in the numerical calculation in the current research.

**2.4.4.3 Non-Steady State Creep Crack Growth**

The steady state solutions described in Section 2.4.4.1 and 2.4.4.2 are only applicable under widespread creep conditions when the stress and creep strain rate fields remain constant
throughout the creeping body. However, there is a transient creep, where creep damage is building up from $K$-controlled SSC to $C^*$-controlled steady state creep as time expands. The transition from short-time elasticity to long-time viscous behaviour was analysed by Riedal and Rice [39] in an elastic-creeping materials. A time and path dependent $C(t)$ fracture mechanics parameter is defined by a line integral similar to $C^*$, as

$$C(t) = \int_{\Gamma \to 0} \left[ \dot{V} \right]_{\text{cx}} \text{d}x$$ \hspace{1cm} (2.50)

where the notation $\Gamma \to 0$ is used to denote that the integral must be evaluated on a contour near the crack tip which is vanishingly small. In this case, elastic strain rate is relatively small compared with creep strain rate. Under non-steady conditions, the creep stress and strain rate within the transient creep zone are written by

$$\frac{\sigma_{ij}}{\sigma_0} = \left[ \frac{C(t)}{\sigma_0^2} \right] \text{HRR}$$ \hspace{1cm} (2.51)

$$\frac{\epsilon_{ij}}{\epsilon_0^2} = \left[ \frac{C(t)}{\epsilon_0^2} \right] \text{HRR}$$ \hspace{1cm} (2.52)

where $C(t)$ is a parameter that characterises the amplitude of the local stress singularity in the creep zone, which varies with time until approaching $C^*$ when widespread conditions are achieved. As $C(t) \to C^*$ the amplitude of these fields (denoted by RR fields after Riedel and Rice) is equal to those in widespread creep conditions as shown in Equation (2.44) and (2.45).

Consider a structure which is loaded at time $t = 0$, the crack tip fields are described by $K$ or $J$. For $t > 0$, the crack tip fields are described by Equation (2.51) and (2.52). However, at short time, only the very high crack tip stresses reduces to the values of HRR fields so that the surrounding stress distribution remains close to the values at $t = 0$. The surrounding stress distribution may be used to estimate the value of $C(t)$. The amplitude of SSC at crack tip field at short times is estimated from [39, 40]
If Equation (2.53) held at long time as $t \to \infty$, $C(t)$ would fall (as $C(t) > C^*$) to the value of $C^*$ at a transition time, $t_T$

$$t_T = \frac{K^2}{(n+1)E't}$$

A general equation is required to describe the value of $C(t)$ from intermediate times to transition time. Numerically evaluations have been performed on the shape of the relaxation curve of $C(t)$, and it is found the curve can be fitted by either an empirical interpolation formula [41] in an elastic-power-law creeping material as

$$C(t) \cong C^* \left(1 + \frac{t_T}{t}\right)$$

or by Anisworth and Budden’s equation [40] as

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

$C(t)$ parameter describes the instantaneous stages of the crack tip fields. Another parameter $C_i$ has been proposed and suggested for characterising crack growth prior to the steady state. This parameter can be experimentally measured from the specimen load line displacement and is also equal to $C^*$ as time expands to the widespread creep. The $C_i$ parameter under SSC conditions has an energy rate interpretation and it can be written by

$$\left(\frac{C_i}{ssc}\right) = \frac{P_i^*}{BW} \frac{dF}{d(a/W)}$$

where $P_i^*$ is creep load line displacement rate, $F$ is the stress intensity factor function as

$$F = \frac{KB}{P} \sqrt{W}$$

An expression has been proposed to describe the range of conditions from SSC to widespread creep, as
\[ C_t = (C_t)_{ssc} + C^* \]  \hspace{1cm} (2.59)

An alternative description of \( C_t \) at short times is derived from [42]

\[ C_t \approx \left[ 1 + \left( \frac{t}{\tau} \right)^{\frac{n-3}{n+1}} \right] C^* \]  \hspace{1cm} (2.60)

Compared with Equation (2.53), \( C_t \) does not have the same time dependency as \( C(t) \) as \( t \to 0 \).

The two methods have compared by numerical analysis with \( n = 5 \) and \( n = 10 \) in C(T) and SEN(T) specimens [43], the results can be seen in Figure 2.10 where \( \tau \) is the normalised time defined by \( \tau = t / t_T \). \( C_t \) has a smaller value than \( C(t) \) but the differences between them decrease with increasing \( n \) and time.

2.5 NSW Creep Crack Initiation and Growth Models

Under steady state creep condition, NSW model has been proposed to predict creep crack initiation (CCI) and growth (CCG) from uniaxial creep strain evolution and rupture properties of studied materials when detailed CCI and CCG data are not available, this model is known after its author (Nikbin, Smith and Webster) [44, 45]. This section starts with NSW CCG models for dealing with steady state creep crack growth rate with a consideration of several types. After that, NSW CCI model has been briefly introduced for predicting creep crack initiation.

2.5.1 Steady State CCG Models

In order to predict creep crack behaviour, the creep crack growth rate, \( \dot{\iota} \), should be determined using an appropriate fracture mechanics parameter. If the crack growth rate is modelled successfully, it can be used to assess the life of a serviced component that cannot be simulated from laboratory experiments. When creep deformation predominates the stress distribution, the steady state crack growth rate, \( \dot{\iota} \) is commonly characterised by \( C^* \), NSW model is introduced below, which is based on Equation (2.46) as \( \dot{\iota} \) is a function of \( C^* \) in steady state (secondary) stage.
2.5.1.1 NSW Model

In this model, it is assumed that creep process zone is expanding at a constant \( \dot{r} \) and remains undamaged until reaching a characteristic distance, \( r_c \) ahead of the crack tip. At that point, failure occurs where \( \tilde{\epsilon} \) (Equation (2.44)) attains its maximum value of unity, \( \tilde{\epsilon} \). The CCG rate using NSW model, \( \dot{\iota} \) can be predicted by

\[
\dot{\iota} \left( \frac{n+1}{\dot{\epsilon}_f^*} \right)^{n} = \frac{n+1}{\dot{\epsilon}_f^*} \left( \frac{1}{I_n \sigma_0^*} \right) \left( \frac{1}{A r_c} \right) \left( \frac{C^*}{I_n} \right)^{n+1}
\]

where \( A, n, \sigma_0 \) and \( \dot{\iota} \) can be obtained from Equation (2.14), \( r_c \) is the size of the creep process zone and \( \dot{\epsilon}_f^* \) is the appropriate crack tip ductility, \( \dot{\epsilon}_f^* \) is equal to the uniaxial creep ductility \( \dot{\epsilon}_f \) under plane stress, and \( \dot{\epsilon}_f / 30 \) under plane strain condition for most relevant engineering materials [46].

An alternative NSW model [47] assumes creeping materials undergo primary, secondary and tertiary creep deformation, the creep ductility can be obtained by using a power-law form average creep strain rate as shown in Equation (2.18). Additionally, \( t_c \) can be expressed as a function of stress as described in Equation (2.16). Assuming \( \sigma_0 \) is identical in Equation (2.16) and (2.18), the failure strain \( \dot{\epsilon}_f \) can be derived from these two equations, as

\[
\dot{\epsilon}_f = \dot{\epsilon}_{f0} \left( \frac{\sigma}{\sigma_0} \right)^{n_d-v}
\]

If \( n_d > v \), the ductility decreases with decrease in stress and if \( n_d = v \), \( \dot{\epsilon}_{f0} \) becomes independent on stress as \( \dot{\epsilon}_{f0}^* = \dot{\epsilon}_f^* \). Substitution of Equation (2.16), (2.18) and (2.62) into the exhaustion ductility based on creep strain data, the CCG rate using NSW-Rupture model, \( \dot{\iota} \) can be therefore written as

\[
\dot{\iota} = \dot{\epsilon}_{f0} \left( \frac{n_d + 1}{n_d + 1 - v} \right)^{\dot{\epsilon}_f^* \left( \frac{\sigma}{\sigma_0} \right)^{n_d-v}} \left( \frac{C^*}{I_n \sigma_0^*} \right)^{n_d-v} \left( \frac{1}{A r_c} \right) \left( \frac{C^*}{I_n} \right)^{n_d-v} \left( \frac{1}{A r_c} \right)
\]

\[
= \left( \frac{n_d + 1}{n_d + 1 - v} \right) \sigma_0^{n_d-v} \dot{\epsilon}_{f0}^* \left( \frac{C^*}{I_n \sigma_0^*} \right)^{n_d-v} \left( \frac{1}{A r_c} \right) \left( \frac{C^*}{I_n} \right)^{n_d-v} \left( \frac{1}{A r_c} \right)
\]

27
Note that if \( n_a = \nu \), Equation (2.63) can be reduced to the form of Equation (2.61).

### 2.5.1.2 Approximate NSW Model

To simplify the predictions for CCG rate with a consideration of most engineering materials, Equation (2.61) can be simplified to the form of Equation (2.64) [48], known as Approximate NSW (NSWA) Model.

\[
i' = \frac{\nu^{0.85}}{\varepsilon_f^*} \quad (2.64)
\]

Note that the dimension of \( i' \) is in [mm/h] and \( C^* \) is in [MPam/h]. The steady state of CCG rate is mainly dependent on creep ductility \( \varepsilon_f^* \). Here \( \varepsilon_f^* \) is again taken as the uniaxial failure strain, \( \varepsilon_f \) for plane stress condition and \( \varepsilon_f / 30 \) for plane strain condition. This was used in the current analysis to calculate the creep crack growth rate.

### 2.5.1.3 Modified NSW Model

A modified NSW model has been derived in [49] which considered the additional dependency of creep strain on the crack tip angle \( \theta \), and the power-law creep stress component, \( n \). Crack growth is assumed to occur where the ratio of equivalent strain to the multiaxial failure strain, \( \tilde{\varepsilon} (\theta, n) \), reaches a maximum value, denoted as \( \tilde{h}_n \). Therefore, the CCG rate using NSW-MOD model is predicted by

\[
i' = \frac{(n+1)i'}{\varepsilon_f^*} \left[ I_\sigma \sigma_{\nu}^* \right]^{\frac{1}{n+1}} \tilde{h}_n \quad (2.65)
\]

In Equation (2.65), the values of multiaxial creep ductility, \( \varepsilon_f^* \) may be estimated from the uniaxial failure strain, using a multiaxial strain factor (MSF), in such a way that

\[
\varepsilon_f^* = MSF \times \varepsilon_f \quad (2.66)
\]

The form of the MSF can be predicted using an appropriate model.
Based on a review of the CCG data for parent P91 [7] and 316 (L) material [50] as well as the life prediction of CCG using the NSW and NSW-MOD model, it is found that the NSW model would over-conservatively predict both the upper and lower bounds of the crack growth rate data while the NSW-MOD can reduce the conservatism [7].

2.5.2 NSW CCI Models

A steady-state situation has been considered so far. For most engineering ductile materials, an incubation period, prior to the onset of creep crack extension may exist when damage develops at the crack tip. Therefore it is important to incorporate the incubation time into creep crack growth prediction in practical use [47]. The initiation (or incubation) time, \( t_i \), can be defined as the time taken for the crack to extend over a small distance after loading, this is typically of the order of 0.2-0.5 mm, according to the sensitivity of the equipment used to detect the crack growth. The experimental definition of creep crack initiation time is regarded as the time as \( \Delta a = 0.2 \) mm or \( \Delta a = 0.5 \) mm of crack extension, denoted as \( t_{0.2} \) and \( t_{0.5} \) [51].

If \( \Delta a \) is the minimum crack extension that can be measured reliably, and \( \Delta a \) is sufficiently small that CCG rate is approximately constant for this amount of crack growth, then the initiation time, \( t_i \), may be estimated by

\[
t_i = \frac{\Delta a}{\dot{a}}
\]  
(2.67)

If the crack begins to grow at the rate of \( \dot{a} \) and reaches to secondary CCG rate, the lower and upper bound prediction of \( t_i \) can be estimated by

\[
\frac{\Delta a - \dot{a}}{\dot{a}} \leq t_i \leq \frac{\Delta a}{\dot{a}}
\]  
(2.68)

The initial creep crack growth rate \( \dot{a} \) for a given value of \( C^* \) is related to the steady state creep crack growth rate \( \dot{a} \) as [21]

\[
\dot{a} = \dot{a}
\]  
(2.69)
Employing Equation (2.68) and (2.69) to NSW CCG models, the upper and lower bound expression of \( t_i \) from the NSW equation as an example can be obtained in such a form as

\[
\begin{aligned}
\varepsilon_f^* & \quad \Delta a \\
\vartheta & \quad \left[ I_n \sigma_0^* \right]
\end{aligned}
\]  

(2.70)

Under conditions of plane stress and plane strain, predictions from Equation (2.70) will also depend on the influence of stress state on \( \varepsilon_f^* \), \( \varepsilon_{f0}^* \) and \( I_n \). Usually, predictions of initiation time under plane strain conditions (based on \( t^* \)) provides the lower bound value, the upper bound value is obtained under plane stress conditions based on \( t^* \). Compared with upper bound prediction, the lower bound provides the more conservative prediction (i.e. shorter initiation times predicted) [47].

2.6 Continuum Damage Mechanics

2.6.1 Introduction

In Section 2.3.2, the micro-structural mechanisms for creep deformation, i.e. diffusion and dislocation movement have been introduced. As a result of microstructural changes during deformation, the processes which may result in material degradation are known as creep damage. It usually occurs in metals and alloys after prolonged exposure to stresses at elevated temperatures. The creep damage is usually associated with the tertiary stage of creep, and brings about an accumulation and growth of microcracks and the onset of creep failure. Continuum damage mechanics originally proposed by Kachanov [52] and Rabotnov [53, 54], is introduced here which captures the overall response of the materials due to the accumulation of damage.

2.6.2 Continuum Damage Mechanics (CDM)

Under creep conditions, assuming the temperature is constant and the procedures are irreversible, the variables for the phenomenological description of creep damage are commonly characterised by a damage parameter, denoted as \( \omega \) [52]. The material is undamaged if \( \omega = 0 \) and fully damaged if \( \omega = 1 \), because there is no recovery, \( \omega \) is an increasing quantity. The Kachanov
damage model [52] was developed to describe brittle creep rupture under uniaxial creep conditions. In a creep rupture test, Kachanov introduced a continuous scalar field variable, $\psi$ and called it ‘continuity’, where $\psi = 1 - \omega$

$$
\psi = \frac{A_{\text{net}}}{A_0}
$$

(2.71)

where $A_0$ is the initial cross-sectional area and $A_{\text{net}}$ is the real carrying area, which decreases as a result of deterioration due to creep. Also, Kachanov defined the rate of change of the continuity as

$$
\frac{d\psi}{dt} = -C \left( \frac{\sigma}{\psi} \right)^n
$$

(2.72)

where $C$ and $n$ are material constants depending on temperature. The quantity $(\sigma / \psi)$ are interpreted as an effective stress, defined as $\tilde{\sigma}$.

Rabotnov [53, 54] assumed that damage could also bring the influence for the rate of strain and damage accumulation in a relationship of

$$
\dot{\varepsilon} = \left( \frac{1}{1 - \omega} \right)^q
$$

(2.73)

$$
\dot{\varepsilon} = \left( \frac{1}{1 - \omega} \right)^r
$$

(2.74)

where $A , C , n, k , q$ and $r$ are material parameters. If $\sigma$ could be simply replaced by $\tilde{\sigma}$, then $n = q$ and $k = r$, therefore, only two exponents are remained. Hayhurst [55] analyzed the experimental data for stationary creep in various temperatures and materials, and he noticed that in most situations, the below equation is sufficiently accurate.

$$
k = 0.7n
$$

(2.75)

Typical Katchanov-Robotnov continuum damage constitutive equations can be expressed in multi-axial form as follow
\[
\dot{\varepsilon} = \frac{(\sigma_e)^{n-1} S_{ij} t^m}{(1 - \omega)^n}
\]  
(2.76)

where \( A', n \) and \( m \) are material constants, \( S_{ij} \) is the deviatoric stress and \( \sigma_e \) is the equivalent (von Mises) stress.

The rate of damage accumulation is

\[
\dot{\omega} = \frac{\varepsilon \chi}{\omega}\dot{t}^n
\]  
(2.77)

where \( B', \phi \) and \( \chi \) are material constants, \( \sigma_r \) is the rupture stress is \( \sigma_1 \) is a function of the maximum principal stress, and \( \sigma_e \) is the equivalent stress:

\[
\sigma_r = \alpha \sigma_1 + (1 - \alpha) \sigma_e
\]  
(2.78)

where \( \alpha \) is a material constant which ranges from \( \alpha = 1 \) (maximum principal stress dominant) to \( \alpha = 0 \) (equivalent stress dominant).

### 2.6.3 Uncoupled Creep Damage

In uncoupled creep damage, the creep deformation, \( \dot{\varepsilon} \) is independent of the damage parameter. Under such circumstances, only \( \omega \) is an indicator of damage where the tertiary creep is not considered. Therefore in the uncoupled approach, failure at a given stress is conceded when an appropriate critical value of creep strain, \( \varepsilon_{\text{crit}} \) or alternatively the time for creep rupture, \( t_r \) is attained. The two conditions are called strain or rupture-time based method respectively [56]. Creep damage rate is:

\[
\dot{\varepsilon}_{\text{crit}} \text{ in the strain based uncoupled method}
\]  
(2.79)

\[
\dot{t}_r \text{ in the rupture time based method}
\]  
(2.80)

considering the definition of average strain rate as seen in Equation (2.17), Equation (2.80) can be rewritten as
If the average creep strain rate is applied to describe the creep deformation rate then the failure time corresponds to the attainment of the creep ductility when \( \varepsilon_{\text{crit}} = \varepsilon_f \).

The uncoupled approach has been used in the current research, it should be noted that it usually gives a conservative prediction on the rupture life.

### 2.6.4 Multiaxial Creep Failure Analysis

Under a uniaxial creep, the damage parameter, \( \omega \) can be defined by the ratio of the creep strain to the creep failure strain, as

\[
\omega = \frac{\dot{\varepsilon}}{\varepsilon_f}
\]  

(2.82)

The existing damage parameter associated with applied stresses can be used to describe the evolutions of creep strain and damage accumulations under uniaxial creeping conditions.

Due to the presence of stress triaxiality in the HAZ region, the creep damage is under multiaxial stress conditions. In the R5 assessment procedure, reheat crack initiation is described by an empirical ductility exhaustion concept [57], given by

\[
\omega = \int_0^t \frac{\dot{\varepsilon}}{\varepsilon_f^*} dt
\]  

(2.83)

where \( \dot{\varepsilon} \) is the instantaneous equivalent creep strain rate and \( \varepsilon_f^* \) is the multiaxial creep ductility which is a function of the stress and the equivalent creep strain rate [58].

Several models have been proposed [59] (Hull and Rimmer, 1959; McClintock, 1968; Rice and Tracey, 1969; Hayhurst, 1972; Cocks and Ashby, 1980; Cane, 1981; Manjoine, 1975,1982; Margolin et al. 1009; Ragab, 2002; Spindler, 1994; Spindler et al. 2001) to account for creep ductility under multiaxial stress states. These models are generally based on the growth of a void in a deforming medium and illustrate that the void growth rate is a function of stress triaxiality, which is defined as the ratio of the mean (hydrostatic) stress to equivalent stress \( (\sigma_m / \sigma_e) \). Two models are briefly introduced here, which are Cocks and Ashby [60] and Spindler [61] model.
Cocks and Ashby [60] proposed a model based on the constrained cavity growth mechanism. This model is appropriate for representing the multiaxial creep ductility of type 316H stainless steel in this research. It can be written as

\[
\frac{\dot{\varepsilon}_i}{\varepsilon_f} = \sinh \left[ \frac{2}{3} \left( \frac{n-1/2}{n+1/2} \right) \right] \sinh \left[ \frac{2}{n+1/2} \frac{\sigma_m}{\sigma_e} \right]
\]

In the R5 procedure [1], a ductility exhaustion approach is used to evaluate creep damage. A creep exponent independent model for the effect of multiaxial stress states on creep ductility is shown below which was proposed and developed by Spindler [61] on type 316 and type 304 stainless steels from biaxial creep data according to the following form as:

\[
\frac{\dot{\varepsilon}_i}{\varepsilon_f} = \exp \left\{ p \left( 1 - \frac{\sigma_m}{\sigma_e} \right) + q \left( \frac{1}{2} - \frac{3\sigma_m}{2\sigma_e} \right) \right\}
\]

where \( p \) and \( q \) are empirical parameter to be determined by multiaxial creep tests. It has shown that \( p = 2.38, q = 1.04 \) give the most satisfactory fit for type 316H stainless steel. All these would give similar results within a limited range of constraint.

### 2.6.5 The Effects of Continuum Damage

The effects of continuum damage have been considered in welded components. Since the cracks usually occur at the boundary between base and weld metal, in an attempt to describe deterioration phenomena in a class of welded component, Hayhurst [62] has applied different material’s creep properties on different regions of weldment, including base metal, weld metal, HAZ and strain affected region. This contribution is an attempt to cope with a complex microstructural deterioration within the studied component, to examine continuum damage behaviours. Extensive experimental support is essential in conducting such an analysis.

Another important parameter incorporated into the damage analysis is the equivalent stress, which can determine the damage parameter and the rate of damage accumulations according to Equation (2.69) and (2.82). The failure may initiatially occur at the positions with maximum effective streses. Therefore, the description of damage can be characterised on the macroscale.
via a change of material properties and the use of a fictitious effective stress. This method is a classical and widely used way to combine deterioration effects and stress-strain response.

According to Equation (2.82) (2.83), the damage is a function of creep ductility and stress-dependent creep strain rate. Also known that the stress distribution could depend on the loading conditions, creep deformation, geometrical constraint, in order to perform a comprehensive study on the damage, the effects of continuum damage have been studied in the present analysis, which has mainly discussed in Chapter 7, including considerations of load conditions, material inhomogeneity, creep stress relaxation, geometrical constraint and creep ductility.

2.7 Creep Cracks in 316H Austenitic Stainless Steels Welded Joints

AISI type 316 steel is an austenitic grade with nominal composition of 18 % Cr, 8 % Ni and 2 % Mo [63]. An additional of 2% molybdenum would improve corrosion resistance and is particularly apparent for pitting and crevice corrosion in chloride environments. 316 stainless steel family exists several subgrades, including 316L (lower carbon content for improved ductility), 316N (higher nitrogen levels for increased strength) and 316H (higher levels of carbon for improved resistance to intergranular corrosion) [64].

Type 316H Austenitic stainless steels are widely used in the fabrication of components operating at high temperatures due to their excellent resistance to creep and oxidation [65]. After a long time high temperature service exposure, creep cracks have been found in the HAZ of type 316H, austenitic stainless steel components containing thick section attachment weldments [66-70]. The mechanics by which these cracks were produced has been identified as reheat cracking, which is driven by the relaxation of post-weld residual stresses from values in excess of the yield stress to relatively low values, therefore, the relaxation of weld-induced residual stresses result in the accumulation of creep damage [71].

2.7.1 Reheat Crack Mechanism in 316H Austenitic Stainless Steels

Reheat-cracking in 316H austenitic stainless steel was first observed in components in 1988. Since then, over the period between 1988 and 1998, 261 incidences of reheat cracking have
been reported from the advanced gas-cooled reactor (AGR) in nuclear power plants [72]. These components usually contain thick section of weldments and operate at a temperature between 490 and 550 °C [73]. The reason for reheat cracks is mainly due to the creep dominated relaxation of the highly triaxial residual stresses induced by welding processes, the cracks result from the initiation and growth of grain boundary creep cavitation within the HAZ of the weldment and all of the cases are associated with low creep ductility [73]. Austenitic stainless steel, unlike the low alloy ferritic steel, remains in highly triaxial residual stresses after welding, the header components are generally not stress relieved before service [67, 74, 75].

In addition to the residual stresses introduced during welding, a range of complex microstructures are produced in austenitic stainless steel weldments as the temperature history varies in the HAZ. Very high temperatures experienced in the fusion zone decrease with distance towards the base materials. A given location in the heat affected zone (HAZ) therefore exhibits a sequence of ascending temperature excursions as the weld head approaches and a string of decreasing excursions on cooling. Many factors such as the existence of grain coarsening, the presence of intra- and inter-granular carbide precipitates, high levels of plastic strain and grain boundary impurity element segregation may result in detrimental results in the HAZ region of weldments [76]. The influence of thermo-mechanical history on microstructural creep properties of 316H austenitic stainless steels are not systematically studied here. The review of microstructural effects on reheat cracking can be seen in Ref [77].

It is recognised that a critical combination of residual stresses, microstructure and temperature is required to promote reheat cracking [73]. As the main mechanisms of reheat cracking, residual stresses have been described in the next.

**2.7.2 Welding Residual Stresses**

**2.7.2.1 Introduction**

Residual stresses are those retained within a body when no external forces are acting. They can have a detrimental effect on component’s structural integrity and performance in fracture, creep and fatigue deformation [78]. Reheat cracking in 316H austenitic stainless steel weldments was attributed to the presence of the triaxial residual stresses in the HAZ and their relaxation driven
by creep deformation. Therefore residual stresses must be accurately quantified to determine their influence on failure mechanisms. Investigation of residual stresses in structural components should combine various factors including [79]:

1. The welding procedure (heat input, number of passes, deposition sequence etc.)
2. The weldment geometry (including the effect of structural restraint)
3. Local geometric features (weld root, toe and capping passes)
4. The weld and base material properties
5. Effects of heat treatment
6. Microstructure changes

2.7.2.2 Residual Stresses in Components

Residual stresses play a significant role in the structural integrity assessment of welded components. The residual stress field could be introduced by a non-uniform plastic deformation when components were joined together by fusion welding or deformed mechanically, being set up across the thickness of the component. At long service time under high operational temperature and stress, engineering structures may fail by creep mechanism [80].

The distribution of residual stress are primarily affected by existing stress, the geometry of welded joints, post-weld heat treatment (PWHT), material properties and service degradation. A brief investigation on some factors has been reviewed here. The detail study on material inhomogeneity between base and weld metal and ex-service degradation has been stated in Chapter 7.

The pre-existing residual stresses could be introduced during manufacturing operations such as casting, forging, rolling, heat treatments, quenching, straightening and carburisation. Significant pre-welding fabrication processes can also bring significant residual stresses during flame, plasma or laser cutting, bending, machining, jigging and correction of misalignment. The pre-existing residual stresses should be accompanied when the residual stresses in welded structures are being evaluated [2].

In addition, a wide range of geometries of welded joints will affect the performance of transverse and longitudinal through-thickness residual stress distributions [80]. For different joint
geometries, the maximum transverse stress usually occurs at the weld toe in terms of tensile stress and then becomes compressive, achieve a minimum value at the half-thickness and tensile again at 75% thickness. The longitudinal residual stresses are tensile throughout the thickness. The results indicate that higher residual stresses are obtained at the weld toe with increasing heat inputs; thicker joint can make higher transverse residual stresses.

PWHT can significantly reduce the magnitude of residual stresses, Stacey et al. [80] and Josefson [81] illustrated that stress relief brought a significant decrease of axial, hoop and radial stress with a temperature-controlled treatment. However, although PWHT is an effective method to release the residual stress, the consideration should be also taken on the cost, excessive distortion or material sensitisation [82].

Residual stress field within the welding area plays the significant role in crack initiation and propagation. For cracks growing perpendicular to the weld, which is subjected to weld induced longitudinal tensile stresses of at least 0.5 of yield stress, crack growth rates have been measured to be between three and seven times larger than those found in parent plate [83]. The tensile residual stress could increase the stress local to a crack tip and result in a subsequent failure in the service [84].

The welding residual stresses can be relaxed after service exposure for a period of time. If a specimen containing residual stresses is maintained at high temperature with no primary load, the residual stresses will relax significantly due to creep as the elastic/plastic strains in the specimen are transferred to creep strains. Under high constraints and low ductility, the relaxation turns into the development and growth of cracks. This is invariably observed in large welded components. In addition if there are primary loads acting on the tested specimens, the residual stress effect can enhance the stresses combined with that primary load, and the introduction of residual stresses leads to an acceleration of the evolution of creep strain during the early stage of loading [85].

### 2.7.2.3 Residual Stress Measurements

It is important to have reliable methods for the measurement of residual stresses and to understand the level of information they can provide [86]. In general, residual stress measurement techniques can be divided into destructive and non-destructive methods.
Mechanical stress measurement generally belongs to destructive techniques, which rely on monitoring of changes in component distortion, either during the generation of the residual stresses, or afterwards, by deliberately removing material to allow the stresses to relax [86]. This types of methods include hole drilling strain gauge [87], deep hole drilling [88], crack compliance and contour techniques [89], the BRSL method [90] and Sach’s boring method [91]. Deep hole drilling method measures through-thickness residual stresses semi-destructively and the method can be applied to complex shaped components.

Diffraction methods for measuring residual stresses in a non-destructive way are very popular as they can non-destructively determine the stress state inside a sample and can provide a fast requisition and good spatial resolution. Those techniques have the capability of determining the deep line stress profile using a 3D stress tensor [92]. The main limitations are the complexity, size and weight of component which can be investigated, also the cost of using a neutron facility and the determination of local stress-free lattice parameter. Frequently used methods are X-ray diffraction, neutron (atomic strain gauge) and ultrasonic diffraction [93].

**2.7.2.4 Residual Stress Simulations**

FE analyses of welding residual stresses and reheat-cracking damage began in the early of 1970s with the pioneering work of Hibbit and Marcal [94], Friedman [95], Westby [96] and Andersson [97]. After more than 40 years of development, computational welding processes have now been carried out for a large number of service plant components [66-69, 98] and most of the physical phenomena involved in the welding process can now be addressed via FE programs such as ABAQUS [99]. This software provides a thermo-elastic-plastic solution based on the heat equilibrium equation for a solid body with heat sources. The code can consider the temperature dependence of all thermal properties and non-linear variation of convective and radiative boundary heat dissipation. The latent heat due to phase changes can also be taken into account in the model [67]. In application, multi-pass welding process may be modelled on a run by run basis to complete the join. Thus, computational welding process can be used to simulate the generation of residual stresses.

In practice however, there are still aspects of welding that have made the rigorous analysis of welds a challenging procedure. At the macroscopic level a weld can be considered to be a
thermo-mechanical problem of computing transient temperature, displacement, stress and strain. At the microscopic level, it can be considered to be a metal physics problem of computing the phase transformations including grain growth, dissolution and precipitation [100]. Due to the complexity of welding process, FE work usually idealises the material properties and decreases the number of weld deposits in order to reduce computational requirements, those simplifications and idealisations may become a source of uncertain prediction. In the current study, the residual stresses are simulated using a simplified method without considering a weld simulation, the results have been compared with Smith et al. and Hayhurst’s work [66-70] using weld simulation method. This method is introduced by defining appropriate stresses type and boundary conditions to reflect the pseudo residual stress distribution, the results are discussed in Chapter 7. Generally, it gives good predictions when compared with the measured data for the studied components.

2.8 Summary

As a consequence of this literature review chapter, some key points have been discussed.

1. Fundamental concepts of elastic-plastic deformation and the mechanisms of creep have been reviewed. The constitutive equations have been considered for plastic and creep deformation. Creep parameters, including $\dot{\varepsilon}, \varepsilon_f, \varepsilon_f$, and $t_f$, have been proposed and discussed for type 316 stainless steel. The principles of elastic, elastic-plastic and time dependent (creep) fracture mechanics have been briefly described. The single parameters, including $K, J, C^*, C(t)$ and $C_i$ have been introduced for characterising crack tip stress and strain condition, together with its application on crack initiation and propagation. Methods for evaluating $J$ and $C^*$ using experimental measurement based on $\eta$ factor, EPRI and reference stress method have been derived.

2. NSW models have been presented for predicting creep crack initiation and steady state crack propagation in terms of creep damage accumulation in a process zone ahead of the crack tip using fracture mechanics concepts.
3. Classical continuum damage mechanics which has evolved as a means to analyse the effect of material deterioration under creep conditions has been briefly reviewed. The effects of damage evolutions have been introduced and are further discussed in Chapter 7.

4. The mechanisms of reheat cracking in 316H stainless steel have been reviewed. Key parameters which affect the susceptibility to the reheat cracking are residual stress, microstructure and temperature. As the main mechanism of reheat cracking, welding residual stress effects have been described. A further analysis of welding residual stresses is performed in Chapter 7.
### 2.9 Tables

Table 2.1: Definition of $H$ and $A$ for each fracture mechanics specimen type [36, 101]

<table>
<thead>
<tr>
<th>Specimen Type</th>
<th>$H^{LLD}$</th>
<th>$H^{CMOD}$</th>
<th>Associated Area, $A$</th>
</tr>
</thead>
<tbody>
<tr>
<td>C(T)</td>
<td>$N/(N+1)$</td>
<td>$N/(N+1)$</td>
<td>$A_p$</td>
</tr>
<tr>
<td>CS(T)</td>
<td>$N/(N+1)$</td>
<td>$N/(N+1)$</td>
<td>$A_p$</td>
</tr>
<tr>
<td>SEN(T)</td>
<td>$N/(N+1)$</td>
<td>$N/(N+1)$</td>
<td>$A_p$</td>
</tr>
<tr>
<td>SEN(B)</td>
<td>$N/(N+1)$</td>
<td>$(2L/W)N/(N+1)$</td>
<td>$A_p$</td>
</tr>
<tr>
<td>DEN(T)</td>
<td>$1/2(N-1)/(N+1)$</td>
<td>$1/2(N-1)/(N+1)$</td>
<td>$A_{sec}$</td>
</tr>
<tr>
<td>M(T)</td>
<td>$1/2(N-1)/(N+1)$</td>
<td>$1/2(N-1)/(N+1)$</td>
<td>$A_{sec}$</td>
</tr>
</tbody>
</table>

### 2.10 Figures

Figure 2.1: Typical stress-strain curve in Ramberg-Osgood material model
Figure 2.2: Deformation mechanics map for type 316 stainless steel [15]

Figure 2.3: Application area in deformation mechanics map for type 316 stainless steel [17]
Figure 2.4: Schematic creep behaviour of a material subjected to a constant stress

Figure 2.5: Fracture mechanism map for type 316 stainless steel [22]
Figure 2.6: Definition of secondary and average creep strain rate

Figure 2.7: Two types of creep ductility behaviours as creep ductility is (a) a function of net section stress (P91-PM at 600 °C) [25]; (b) not a function of net section stress (316L(N) at 550 °C) [25]
Figure 2.8: Definition of $J$-integral based on an arbitrary contour around the crack tip

Figure 2.9: Load-displacement response and definition of (a) plastic displacement $\Delta_P$ and plastic area $A_P$, (b) plastic secant area $A_{sec}$, under the load-plastic displacement curve

Figure 2.10: Comparison of $C(t)$ and $C_i$ for C(T) and SEN(T) specimens with (a) $n = 5$ and (b) $n = 10$ [43]
Chapter 3

Evaluation of Fracture Mechanics Parameters for Weldment Geometries

3.1 Introduction

In order to characterise the fracture or creep behaviour of engineering components, it is important to provide acceptable estimations of the $J$ and $C^*$ fracture mechanics parameters. The observations have shown that in some engineering components operating at elevated temperatures, the crack may often initiate in or near the heat-affected zone (HAZ) as a result of creep damage accumulation in this region, thus limiting the effective components’ lifetime [5].

The use of weldments usually requires an over-matched condition where the strength of weld metal is higher than base metal. This condition can promote gross section yielding of the base metal, facilitate a shift of plastic zone from higher strength of weld metal to base metal with lower strength, and therefore, reduce the probability of structural failure stemmed from an undetected weld defect in operation [102-105]. In the welded structure, the complex interactions between global loading and local crack tip driving forces can be significantly influenced by the mismatch factor, $M$, which is defined as the ratio of the weld metal (WM) to the base metal’s (BM) yield strength [106], as

$$M = \frac{\sigma_{0WM}}{\sigma_{0BM}}$$  \hspace{1cm} (3.1)

The investigations on mismatch ratio is essential for mismatched components in order to accurately evaluate the fracture mechanics parameters, namely $J$ and $C^*$.

A major framework for evaluating the non-linear component of the $J$ -integral and $C^*$ parameter for common fracture mechanics geometries upon experimental measurement is based on the plastic work according to the load-displacement behaviour. The theory is built by Sumpter and Turner in 1976 [107] who expanded upon the contributions from early work of
Rice [32], Merkle and Corten [108] and firstly defined a non-dimensional parameter, $\eta$ in a form of

$$\eta = \frac{J^* B(W - a)}{A_p}$$  \hspace{1cm} (3.2)

Here, $\eta$ is assumed to be a function of the specimen geometry, which is sensitive to normalised crack length but is insensitive to load. Numerous investigations [36, 102-105, 109-115] have shown that the $\eta$ factor based on either load-line displacement (LLD) or crack mouth opening displacement (CMOD) measurement can provide an accurate and effective toughness measurement for different fracture geometries. Since the requirement for those users who rely on $\eta$-based methods for $J$ and $C^*$ measurement, a number of standardised codes for fracture toughness measurement have already been developed, such as ASTM 1290 [116] and ASTM 1820 [117]. An alternative method to evaluate $J^*$ is to use load separation analysis to evaluate $\eta$ for conventional fracture specimens. This theory was initially proposed by Paris et. al. [118] but is not applied in the present study.

Standardised measurement for crack growth resistance testing usually propose compact tension, C(T) specimens containing deep cracks as $a/W \geq 0.45 - 0.5$. C(T) specimen is often employed in fracture tests because it ensures the crack growth under a high crack-tip constraint condition with a limited-scale of plasticity [104]. Also, other laboratory geometries have been developed and proved to be useful for investigating various crack tip constraint configurations on fracture toughness. For example, in high pressure piping systems, the near-surface cracks occurred during welding fabrication require an investigation of low constraint crack configurations [104, 105, 111, 119]. Based on application requirement, a systematic review of the fracture toughness parameter, $\eta$ factor is performed for a range of laboratory geometries in the present study. The reliability of fracture toughness parameters, $\eta$ factor, for each type of geometry is discussed by considering both $LLD$ and $CMOD$ records as well as comparing with the results available in the literatures.

Experimental procedures for evaluation of $J$ and $C^*$ parameter based on $\eta$ factor have been introduced in Section 2.4.3.3 and 2.4.4.2. The evaluation procedures as well as all of standards
mentioned above are limited to homogeneous materials. However, as for welded components, the mismatch between weld metal and base metal would influence the mechanical behaviour of the laboratory specimens from the load-displacement response and further affect the crack tip stress fields [103, 105]. The use of over-matched conditions would overestimate the actual toughness of the welded joints whereas an under-matched condition would result in an under prediction. A comprehensive study on the influence of weldment material mismatch on the $\eta$ factor is made. Also, it is important to use appropriate data from validated test methods where available to use in FE analysis.

The overall objective of this work is to expand previous investigations on $J$ or $C^*$ estimation to welded specimens using -based procedures [36], stressing on the mismatch effect under a range of crack lengths. Previous research on $\eta$ -based methods deriving $J$ and $C^*$ fracture mechanics parameter have been reviewed and compared to the present results for weldment analysis where applicable. The current calculations considered the different effects on $\eta$ factors from $LLD$ and $CMOD$ analysis with respect to geometric and material variables. These variables comprise of a range of crack sizes (as a measure of $a/W$), elastic-plastic material properties, including different hardening exponents $N$, weld width ratios $h/W$ and stress states condition $P\varepsilon, P\sigma$, with particular emphasis placed on the sensitivity of mismatch ratios $M$. The $\eta$ solutions obtained will improve the evaluation of $J$ and $C^*$ in accordance with ASTM E1820 [117], E647 [120] and E1457 [51] standard procedures used for fracture, fatigue and creep crack growth analyses, respectively. The results from this work have been implemented in ASTM E1457-2014.

### 3.2 Analysis

#### 3.2.1 Numerical Analysis

Numerical analysis has been performed to calculate $\eta$ values based on the evaluation procedure described in Section 2.4.3.3 and 2.4.4.2. According to Equation (3.1), $\eta$ is obtained from the plastic component of $J$ and area under the load-displacement curve. As $J$ is evaluated in an elastic-plastic FE analyses, $J_p$ is obtained according to Equation (2.30) and (2.31). The plastic area $A_p$ is obtained from Equation (2.36) and (2.37). Detailed FE analyses have been performed
for six geometries containing a weld region in the crack growth area with different $h/W$ and mismatch ratios. The examined geometries include compact tension specimen C(T), C-shaped tension specimen, CS(T), single edge notch specimen in tension SEN(T) and bending SEN(B), double edge notch specimen in tension DEN(T) and middle crack specimen in tension M(T). Schematic illustrations of the fracture mechanics geometries examined are shown in Figure 3.1 in which $h/W$, specimen length, width, and radius are defined for each geometry. The specimen dimensions chosen for FE analyses are based on ASTM E 1457 [51], as listed in Table 3.1. The analysis matrix has considered the effect of $a/W$, $h/W$ and $P\varepsilon$, $P\sigma$ and elastic plastic properties for the recommended geometries. FE analyses for each geometry and condition have been performed to calculate $\eta$ as a function of $M$ for a range of $a/W$ based on LLD or CMOD responses.

All analyses were conducted in ABAQUS v6.12 [99] using four-noded plane stress (CPS4) and plane strain (CPE4H) elements. Figure 3.2 shows the FE models constructed for each geometry showing load and boundary conditions. It should be noted the loading conditions in SEN(T), DEN(T) and M(T) are using surface tension methods which can avoid any stress concentration on a single pulled node. After comparison with another method where the tension surface was constrained with a reference node, it was found that the results are exactly same, proving that this method can be used in the experimental analysis. Due to the geometric symmetry of the specimens only half of the full geometry were modelled with appropriate symmetrical boundary conditions in SEN(T), SEN(B), C(T) and CS(T), and a quarter of mesh were modelled to represent M(T) and DEN(T). The design of near crack region is similar in all the different geometries as shown in Figure 3.2.

3.2.2 Modelling the Geometries and Material Behaviour

Evaluation of $\eta$ factors in FE solutions requires the numerical solution of the $J$ parameter and the load-displacement response prediction. In this study, the Ramberg-Osgood material model has been utilised as can be seen in Equation (2.6). For power-law creeping material, the steady state creep strain rate is based on Norton’s law as shown in Equation (2.14).
The FE analyses have taken into account a wide range of strength mismatch values: 50% under-match, even-match, 50% and 100% over-match \((M = 0.5, 1.0, 1.5 \text{ and } 2.0)\). Since the HAZ is not considered due to its complicated properties, weld fracture specimens under this study can be considered as a bi-material model. The yield stress of weld metal has been fixed at \(\sigma_{0WM} = 660\) MPa and the homogeneous base metal’s yield strength, \(\sigma_{0BM}\), was varied in the analyses. Power-law hardening exponents of \(N = 5, 10\) and \(20\) were used in the analyses. Therefore, the sensitivity of the \(\eta\) solution to the mismatch ratio \(M\), hardening exponent \(N\) under \(P\varepsilon\) and \(P\sigma\) has been examined.

### 3.3 Review of the Existing \(\eta\) Factor Solutions in the Literature

A review of the existing analytical and numerical solutions of the \(\eta\) factor for a range of specimen geometries in the literature is summarised in Table 3.2. As seen in this table the majority of the values have been investigated for homogeneous materials. Therefore further investigations are required to consider the influence of inhomogeneity on solutions. Some discrepancies in the values of \(\eta\) have been observed between different sources, which may be attributed to different material properties, mesh designs, boundary conditions and FE packages employed. The data from different resources have been compared in the following study. The literature survey results [36, 102-106, 109-115, 121-132] shown in Table 3.2 have been included where possible to those obtained in this study. In all cases the results fall within the bounds discussed for the present analysis. Although there is no clear correlation observed when they are compared individually.

### 3.4 Numerical Solutions of the \(\eta\) Factor

In this section, the calculated \(\eta\) factors for six fracture mechanics geometries are individually presented and discussed in terms of normalised crack lengths \((a/W)\), \(h/W\) ratios, hardening exponents \((N)\), mismatched ratios \((M)\) under plane strain and plane stress \((P\varepsilon, P\sigma)\) conditions. \(\eta\) solutions for weld mismatch ratio of \(0.5 \leq M \leq 2\), \(h/W\) of \(0.05 \leq h/W\) (or \(2W\)) \(\leq 0.10\),
hardening exponent of $5 \leq N \leq 20$, normalised crack length of $0.1 \leq a/W \leq 0.7$ under plane strain and plane stress conditions are individually examined and tabulated in Table A.1-Table A.11 in Appendix A. The effects of each variable have been briefly introduced here.

### 3.4.1 Stress State ($P\sigma, P\varepsilon$) and Hardening Exponent ($N$) Effects

According to the $\eta$ results in Table A.1-Table A.11, for a wide range of $N$ under both $P\sigma$ and $P\varepsilon$, the scatter of $\eta$ in most cases is within $\pm 20\%$ for all geometries when mismatch factor is between $0.5 < M < 2.0$. This confirms previous work [131] for C(T) specimens. Therefore, it can be concluded that stress state and hardening exponent would bring limited contributions to the variations of $\eta$, which has proved to be satisfied with any mismatched conditions.

### 3.4.2 Weld width Effects

Due to the insensitivity in $N$ under $P\sigma$ and $P\varepsilon$, a detailed study looking at the influence of $h/W$ on $\eta^{LLD}$ and $\eta^{CMOD}$ calculations has been performed as $N = 10$ under plane strain conditions. The trends of $h/W$ variations on $\eta^{LLD}$ and $\eta^{CMOD}$ are examined on each fracture mechanics geometry for three overmatched values such that $M = 1.25, 1.5$ and $2.0$ and three selected crack lengths ranging from shallow crack of $a/W = 0.2$ (0.35 for C(T) specimen) to medium crack of $a/W = 0.5$, and a deep crack of $a/W = 0.7$. The results are shown in Figure 3.3-Figure 3.13.

In all cases as shown in Figure 3.3-Figure 3.13, parabolic profiles can be seen where $\eta$ decreases as $h/W$ increases until a minimum value is attained, beyond which $\eta$ increases and tends towards the homogeneous value ($h = 0$). Based on the analysis in this work the point of inflection in different crack length varies and may be dependent on specimen geometry and material properties. The point of inflection occurs at lower values of $h/H$ as crack length increases, indicating that deformation of the stronger (weld) material becomes more significant than the surrounding softer (base) material. The minimum value for $\eta$ with respect to $a/W$ occurs for $M = 2$ for all the geometries. The maximum variations of $\eta^{CMOD}$ and $\eta^{LLD}$ are calculated and shown in Figure 3.14. The average deviation from the mean value is about $\pm 20\%$ for $\eta^{CMOD}$ and $\pm 50\%$ for $\eta^{LLD}$. Clearly $\eta^{CMOD}$ would be the preferred method for measurements.
in these cases as they show a lower scatter. As shown in Figure 3.3-Figure 3.13 for plane strain conditions, generally, if the weld width is limited to \(0.05 \leq h/W \) (or \(2W\)) \(\leq 0.10\), the variations of both \(\eta^{\text{CMOD}}\) and \(\eta^{\text{LLD}}\) due to \(h/W\) reduce to around 15% less than that of homogeneous materials. This is also the case for plane stress conditions.

### 3.4.3 Crack Length and Mismatch Ratio Effects

Having considered the scatter band for \(\eta\) with respect to \(N, h/W\) and \(P\varepsilon, P\sigma\), the main objective was to only consider the sensitivity of \(\eta\) to \(M\) and \(a/W\). The analysis is comprehensively considered for each geometry in the following section. The upper/lower bound values of \(\eta^{\text{LLD}}\) and \(\eta^{\text{CMOD}}\), based on the linear fits and the level of deviations from the means are then tabulated in Table B.1-Table B.6 in Appendix B for both homogeneous and a range of mismatch values of \(M = 0.5, 1.5\) and 2.0. These values are needed to derive appropriate \(J\) and \(C^*\) estimations for weldment specimens as discussed in test standards [51, 117, 120]. Therefore on the basis of the analyses presented, recommendations for mean and lower/upper bound values of \(\eta\) factors in welded components are made for each geometry examined.

### 3.5 Results

#### 3.5.1 Compact Tension C(T)

Standard \(J\)-integral and \(C^*\) testing procedures [51, 133], typically involve C(T) specimens which are the most widely used creep fracture mechanics geometry. This standard testing specimen ensures highly crack tip constraints and thus provides conservative toughness estimates for the materials of interest. For C(T) specimen, the LLD is measured from the clip gauge attached to the stepped notch as shown in Figure 3.1, thus, the \(\eta\) value estimated from LLD is identical with that from CMOD.

The \(\eta^{\text{LLD}}\) results obtained from the current study on a homogeneous material \((M = 1)\) for a range of crack lengths have been compared to those available in the literature [36, 51, 115, 134] for different \(N\) values under \(P\varepsilon\) and \(P\sigma\) conditions. The results can be seen in Figure 3.15. Generally,
good agreement is observed from different resources as the normalised crack length is between
0.45 ≤ a/W ≤ 0.7 where \( \eta^{LLD} \) is relatively insensitive to \( N, P\varepsilon, P\sigma \). As \( a/W < 0.45 \), the scatter band of \( \eta^{LLD} \) expands since \( \eta^{LLD} \) would be sensitive to \( N \) and \( P\varepsilon, P\sigma \) in the shallow crack.

The mean value of \( \eta^{LLD} \) obtained from numerical analyses is shown in Figure 3.16. The error bars indicate maximum range derived from 12 individual results with various \( N \) and \( h/W \) under \( P\varepsilon, P\sigma \) conditions. It has been found that for the conditions examined, for a given crack length in a range of 0.45 ≤ a/W ≤ 0.7 and mismatch ratio, the \( \eta^{LLD} \) values differ by a maximum of 12% from the mean values. Deviations from the mean in under-matched conditions generally show a larger discrepancy compared with that in over-matched conditions.

\( \eta^{LLD} \) is relatively insensitive to crack length for 0.45 ≤ a/W ≤ 0.7, but decreases significantly for \( a/W < 0.45 \) at all values of \( M \). As an approximately rule for crack length between 0.45 ≤ a/W ≤ 0.7, the variation due to mismatch is about ±15% above and below the base material. As \( \eta^{LLD} = 2.20 \) for homogeneous material, it can be taken as 2.6 for under-matched condition ( \( M = 0.5 \) ) and 2.0 ( \( M = 2.0 \) ) for over-matched conditions.

The average of \( \eta^{LLD} \) for a given \( M \) has been calculated for 0.45 ≤ a/W ≤ 0.7 as shown in Figure 3.17. To describe the influence of mismatch ratio on \( \eta^{LLD} \), a trend line has been fitted to the data in Figure 3.17, and is described by

\[
\eta^{LLD} = 2.71 - 0.38M \pm 0.20 \quad \text{for} \ 0.45 \leq a/W \leq 0.7
\]

(3.3)

### 3.5.2 C-Shape Tension CS(T)

CS(T) which is firstly introduced by Kendall and Hussain [135] has been used in evaluating structural components such as pressure vessels due to its highly triaxial constraint at the crack tip [136]. However, the investigation on \( \eta \) in CS(T) is very limited. In Ref [36], the effect of \( \eta \) has been discussed in homogeneous materials with different normalised crack length, \( N \) under \( P\varepsilon \) and \( P\sigma \) condition. The comparison between the current results and those data on homogeneous materials is made in Figure 3.18.
As seen in Figure 3.18 (a), the results of $\eta_{\text{CMOD}}$ lies within an agreeable region compared with those in Ref [36]. $\eta_{\text{CMOD}}$ is relatively insensitive to $N$, $P\varepsilon$, $P\sigma$ in the studied range of crack length, especially in deep crack. In Figure 3.18 (b), current study reveals that $\eta_{\text{LLD}}$ remains a constant value of around 2.26 and is independent on $N$, $P\varepsilon$, $P\sigma$. It should be mentioned that as $a/W = 0.2$, there is a large discrepancy with previous results in Ref [36], but current results are satisfied with the approximate fitting curve as illustrated in Ref [137].

The effects of crack length and mismatch on $\eta_{\text{CMOD}}$ and $\eta_{\text{LLD}}$ are shown in Figure 3.19 (a) and (b), respectively. Same as homogeneous materials, for a given crack length, $\eta_{\text{CMOD}}$ and $\eta_{\text{LLD}}$ are relatively insensitive to $N$, $h/W$ and $P\varepsilon$, $P\sigma$. The standard deviations are shown to be about ±10% from the means.

As seen in Figure 3.19 (a), the average $\eta_{\text{CMOD}}$ decreases with increasing normalised crack length, the variation due to $a/W$ are within 10% from the means when $0.3 \leq a/W \leq 0.7$. The $\eta_{\text{LLD}}$ results demonstrated in Figure 3.19 (b) shows that $\eta_{\text{LLD}}$ initially increases and then deceases. In the range of $0.3 \leq a/W \leq 0.7$, $\eta_{\text{LLD}}$ only changes in a narrow range of 8%. Therefore, the effect of $M$ on $\eta_{\text{CMOD}}$ and $\eta_{\text{LLD}}$ in the studied range of normalised crack length can be calculated. The variations due to mismatch ratio is within ±10% above and below the base material. The best linear fits for $\eta_{\text{CMOD}}$ and $\eta_{\text{LLD}}$ with respect to $M$ are shown in Figure 3.20 and can be expressed as

$$\eta_{\text{CMOD}} = 4.27 - 0.49M \pm 0.38 \quad \text{for} \quad 0.3 \leq a/W \leq 0.7$$  \hspace{1cm} (3.4)

$$\eta_{\text{LLD}} = 2.50 - 0.29M \pm 0.12 \quad \text{for} \quad 0.3 \leq a/W \leq 0.7$$  \hspace{1cm} (3.5)

It is clear that $\eta$ based on LLD record is relatively insensitive to crack length in the range of $N$, $h/W$ and $P\varepsilon$, $P\sigma$ studies and therefore, provides a more reliable solution for deriving $J$ and $C^*$.

### 3.5.3 Single Edge Notch in Tension SEN(T)

Previous research efforts indicate that the cracked SEN(T) specimen is very useful for investigating low crack tip constraint effects [104, 111, 138-141]. The use of SEN(T) specimen to obtain information of pipelines and cylindrical vessels has received a great interest since the
SEN(T) specimen is safe to be used in low constraint engineering applications due to the similar geometry constraint found in SEN(T) and real pipeline defects.

Evaluations of $\eta$ factors on homogeneous SEN(T) specimen are briefly provided and reviewed with previous work as seen in Figure 3.21. Consideration is taken on the results of $\eta$ using CMOD records firstly. The previous research concluded that the $\eta^{CMOD}$ is insensitive to the $N$, and $P\varepsilon, P\sigma$ [36, 104, 105, 109, 111]. Figure 3.21 shows a full set of results from the current analysis which is in agreement with previous findings covering different conditions mentioned above. In present study, $\eta^{CMOD}$ is around 1.0, this results are consistent to the simulation curve by Kim and Budden [111].

In Figure 3.21 (b), the distribution of $\eta^{LLD}$ solution is not as simple as that for $\eta^{CMOD}$. Based on the previous findings, the accuracy of the $\eta$ factors for very shallow crack ($a/W \leq 0.4 - 0.5$) was strongly questioned, because $\eta^{LLD}$ is sensitive to the specimen thickness [111], stress conditions [36, 109], strain hardening [104, 109, 111] and specimen length [36, 111]. When the normalised crack length is over 0.5, $\eta^{LLD}$ becomes insensitive to the strain hardening, specimen thickness and length, therefore $\eta^{LLD}$ in different resources have a good agreement. Kim and Budden [111] provides upper-bound simulation for $\eta^{LLD}$.

Figure 3.22(a) and (b) illustrate the influence of crack length on average $\eta^{LLD}$ and $\eta^{CMOD}$ values over a range of mismatched ratios. In Figure 3.22(a), the scatter due to $N$, $h/W$ and $P\varepsilon, P\sigma$ is within a range of 10%, thus $\eta^{CMOD}$ is independent on $N$, $h/W$ and $P\varepsilon, P\sigma$. In Figure 3.22(b), for very shallow cracks ($a/W \leq 0.4 - 0.5$), the range of standard derivation is as high as 30-50%, mainly because $\eta^{LLD}$ significantly depends on $h/W$ and $N$. For deep cracks, $\eta^{LLD}$ becomes insensitive to $N$, $h/W$ and $P\varepsilon, P\sigma$ with a narrow scatter of $\pm 10\%$.

As shown in Figure 3.22(a), $\eta^{CMOD}$ is also relatively insensitive to $a/W$ ratio. The variation of $\eta^{CMOD}$ due to mismatch is at maximum $\pm 15\%$ compared to the even-match conditions. Based on ASTM E 1457 [51], $\eta^{CMOD}$ can be taken as 1.0 for homogeneous materials. As a welded specimen, the upper limit of 1.06 for under-matched welds ($M = 0.5$) and lower limit of 0.86
(M = 2) for over-matched welds is appropriate. As shown in Figure 3.22(b), $\eta_{LLD}$ value initially increases when $a/W \leq 0.5$ and stabilises when $a/W > 0.5$. For deep cracks, under even- or over-matched conditions, $\eta_{LLD}$ is independent on crack length, but in under-matched condition, $\eta_{LLD}$ would decrease due to the variations of $h/W$. The values of $\pm 10\%$ above and below the base material can be predicted on the welded specimens depends on the mismatched ratio. The variation of $\eta_{LLD}$ due to mismatch ratio may greater than 50% for the shallow cracks.

$J$ and $C^*$ can best be estimated using $\eta_{CMOD}$ rather than $\eta_{LLD}$ for SEN(T) specimen confirming the results in previous studies [36, 104, 105, 109, 111]. The average values of $\eta_{CMOD}$ for a given mismatch ratio has been plotted in Figure 3.23(a). A simple linear relationship can describe $\eta_{CMOD}$ in terms of $M$ as

$$\eta_{CMOD} = 1.13 - 0.13M \pm 0.09 \quad \text{for} \quad 0.1 \leq a/W \leq 0.7 \quad (3.6)$$

$\eta_{LLD}$ appears more dependent on stress conditions, strain hardening, $h/W$ and normalised crack length for shallow cracks, $a/W \leq 0.5$. However, for deep cracks, when $a/W \geq 0.5$ $\eta_{LLD}$ is less sensitive to the $a/W$, $N$, $P\varepsilon$, $P\sigma$ and $h/W$. The relationship between average $\eta_{LLD}$ and $M$ is shown in Figure 3.23(b) and this relationship can be described by a linear fit as

$$\eta_{LLD} = 3.03 - 0.49M \pm 0.22 \quad \text{for} \quad 0.5 \leq a/W \leq 0.7 \quad (3.7)$$

### 3.5.4 Single Edge Notch in Bending SEN(B)

Current standardised defect assessment procedures also suggest SEN(B) as another laboratory testing method for fracture specimens. Compared with C(T), SEN(B) enables existence of shallow crack in such a way that low constraint effects on fracture toughness can be examined [142].

$\eta$ values for experimental $J$ and $C^*$ estimation for homogenous SEN(B) specimens are compared with previous findings [36, 103, 109, 127] as shown in Figure 3.24 from which a good agreement has been obtained. In Figure 3.24 (a), the deviations of $\eta_{CMOD}$ in $a/W = 0.1$ in current results are remarkable since $\eta_{CMOD}$ shows a large dependency on $N$ and $P\varepsilon$, $P\sigma$. But if
consideration is not taken on \( a/W = 0.1 \), \( \eta^{CMOD} \)'s dependency on \( N \) and \( P \varepsilon, P \sigma \) is no longer obvious.

In Figure 3.24 (b), FE investigation has found that \( \eta^{LLD} \) on shallow cracks (\( a/W \leq 0.3 \)) would increase in increasing load, which deviated from the definition of \( \eta \). Therefore, large deviations on \( \eta^{LLD} \) are obtained in different literature. However, as \( a/W \geq 0.4 \), \( \eta^{LLD} \) becomes a good prediction as its insensitivity to \( N \) and \( P \varepsilon, P \sigma \). \( \eta^{LLD} \) is around 2.0 under different conditions.

Figure 3.25 (a) and (b) describe the effect of normalised crack length and mismatched ratio on average \( \eta^{CMOD} \) and \( \eta^{LLD} \). In Figure 3.25 (a), similar to homogeneous materials, the standard deviations at \( a/W = 0.1 \) are larger than that for longer normalised crack lengths. When \( \eta^{CMOD} \) is in the range of \( 0.2 \leq a/W \leq 0.7 \) the standard deviations due to \( N \) and \( P \varepsilon, P \sigma \) are within a range of 8%. The result of \( \eta^{CMOD} \) as shown in Figure 3.25 (a) also shows an increase versus normalised crack length. As for \( \eta^{LLD} \) seen in Figure 3.25 (b), \( \eta^{LLD} \) increases with crack length, showing standard deviations of about \( \pm 25\% \). When \( a/W > 0.4 \), \( \eta^{LLD} \) shows less scatter with lower standard deviations (\( \pm 10\% \)) since \( \eta^{LLD} \) becomes less sensitive to \( N \) and \( P \varepsilon, P \sigma \).

Figure 3.26(a) shows the relation between \( \eta^{CMOD} \) and \( M \). A linear fit equation to the data is given by

\[
\eta^{CMOD} = 0.79 - 0.08M \pm 0.12 \quad \text{for} \quad 0.2 \leq a/W \leq 0.7 \tag{3.8}
\]

For \( \eta^{LLD} \) versus \( M \) in Figure 3.26 (b), consideration of the average \( \eta^{LLD} \) on mismatched ratios is taken in medium to deep crack lengths, as \( 0.4 \leq a/W \leq 0.7 \). In this range, \( \eta^{LLD} \) is not remarkably dependent on \( N \), stress state, weld width and crack length. This relationship can also be described by a linear fit as

\[
\eta^{LLD} = 2.21 - 0.25M \pm 0.14 \quad \text{for} \quad 0.4 \leq a/W \leq 0.7 \tag{3.9}
\]

The recommended method to calculate \( J \) and \( C^\prime \) for this geometry is to use the \( CMOD \) measurements for \( a/W < 0.4 \) whilst for \( a/W > 0.4 \) \( \eta^{LLD} \) seems to be more accurate than \( \eta^{CMOD} \) due to its insensitivity to crack length and other test variables.
3.5.5 Double Edge Notch in Tension DEN(T)

Similar to SEN(T), DEN(T) can be also used to analyse near crack tip with relatively low constraint and low triaxility. Limited research [36, 110, 112] on $\eta$ has been performed on DEN(T) specimen. The available data was compared with the current study, shown in Figure 3.27.

In Figure 3.27 (a), results of $\eta$ using CMOD records are examined firstly. In ref [110], the effects of $P\varepsilon$ and $P\sigma$ on $\eta^{CMOD}$ were examined, showing that $\eta^{CMOD}$ under plane strain and plane stress would have different values. Also, $\eta^{CMOD}$ is a function of $N$, as proved in Ref [36, 110, 112]. However, after validating the uncertainty, it is found that data distribution is within $\pm20\%$ of the average, regardless of $N$ and $P\varepsilon$, $P\sigma$.

In Figure 3.27 (b), the investigation of $\eta$ using LLD records are also studied. Previously, ref [36] found that $\eta^{LLD}$ is strongly dependent on the ratio of specimen gauge length to half-width ($L/W$) for small cracks; Ref [110] suggested the specimen thickness can affect $\eta^{LLD}$ value by up to $60\%$. $\eta^{LLD}$ is also weakly dependent on $N$ and $P\varepsilon$, $P\sigma$ according to the current study. The $\eta^{CMOD}$ and $\eta^{LLD}$ for high hardening material in plane stress conditions provide lower bound values, whilst, low hardening material in plane strain conditions provide upper bound values.

The effects of crack length and mismatched ratio on average $\eta^{CMOD}$ and $\eta^{LLD}$ on DEN(T) specimen can be seen in Figure 3.28(a) and (b). Also found in the mismatched conditions is that the uncertainty in $\eta^{CMOD}$ and $\eta^{LLD}$ is about $\pm20\%$ due to $N$, $h/W$ and $P\varepsilon$, $P\sigma$.

As seen in Figure 3.28 (a), the effect of crack length to $\eta^{CMOD}$ is insensitive in under- or even-matched conditions. In under-matched condition $\eta^{CMOD}$ is $10\%$ higher than that of homogeneous values. In over-matched condition, as crack length increase, $\eta^{CMOD}$ would decrease and the variation of $\eta^{CMOD}$ to homogeneous condition also increases by up to $30\%$ difference for deep cracks. Therefore, the use of $\eta^{CMOD}$ method to evaluate $J$ and $C^*$ in deep cracks is not recommended. Seen in Figure 3.28 (b), increasing the normalised crack length leads to an
increase in $\eta^{LLD}$. But for deep cracks, compared to the variation of $\eta^{CMOD}$, $\eta^{LLD}$ provides a better prediction due to its lower variability. The variation due to mismatch is approximately $\pm$ 30% above and below the even-matched condition.

The mean values of $\eta^{CMOD}$ for different $M$ ratios are provided when $0.3 \leq a/W \leq 0.7$, as shown in Figure 3.29(a) and given an equation as

$$\eta^{CMOD} = 1.05 - 0.24M \pm 0.16 \quad \text{for} \quad 0.3 \leq a/W \leq 0.7$$

(3.10)

The scatter in over-matched conditions are about 20% which is due to the effects of $N$, $h/W$ and $P\varepsilon$, $P\sigma$, which also exhibit no clear correlation with respect to $\eta^{CMOD}$.

The relationship between $\eta^{LLD}$ and $M$ are shown in Figure 3.29 (b) for deep cracks when $0.5 \leq a/W \leq 0.7$ and given as a linear fit equation as

$$\eta^{LLD} = 1.06 - 0.34M \pm 0.18 \quad \text{for} \quad 0.5 \leq a/W \leq 0.7$$

(3.11)

In these cases within the scatter range of 15%, $\eta^{LLD}$ is not very sensitive to $N$, $h/W$ and $P\varepsilon$, $P\sigma$, values but still dependent on crack length. Compared the two methods on $\eta$ evaluation, $\eta^{CMOD}$ is better than $\eta^{LLD}$ due to its less variability on $N$, $h/W$ and $P\varepsilon$, $P\sigma$ under a wider range of normalised crack length.

3.5.6 Middle Crack in Tension, M(T)

When considering the welded structures, many defects may exist in the form of shallow cracks. The fracture toughness values in term of $J_c$ at crack initiation are higher in shallow cracks than in deep cracks [143], thus an unduly conservative values would be obtained. With the aim of better understanding the behaviour of shallow cracks, M(T) provides low constraint and low triaxiality, and also offers some advantages over bending specimens. The previous findings concluded that the corresponding $\eta^{LLD}$ and $\eta^{CMOD}$ are same for both plane strain and plane stress conditions, and thus can be applied to any specimen with a finite thickness [125].

The results like other geometries are also compared with literature [36, 109, 124, 144] firstly for homogeneous materials, as can be seen in Figure 3.30. The results of $\eta^{CMOD}$ shown in Figure
3.30 (a) was within the range of previous discussion in Ref [109] and [36]. In Ref [125], Kim and Schwalbe assumed \( \eta^{\text{CMOD}} = 1 \), which provides lower bound of \( \eta^{\text{CMOD}} \) value in current study, the upper bound of \( \eta^{\text{CMOD}} \) in current analysis is around 1.18 in shallow crack and 1.10 in deep crack. As mentioned, \( P \varepsilon, P \sigma \) would not bring any influence on \( \eta^{\text{CMOD}} \). Therefore \( \eta^{\text{CMOD}} \) is a simple \( N \)-dependent factor. The lower \( N \) values would result in a corresponding lower \( \eta^{\text{CMOD}} \) values, the dependency reduces for deep cracks and the maximum variation on \( N \) is within 10%.

In Figure 3.30 (b), \( \eta^{\text{LLD}} \) in current study is well consistent with previous study in [36, 109, 126]. \( \eta^{\text{LLD}} \) is still independent on \( P \varepsilon, P \sigma \) as \( 0.5 \leq a/W \leq 0.7 \). \( \eta^{\text{LLD}} \) would also vary with \( N \) values and the variations could be 20% for deep cracks.

The mean values of \( \eta^{\text{CMOD}} \) and \( \eta^{\text{LLD}} \) are plotted against different crack lengths and mismatch ratios, the graphs are shown in Figure 3.31 (a) and (b). In Figure 3.31(a), \( \eta^{\text{CMOD}} \) can be assumed to be insensitive to \( N, h/W \) and \( P \varepsilon, P \sigma \) within a range of ±12% scatter. In Figure 3.31(b), \( \eta^{\text{LLD}} \) shows a larger standard deviation especially for shorter normalised crack length of \( a/W < 0.5 \). It is found that for deep cracks, \( \eta^{\text{LLD}} \) would vary ±20% due to \( N, h/W \) and \( P \varepsilon, P \sigma \). Note that standard deviations of \( \eta^{\text{CMOD}} \) and \( \eta^{\text{LLD}} \) in under-matched conditions show large differences compared to over-matched conditions.

As shown in Figure 3.31 (a), \( \eta^{\text{CMOD}} \) slightly decreases as increasing normalised crack length within a rate of 5%. \( \eta^{\text{CMOD}} \) under different mismatched conditions show a maximum of ±20% variations from even-match conditions. In Figure 3.31 (b), \( \eta^{\text{LLD}} \) keeps increasing over a range of normalised crack lengths in even- or over-matched conditions, but is more stable in under-matched conditions. The variations on \( \eta^{\text{LLD}} \) due to mismatch are about ±50% in shallow crack compared to values of even-matched condition. Usually in M(T) specimens, initial crack lengths are between \( 0.1 \leq a/W \leq 0.5 \), hence using \( \eta^{\text{LLD}} \) compared to \( \eta^{\text{CMOD}} \) seems less accurate as the results are more sensitive to the changes in crack length, \( N, h/W \) and \( P \varepsilon, P \sigma \) in the former.
The mean $\eta^{\text{CMOD}}$ and $\eta^{\text{LLD}}$ versus $M$ in Figure 3.32 show a linear variation and the equations are described as

$$\eta^{\text{CMOD}} = 1.29 - 0.26M \pm 0.14 \quad \text{for} \quad 0.2 \leq a / W \leq 0.7 \quad (3.12)$$

$$\eta^{\text{LLD}} = 1.35 - 0.38M \pm 0.15 \quad \text{for} \quad 0.2 \leq a / W \leq 0.7 \quad (3.13)$$

The recommended method to evaluate the $J$ or $C^*$ for M(T) is to use $\eta^{\text{CMOD}}$ as it is less variable than $\eta^{\text{LLD}}$ especially for shallow cracks when $a / W < 0.5$.

### 3.5.7 Summary of Results

The calculated values of $\eta^{\text{CMOD}}$ and $\eta^{\text{LLD}}$ for specific normalised crack lengths and mismatch ratios have been examined based on simple linear fits for each of the geometries. A summary of best fit equations in six fracture mechanics specimens from the analysis data is tabulated in Table 3.3. Given the variability in trying to compare the effects due to $N$, $h/W$ and $P\varepsilon$, $P\sigma$, it is logical to make them independent of $\eta^{\text{CMOD}}$ and $\eta^{\text{LLD}}$ to be within the observed scatter. Therefore the final results shown in Appendix B are comprehensive and generally material independent within the scatter range specified. This means that they are sufficiently accurate to be adopted in codes of practice and standards for derivations of fracture mechanics parameters. These calculations are based on the assumption that the mechanical properties of welds and base material vary with the mismatch ratio range of between 0.5 and 2.0. The values are a maximum $\pm 25\%$ above and below the mean values of $\eta^{\text{CMOD}}$ and $\eta^{\text{LLD}}$, depending on the mismatch and the level of conservatism needed, for homogenous materials may be adopted for most situations when the testing specimen contains a thin section of welds with $h/W < 0.1$. Thin welds condition should be the preferred approach for most practical conditions in weldment test specimens.

### 3.6 Summary

The numerical solutions of $\eta^{\text{CMOD}}$ and $\eta^{\text{LLD}}$, calculated based on load line displacement and crack mouth opening displacement, have been systematically investigated for a range of centre-cracked, square-grooved weld fracture geometries. The examined geometries include compact
tension specimen C(T), and C-shaped tension specimen, CS(T), single edge notch specimen in tension SEN(T) and bending SEN(B), double edge notch specimen in tension DEN(T), middle crack specimen in tension M(T). This work addresses the effect of weld mismatch ratio for a range of $0.5 \leq M \leq 2$, $0.05 \leq h/W$(or $2W) \leq 0.10$, hardening exponent, $5 \leq N \leq 20$, and crack length $0.1 \leq a/W \leq 0.7$ on the solutions of the fracture mechanics parameters under both plane strain and plane stress conditions. It should be noted that this extensive analysis and comparison work employed to derive accurate values for $\eta$ has yielded an unexpected conclusion in so far that there is no clear trend between $\eta$ and the material variables and that for the present geometries the scatter of data ranges to a maximum of about $\pm 30\%$ which in real has little effect on the accuracy of the correlating parameters that are derived. At best an upper/lower bound value for $J$ and $C^*$ would be derived for use. In any case the present study enables either $\eta^\text{CMOD}$ and $\eta^\text{LLD}$ to be used for $J$ and $C^*$ estimations and the range of results are presented as equations for mismatch variations and tabulated in Table 3.3 for the appropriate crack length and material mismatch highlighting upper and lower bounds that can be applied. It is generally found that

1. $h/W$ influences the values of $\eta^\text{CMOD}$ and $\eta^\text{LLD}$ with respect to $M$ where average variations of $\pm 20\%$ for $\eta^\text{CMOD}$ and $\pm 50\%$ for $\eta^\text{LLD}$ can be expected. By limiting the $h/W$ to a thin welded region with $h/W$(or $2W) \leq 0.1$, the variations due to $h/W$ reduces to $15\%$. This is sufficient for most practical test configurations.

2. The level of weld strength mismatch affects the value of $\eta^\text{CMOD}$ and $\eta^\text{LLD}$ for a given crack length decreases as mismatch ratio increases. Based on the values calculated for different mismatch conditions, both $\eta^\text{CMOD}$ and $\eta^\text{LLD}$ can be obtained for inhomogeneous bi-material fracture mechanics specimens using suggested equation derived from the present numerical studies. These are summarised in Table 3.3. Up to $\pm 25\%$ variations due to mismatch effects mean that the $\eta$ factor in thin welded geometry is close to that in homogeneous specimens.

3. The use of $\eta$ using CMOD measurements is the recommended method to evaluate $J$ and $C^*$ in SEN(T), DEN(T), M(T) and SEN(B) in both homogenous and inhomogeneous
materials for shallow cracks since the results shows less dependency on specimen geometries, material properties and $h/W$ and show $\pm 20\%$ upper/lower bounds variations in $\eta^{CMOD}$.

4. $\eta^{LLD}$ is identical to $\eta^{CMOD}$ in C(T) specimen, therefore both of $\eta$ values have proved to be equally accurate methods to evaluate $J$ and $C^*$. The use of $\eta$ using $LLD$ measurements is preferred for deep cracked CS(T) and SEN(B) specimens. Compared with $\eta^{CMOD}$, $\eta^{LLD}$ values seem slightly less dependent on crack length.
3.7 Tables

Table 3.1: Specimen dimensions employed in numerical $P_{\sigma}, P_{\varepsilon}$ analyses

<table>
<thead>
<tr>
<th>Specimen Type</th>
<th>$W$ mm</th>
<th>$B$</th>
<th>$H, L, R_o$</th>
<th>$a/W$</th>
<th>$2h$</th>
</tr>
</thead>
<tbody>
<tr>
<td>C(T)</td>
<td>50</td>
<td>$W/2$</td>
<td>$H=1.2W$</td>
<td>0.35, 0.40, 0.45, 0.50, 0.60, 0.70</td>
<td>$h/W=0.05, 0.1$</td>
</tr>
<tr>
<td>CS(T)</td>
<td>25</td>
<td>$W/2$</td>
<td>$R_o=2W$</td>
<td>0.1-0.7 with an increment of 0.1</td>
<td>$h/2W=0.05, 0.1$</td>
</tr>
<tr>
<td>SEN(T)</td>
<td>50</td>
<td>$W/4$</td>
<td>$L=2W$</td>
<td>0.1-0.7 with an increment of 0.1</td>
<td>$h/W=0.05, 0.1$</td>
</tr>
<tr>
<td>SEN(B)</td>
<td>25</td>
<td>$W/2$</td>
<td>$L=2W$</td>
<td>0.1-0.7 with an increment of 0.1</td>
<td>$h/2W=0.05, 0.1$</td>
</tr>
<tr>
<td>DEN(T)</td>
<td>25</td>
<td>$W/2$</td>
<td>$L=4W$</td>
<td>0.2-0.7 with an increment of 0.1</td>
<td>$h/2W=0.05, 0.1$</td>
</tr>
<tr>
<td>M(T)</td>
<td>25</td>
<td>$W/2$</td>
<td>$L=4W$</td>
<td>0.2-0.7 with an increment of 0.1</td>
<td>$h/2W=0.05, 0.1$</td>
</tr>
</tbody>
</table>

Table 3.2: Summary of numerical and analytical study of $\eta$ factor for a range of geometries in literature

<table>
<thead>
<tr>
<th>Author</th>
<th>Specimen Type</th>
<th>Mismatch Ratio ($M$)</th>
<th>$N$</th>
<th>Material Properties</th>
<th>$a/W$</th>
<th>Geometry</th>
<th>$P_{\sigma}$ / $P_{\varepsilon}$ / 3D</th>
</tr>
</thead>
<tbody>
<tr>
<td>Marie et. al. (2011)[102]</td>
<td>C(T)</td>
<td>2.3</td>
<td>Experimental</td>
<td>$\sigma_{BM}=305$ MPa</td>
<td>0.50</td>
<td>CT12, CT25, CT50</td>
<td>$P_{\varepsilon}$</td>
</tr>
<tr>
<td>Donato (2009)[103]</td>
<td>SEN(B)</td>
<td>1.0 0.8 1.2 1.5 2.0</td>
<td>WM 10.0 11.5 12.8 14.5</td>
<td>BM 10 10 10 10</td>
<td>$\sigma_{BM}=412$ MPa, $\sigma_{WM}=453$ MPa, $\sigma_{BM}=494$ MPa, $\sigma_{WM}=536$ MPa, $\sigma_{BM}=412$ MPa, $E=206$ GPa</td>
<td>0.1, 0.15, 0.2, 0.25, 0.3, 0.5, 0.7</td>
<td>W=2B, S/W=4, 2h=5-20 mm</td>
</tr>
<tr>
<td>Ruggieri (2012) [104]</td>
<td>SEN(T)</td>
<td>Even match ($M=1$)</td>
<td>5</td>
<td>10 20</td>
<td>$E/\sigma_0=800$</td>
<td>0.1-0.7 with an increment of 0.05</td>
<td>Pin-loaded: $L/W=3$, Clamp-loaded: $L/W=3$ or 5</td>
</tr>
<tr>
<td>Paredes and Ruggieri (2012) [105]</td>
<td>SEN(T)</td>
<td>1.0 1.1 1.2 1.3</td>
<td>WM 10.0 11.5 12.8 14.5</td>
<td>BM 10 10 10 10</td>
<td>$\sigma_{BM}=412$ MPa, $\sigma_{WM}=453$ MPa, $\sigma_{BM}=494$ MPa, $\sigma_{WM}=536$ MPa, $\sigma_{BM}=412$ MPa, $E=206$ GPa</td>
<td>0.1-0.7 with an increment of 0.1</td>
<td>Pin-loaded: $L/W=3$, Clamp-loaded: $L/W=3$ or 5, 2h=10, 15, 20 mm</td>
</tr>
<tr>
<td>Kim et. al. (2004)[109]</td>
<td>SEN(T)</td>
<td>Even match ($M=1$)</td>
<td>5</td>
<td>10</td>
<td>$E=200$ GPa $\sigma_0=400$ MPa</td>
<td>0.5</td>
<td>$B/W=0.5, 0.25$ and 0.125</td>
</tr>
<tr>
<td>Davies et. al.(2007) [36]</td>
<td>SEN(T)</td>
<td>Even match ($M=1$)</td>
<td>3 3</td>
<td>5 5</td>
<td>$E=210$ GPa $\sigma_0=660$ MPa</td>
<td>0.1-0.7 with an increment of 0.1</td>
<td>Same as Table 3.1</td>
</tr>
</tbody>
</table>
Table 3.2 continued

<table>
<thead>
<tr>
<th>Source</th>
<th>Condition</th>
<th>Test Type</th>
<th>BM</th>
<th>Specimen</th>
<th>Width</th>
<th>E/σ₀</th>
<th>Test Parameters</th>
<th>Results</th>
</tr>
</thead>
<tbody>
<tr>
<td>Kim (2004) [110]</td>
<td>DEN(T)</td>
<td>Even match (M = 1)</td>
<td>5 10 20</td>
<td>0.1-0.8 with an increment of 0.05</td>
<td>W/B=0.5,1,5 and 10</td>
<td>P σ and P ε</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Kim and Budden (2001) [111]</td>
<td>SEN(T)</td>
<td>Even match (M = 1)</td>
<td>5 20</td>
<td>-</td>
<td>L/W=1.5</td>
<td>P σ and P ε</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Wilson (2008) [112]</td>
<td>DEN(T)</td>
<td>Even match (M = 1)</td>
<td>2,3,5,7,10, 13,16,20</td>
<td>E = 72 GPa σ₀ = 385 MPa</td>
<td>0.125-0.875 with an increment of 0.125</td>
<td>P σ and P ε</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Savioli et. al. (2011) [113]</td>
<td>C(T)</td>
<td>Even match (M = 1)</td>
<td>1.0 1.1 1.2 1.3</td>
<td>0.4-0.7 with an increment of 0.05</td>
<td>B = 25.4 mm W = 2B 2h = 10, 15, 20 mm</td>
<td>P ε</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Xuan et. al. (2005) [114]</td>
<td>C(T)</td>
<td>0.25,0.5,0.75, 1.25,1.5, 2</td>
<td>9.03</td>
<td>σ₀/E = 175 GPa</td>
<td>0.1-0.9 with an increment of 0.1</td>
<td>2h = 12,10,8,6,4</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td>Panontin et. al. (2000) [115]</td>
<td>C(T)</td>
<td>Even match (M = 1)</td>
<td>5 10 20</td>
<td>E/σ₀ = 800 E/σ₀ = 500 E/σ₀ = 250 (Ferritic steels)</td>
<td>0.4-0.7 with an increment of 0.05</td>
<td>-</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Smith (1992)[121]</td>
<td>C(T)</td>
<td>Even match (M = 1)</td>
<td>-</td>
<td>-</td>
<td>0-1 with an increment of 0.1</td>
<td>-</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Cassanelli et. al. (2003) [122]</td>
<td>C(T)</td>
<td>Even match (M = 1)</td>
<td>-</td>
<td>2.25Cr 1.0Mo, ASTM A387-Gr22 type steel plate</td>
<td>0.65</td>
<td>W = 60 mm side groove ratio: 0%, 25% and 50%</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td>Kim and Chao (2003) [123]</td>
<td>CS(T)</td>
<td>Even match (M = 1)</td>
<td>Steel 304 L</td>
<td>-</td>
<td>0.54</td>
<td>R₀=18.9 R₁=9.14</td>
<td>3D and P ε</td>
<td></td>
</tr>
<tr>
<td>Wilson (2002)[124]</td>
<td>M(T)</td>
<td>Even match (M = 1)</td>
<td>2,3,5,7,10, 13,16,20</td>
<td>E = 72 GPa σ₀ = 385 MPa</td>
<td>0.125-0.875 with an increment of 0.125</td>
<td>P σ and P ε</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Kim and Schwalbe (1999) [125]</td>
<td>M(T) C(T) SEN(B)</td>
<td>Even match (M = 1)</td>
<td>A range of materials</td>
<td>M(T) SEN(B): 0.1-0.7 C(T): 0.45-0.7</td>
<td>-</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>O'Dowd (1997) [126]</td>
<td>M(T)</td>
<td>Even match (M = 1)</td>
<td>4 10 20</td>
<td>E/σ₀ = 500</td>
<td>0.1,0.25,0.5</td>
<td>L/W=2</td>
<td>P σ and P ε</td>
<td></td>
</tr>
<tr>
<td>Kirk and Dodds (1992) [127]</td>
<td>SEN(B)</td>
<td>Even match (M = 1)</td>
<td>4 5 10 50</td>
<td>0.05,0.15,0.25, ,0.5 and 0.7</td>
<td>L/W=2</td>
<td>B/W=1, 0.5,0.25</td>
<td>3D and P ε</td>
<td></td>
</tr>
<tr>
<td>Nevalainen and Dodds (1995) [128]</td>
<td>SEN(B) C(T)</td>
<td>Even match (M = 1)</td>
<td>5 10 20</td>
<td>E/σ₀ = 500</td>
<td>SEN(B): 0.1, 0.5 C(T): 0.6</td>
<td>-</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
Table 3.3: Summary of best fit equations to show the sensitivity of the mean values of $M$ to $\eta$ in six fracture specimens giving the approximate upper/lower bound as estimated from the analysis data

<table>
<thead>
<tr>
<th>Specimen Type</th>
<th>Equation</th>
<th>Lower Bound</th>
<th>Upper Bound</th>
</tr>
</thead>
<tbody>
<tr>
<td>C(T)</td>
<td>$\eta_{LLD} = 2.71 - 0.38M \pm 0.20$</td>
<td>$0.45 \leq a/W \leq 0.7$</td>
<td></td>
</tr>
<tr>
<td>CS(T)</td>
<td>$\eta_{CMOD} = 4.27 - 0.49M \pm 0.38$</td>
<td>$0.3 \leq a/W \leq 0.7$</td>
<td>$0.3 \leq a/W \leq 0.7$</td>
</tr>
<tr>
<td>SEN(T)</td>
<td>$\eta_{CMOD} = 1.13 - 0.13M \pm 0.09$</td>
<td>$0.1 \leq a/W \leq 0.7$</td>
<td></td>
</tr>
<tr>
<td>SEN(B)</td>
<td>$\eta_{CMOD} = 0.79 - 0.08M \pm 0.12$</td>
<td>$0.2 \leq a/W \leq 0.7$</td>
<td>$0.4 \leq a/W \leq 0.7$</td>
</tr>
<tr>
<td>DEN(T)</td>
<td>$\eta_{CMOD} = 1.05 - 0.24M \pm 0.16$</td>
<td>$0.3 \leq a/W \leq 0.7$</td>
<td>$0.5 \leq a/W \leq 0.7$</td>
</tr>
<tr>
<td>M(T)</td>
<td>$\eta_{CMOD} = 1.29 - 0.26M \pm 0.14$</td>
<td>$0.2 \leq a/W \leq 0.7$</td>
<td>$0.2 \leq a/W \leq 0.7$</td>
</tr>
</tbody>
</table>

Note: $\eta_{LLD}$ and $\eta_{CMOD}$ are the lower and upper bounds of the mean values of $M$, respectively.
3.8 Figures

Figure 3.1: Schematic illustration of the fracture geometries and their dimensions, showing the weld/base distribution
Figure 3.2: Loading condition, meshing and boundary condition designs for the six fracture mechanics geometries, including the collapsed mesh design of near tip region.

Figure 3.3: The variation of $\eta^{\text{LD}}$ with $h/W$ ratios for different over-matched ratios in C(T) in (a) shallow crack of $a/W = 0.35$; (b) medium crack of $a/W = 0.5$; (c) deep crack of $a/W = 0.7$. 
Figure 3.4 The variation of $\eta_{CMOD}$ with $h/W$ ratios for different mismatch ratios in CS(T) in (a) shallow crack of $a/W = 0.3$; (b) medium crack of $a/W = 0.5$; (c) deep crack of $a/W = 0.7$

Figure 3.5 The variation of $\eta_{LLD}$ with $h/W$ ratios for different over-matched ratios in CS(T) in (a) shallow crack of $a/W = 0.3$; (b) medium crack of $a/W = 0.5$; (c) deep crack of $a/W = 0.7$

Figure 3.6 The variation of $\eta_{CMOD}$ with $h/W$ ratios for over-matched ratios in SEN(T) in (a) shallow crack of $a/W = 0.3$; (b) medium crack of $a/W = 0.5$; (c) deep crack of $a/W = 0.7$
Figure 3.7 The variation of $\eta^{LLD}$ with $h/W$ ratios for over-matched ratios in SEN(T) in (a) shallow crack of $a/W = 0.3$; (b) medium crack of $a/W = 0.5$; (c) deep crack of $a/W = 0.7$

Figure 3.8 The variation of $\eta^{CMOD}$ with $h/W$ ratios for over-matched ratios in SEN(B) in (a) shallow crack of $a/W = 0.3$; (b) medium crack of $a/W = 0.5$; (c) deep crack of $a/W = 0.7$

Figure 3.9 The variation of $\eta^{LLD}$ with $h/W$ ratios for over-matched ratios in SEN(B) in (a) shallow crack of $a/W = 0.3$; (b) medium crack of $a/W = 0.5$; (c) deep crack of $a/W = 0.7$
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Figure 3.11 The variation of $\eta_{LLD}$ with $h/W$ ratios for over-matched ratios in M(T) in (a) shallow crack of $a/W = 0.3$; (b) medium crack of $a/W = 0.5$; (c) deep crack of $a/W = 0.7$

Figure 3.12 The variation of $\eta_{CMOD}$ with $h/W$ ratios for over-matched ratios in DEN(T) in (a) shallow crack of $a/W = 0.3$; (b) medium crack of $a/W = 0.5$; (c) deep crack of $a/W = 0.7$
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Figure 3.24: Studies of $\eta$ results in homogeneous SEN(B) specimen under $P_\varepsilon$ and $P_\sigma$ conditions compared with (a) [36, 103, 109, 127] for $\eta_{\text{CMOD}}$ (b) [36, 109, 127] for $\eta_{\text{LLD}}$
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Figure 3.26: Linear fit to the mean \( \eta \) values for different mismatch ratios in SEN(B) (a) \( \eta^{CMOD} \) for \( 0.2 \leq a/W \leq 0.7 \) (b) \( \eta^{LLD} \) for \( 0.4 \leq a/W \leq 0.7 \)
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Chapter 4

Materials and Experimental Techniques

4.1 Introduction

Experimental tests have been performed to investigate the remaining life of the service-exposed AISI 316H stainless steel welded components with the existing of welding residual stresses. This chapter details the experimental and data processing techniques used for all the experiments performed in this project. This chapter firstly provides a short description of the materials used. A typical type 316H austenitic stainless steel welded steam header component from 1B1/1 header with cast number of 53415 has been mainly studied, focusing on the cross-weld region. A previous study on this alloy [134] has investigated as-received and pre-strain behaviour of the base metal (BM) from the same header, the results of which have been compared with the data obtained in this work. Also, the behaviour of welded 316H stainless steel headers operated at 550 °C from other resources have also been reviewed in order to make comprehensive and meaningful comparisons. In addition, the experimental techniques and testing procedures used are presented. The methods used include micro-hardness tests, tensile tests measured by digital image correlation technique (DIC), uniaxial creep tests and creep crack growth tests. Regarding the tensile test results, the values of Young’s modulus, Passion’s ratio and yield stress are required to perform the basis stress analysis. Creep rupture data are required to calculate the rupture life of studied materials and to estimate the current continuum damage level as the defect grows. Creep crack growth data are required to calculate crack growth under steady loading conditions and to estimate the crack extension.

4.2 Material Specification

4.2.1 Material Acquisition

Type 316H stainless steel used in this research was provided by EDF Energy from a welded steam header component, denoted 1B1/1 (cast 53415), taken from Heysham power plant in the
UK. Figure 4.1 shows a photograph and a schematic illustration of the header, which identify the region of interest (seen red area) for this study where reheat cracks have been observed in the welded region of the nozzle. An outlet nozzle has been attached by welding onto the side of the header cylinder. The nozzle has been joined to the header via a multi-pass manual metal arc (MMA) weld [69]. The nozzle was removed from the header to investigate the weldment and the cracking regions. Figure 4.2 illustrates the shape and dimensions of the nozzle used in the present research.

The header and nozzle was manufactured from type 316H austenitic stainless steel, the chemical composition (by % weight) of header 1B/1 is provided in Table 4.1. The weld is type 316L stainless steel, the chemical composition requires further examination.

4.2.2 Service Condition

The header was heat treated at 1,050 °C around 3 hours and water quenching before being put into service. The header was removed from service after 87,790 hours at around 523°C, due to the detection of creep cracks.

4.2.3 Sample Descriptions

The service exposed steam header has been studied. Figure 4.3 shows a more detailed view of the region of main interest in the header 1B1/1, which is located in the welding area in the vicinity of the branch between the header and the nozzle. This tapered region has a thickness from 37 mm to 70 mm, an inner radius ($R_i$) of 51 mm and a height of approximately 135 mm. The cylindrical part of the header itself is approximately 65 mm in thickness. This region is therefore expected to be a highly constrained geometry.

As seen in Figure 4.3, re-heat cracks have been observed in the header 1B1/1 which initiated on the transition radius from the nozzle to the header, the partially circumferential cracks grew normal to the corner of the joint and in the vicinity of, but not on, the fusion line. Research on the cracking mechanism is described in Chapter 7.

In order to clearly inspect the cross section of the nozzle/header branch shown in Figure 4.3, the cross sectional surface, after cutting, was ground, polished and then macro-etched by ferric
chlorine (FeCl₃) solution. A typical macro-etched graph of the cross-weld region is shown in Figure 4.4, where the weld metal appears shinier than the base metal and there is a narrow transition zone that illustrates the HAZ section, the HAZ boundary would be clearly inspected after micro-etching.

Standard fracture mechanics specimens according to ASTM E1457 were extracted from the welded components. Designed C(T) and SEN(T) specimen geometry is shown in Figure 4.5 and Figure 4.6, respectively. In some specimens, in order to investigate the cracking behaviour similar to the real situation, the initial crack was introduced within the base metal but near the fusion line to replicate to the region of the crack in the header (shown in Figure 4.5 and Figure 4.6). As an example, the C(T) specimen cut from the studied nozzle/header are shown in Figure 4.7. In order to accurately position the initial cracks, the cross-weld fracture specimens were firstly polished and etched to show the BhyM/HAZ boundary before Electric Discharge Machining (EDM). Uniaxial and tensile cross-weld specimens were also made from the studied nozzle aiming to test the cross-weld region which was manufactured, such that the cross-weld was within the gauge length.

Figure 4.8 shows the drawing to cut testing specimens from the nozzle/header components under investigation. In order to maximise the number of specimens that can be made for testing, a CAD model was developed to optimise positions for specimen extraction. Based on this design, several uniaxial C(T), SEN(T) specimens containing cross-weld regions and C(T), SEN(T) specimens made in homogeneous base metal were manufactured.

**4.2.4 Specimen Design and Orientation**

The dimensions of C(T) and SEN(T) have been defined according to ASTM standard E1457 [51]. The standardised specimen design on uniaxial and tensile specimens were based on ASTM standard E8M [145] and E21 [146] respectively. The dimensions for each specimen are shown in Figure 4.10, Figure 4.13, Figure 4.15 and Figure 4.17.

Consideration was also taken to specimen orientations. After long term exposure at high temperature, some defects may trigger a weakness in a specific direction and result in earlier crack initiation and propagation. As for the welded structure, the material was no longer homogeneous and isotropic, under this circumstance, sensitivity to orientation was particularly
pronounced in fracture toughness and creep testing [11]. For this reason, the specimens have been machined in the same orientation in order to obtain the comparable test results. There are altogether six possible orientations for the specimen alignment within the axes of symmetry in the plate. The letters C, R and L denote the circumferential, radial and longitudinal directions, respectively [11].

Figure 4.8 shows schematic illustration of the specimen orientation. As-welded fracture specimens were taken from the L-R orientation, which means loading in the longitudinal (L) direction and the crack was to be propagated in radial (R) direction.

### 4.3 Test Matrix

Test specimens extracted from the studied nozzle$header component$s were designed to perform tensile, uniaxial creep (UC) and CCG testing at room or elevated temperature at 550 °C. The full details of the proposed experiments are shown in Table 4.2 which specifies the specimen type, ID, main geometries and testing temperatures.

### 4.4 Available Material Datasets

In order to comprehensively identify the tensile, uniaxial creep rupture and creep crack growth behaviour of service exposed 316H austenitic stainless steel welded component, and also due to the limited resources from this study, the testing results from these tests are compared to those of available on the material pre-compressed to 8% plastic strain at room temperature (8% PC) and as-received (AR) materials from same or different header [134]. An analysis of testing results has been made combined with the previous tests on ex-service type 316H stainless steels at 550 °C, mostly performed at Imperial College London [56, 134, 147-151]. A summary of collected data from tensile, uniaxial creep rupture and creep crack growth test are shown in Table 4.3.

### 4.4.1 Pre-straining Effects

Material pre-straining has been recognised as an important fabrication process as it can strengthen metallic materials by creating dislocation barriers to restrain subsequent dislocation
movement [152]. It is found that pre-straining can introduce resistance on the subsequent high temperature creep, in such a way the creep fracture strain is influenced.

The influence of plastic pre-strain on the creep deformation and crack growth properties in 316H stainless steel from the same header has been investigated by Mehmanparast, A. et al.[134, 153, 154]. It is found that pre-compressed 316H material has material hardening effects and the ductility is much lower than as-received materials. These findings are similar to cross-weld specimens under the current investigations. Based on this similarity, the results obtained from the present study are directly compared to those of available on the material pre-compressed to 8% plastic strain at room temperatures.

4.5 Micro-Hardness Measurement

The Vickers micro-hardness of a service exposed cross-weld sample containing reheat cracks and an electron beam new-welded (NW) compact tension C(T) specimen with half BM and half WM (CT-NW-1) were measured using Zwick ZDU 0.2 Hardness Tester. All hardness readings were obtained at indentation load of 5 kg-f with an indentation time of 12 seconds, reported as HV5/12. The testing procedure is according to the standard ASTM E 384 [155].

4.6 Tensile Tests Measured by Digital Image Correlation (DIC)

4.6.1 Introduction

DIC is an effective tool for displacement and strain measurement that has been used in the domain of experimental mechanics [156]. Due to its capability for fast data acquisition, this technique is well suited for the characterisation of material properties both in the elastic and plastic ranges. It also has advantages of full field, non-contact, and considerately high accuracy. Most often, the traditional extensometer technique is used to measure the tensile strain between two points which can only provide average strain development over an area. However, it is practically difficult to find the detailed strain deformation on a narrow cross-weld specimen because the traditional measurement cannot provide a full-range strain distribution during the tensile test.
DIC, on the other hand, is based on sophisticated computational algorithms that can track the position of the same physical points shown in a reference image and a deformed image, therefore, the strain deformation before and after a loading event can be detected [157]. To achieve this, a square subset of pixels are identified on the speckle pattern around point of interest on the reference image and their corresponding location can be determined on the deformed image [156].

DIC techniques provide an accurate strain on a specific stress applied on the sample. According to the true stress vs. true strain curve, the 0.2% proof stress, $\sigma_{0.2}$ and Young’s modulus, $E$ of base metal, weld metal and cross-weld can be determined from a single test. The current study measured the strain variations on three cross-weld tensile specimens at room temperature.

### 4.6.2 Tensile Specimens

The tensile specimens were cut from a block extracted from the weldments on the studied header; schematic diagram of cross-weld specimens in blocks is shown in Figure 4.9. The dimensions of room temperature tensile specimen are illustrated in Figure 4.10.

### 4.6.3 Testing Procedure

Prior to the DIC tensile test, one side of the specimen was ground and etched to reveal the structure of the weld. The weld areas were marked in order to observe the failure positions. Figure 4.11 illustrates the weld regions of the three tensile specimens.

The specimen was then lightly coated with black paint and dusted lightly with white powders to obtain a random speckle black and white pattern on the through-thickness cross-section [158], the speckle pattern can be seen in Figure 4.12. The average speckle diameter is approximately 0.03 mm.

Once the pattern applied, the prepared specimen was loaded into the tensile testing machine. Each tensile test was performed with a 30kN load cell under constant displacement rate of 3mm/min. The camera was positioned so that the line of sight was normal to the specimen surface to minimise out-of-plane effects [159]. An initial image capturing the undeformed image was required and subsequent images were acquired every 5 second. Each deformed image was
correlated with the initial undeformed image to compare to the appropriate deformed image and converted the displacement data into the full-field displacement gradient tensor. The tensile test specimens were loaded until fracture. The fractural position usually occurs in the weld metal or in the boundary between the weld metal and the base metal. The strain was averaged across the width of the specimen and the evolution of the strain variation along the gauge length was obtained under the computational investigation. Therefore, the corresponding stress-strain curve was obtained.

4.7 Uniaxial Creep-Rupture Tests

4.7.1 Introduction

Uniaxial creep-rupture tests can determine the creep curve and strain at failure of a material under constant tensile load at constant temperature [19]. The creep curve has been introduced in Section 2.3.3 where creep strain is plotted against time, as described in Figure 2.4. As indicated in this figure, some important parameters, including duration of each of the creep stages, secondary creep strain rate, average creep strain rate, time to rupture, and creep failure strain (creep ductility) can be deduced.

4.7.2 Uniaxial Creep-Rupture Specimens

The uniaxial creep rupture specimens geometry was designed according to ASTM standard E 8M [145], the ridges were added to the top and bottom of the gauge region for extensometer purposes as shown in Figure 4.13. Four uniaxial creep rupture specimens containing cross-weld regions within the gauge length (illustrated in Figure 4.14) were tested.

4.7.3 Test Procedure

The general procedure for creep rupture tests are standardised in ASTM E 139-11 [19]. All the tests were conducted under a dead weight level arm machine with a level arm ratio of 10:1. The temperature was monitored by two thermocouples attaching along the upper and lower positions of gauge length. The temperatures were limited to less than 2° C from the test temperature of 550° C.
Axial displacement was measured by a displacement transducer called Linear Variable Differential Transformer (LVDT). The output voltage from the LVDT was converted into a displacement based on a linear relationship between those two signals. The calibration of LVDT was performed before and after each test.

During the creep test, the temperature and displacement were continuously measured every 5 minutes throughout the test until rupture.

### 4.7.4 Nominal Stress and Creep Failure Strain

Uniaxial rupture tests were under constant load. The assumption of constant stress allows the elastic-plastic-creep material properties to be easily derived from a uniaxial test [160]. Engineering nominal net section stress, $\sigma_{nom}$ is equal to the ratio of applied load $P$ to initial net cross section $A_0$ as

$$\sigma_{nom} = \frac{P}{A_0} \quad (4.1)$$

For the specimen under large deformation, the cross-sectional area and the length of the specimen can change substantially thus the engineering definition of stress ceases to be an accurate measurement, under this circumstance, the true definition net section stress after loading is defined as

$$\sigma_{true} = \sigma_{nom} (1 + \varepsilon_{load}^{eng}) \quad (4.2)$$

where $\varepsilon_{load}^{eng}$ is the engineering strain during load up.

The creep failure strain (creep ductility) is a material property that describes how a material can be deformed without failure in terms of creep deformation. For a uniaxial creep rupture test, engineering strain definition assumes that strain is uniform throughout the specimen, so that engineering creep ductility can be written as

$$\varepsilon_{f}^{eng} = \frac{\Delta_f - \Delta_{load}}{l_0} \quad (4.3)$$
where $\Delta_f$ is the axial creep displacement during the test, $\Delta_{\text{load}}$ is the axial displacement due to the instantaneous elastic-plastic material response and $l_0$ is the initial gauge length. If large deformation occurs, the reliable true failure strain can be defined as

$$\varepsilon_f^{\text{true}} = \ln(1 + \varepsilon_f^{\text{eng}}) \quad (4.4)$$

For some creep ductile materials, substantial necking was observed local to the failure region, an assumption of uniform strain was no longer valid. Under this circumstance, a failure strain considering the reduction of area (ROA) is regarded as a more reliable method. Engineering reduction of area failure strain $\varepsilon_f^{\text{eng}}$ is defined as

$$\varepsilon_f^{\text{eng}} = \frac{A_0 - A_f}{A_0} - \varepsilon_{p,\text{load}}^{\text{eng}} \quad (4.5)$$

where $A_f$ is failure cross sectional area and $\varepsilon_{p,\text{load}}^{\text{eng}}$ is axial elastic plastic strain at loading. The cross sectional area after loading up cannot be measured thus Equation (4.5) provides an estimated value of the creep ductility. The true definition of ROA failure strain can be calculated from

$$\varepsilon_f^{\text{true}} = 2 \ln \left( \frac{d_0}{d_f} \right) - \varepsilon_{p,\text{load}}^{\text{true}} \quad (4.6)$$

where $d_0$ and $d_f$ are the initial and failure diameters. $\varepsilon_{p,\text{load}}^{\text{true}}$ is the true axial plastic strain at loading which can be calculated from $\varepsilon_{p,\text{load}}^{\text{eng}}$ using a similar form of Equation (4.4). Usually localised creep ductility under ROA estimation provides greater value than axial measurement.

### 4.8 Creep Crack Growth (CCG) Tests

#### 4.8.1 Introduction

Creep crack growth (CCG) tests have been performed on two types of geometries, i.e. C(T) and SEN(T). The test procedure was based on the standard ASTM E1457 [51]. During the CCG test, the load line displacements were recorded and the crack length was monitored using the potential drop (PD) method. The load line displacement rates and the corresponding crack growth rate were then used to calculate $C^*$ parameters.
4.8.2 Specimen Geometries

As shown in the test matrix (Table 4.2), 2 C(T) and 1 SEN(T) specimens have been performed in the CCG tests. These specimens were extracted from the cross-weld region of ex-service steam header in such a way that initial cracks were located on the BM metal but near the weld line. In addition, 1 electron beam new-welded C(T) specimen with half BM and half WM, denoted as CT-NW-1 was tested, in this C(T) specimen, the initial crack was located on the boundary between the BM and WM material.

The geometry of CT-1 and CT-2 extracted from ex-service header is shown in Figure 4.15, all dimensions are dependent on $W$ where $W = 40$ mm. 15% side groove in each side was machined into the sample to meet crack front straightness requirements. CT-NW-1 was side-grooved by 10% and the initial normalised crack length was 0.50. The specimen geometry of CT-NW-1 can be seen in Figure 4.16. The geometry of SENT-1 is shown in Figure 4.17 where the normalising crack length is 0.30, and no side-grooves are present.

4.8.3 Test Procedure

The tests were static load creep tests and performed at 550°C ±2°C. Potential drop (PD) input and output cables were spot welded onto the specimen’s surface in order to measure the crack length. Three thermocouples were welded to the specimen to monitor the temperatures in top, middle and bottom surfaces to avoid temperature deviations in space. An example set-up of a C(T) specimen is shown in Figure 4.18. The specimen was then housed within a three zone electrical resistance furnace. C(T) and SEN(T) specimens were tested on the dead weight level arm machines with a level arm ratio of 10:1. After a 12h preheating period, the sample was carefully loaded.

During the test, axial displacements were measured either by capacitance gauge (CG) or by LVDT. The signal from LVDT has a significant noise during a long term creep test; CG could generate more accurate results, as the measurement is local and does not require extensometer, though this precise measurement requires higher cost.
4.8.4 Crack Length Measurement

An electrical current was passed through the specimen such that a potential distribution forms across the specimen [64]. The potential difference was measured from the output leads positioned above and below the crack plane. The change in potential difference measured between the output leads could be related to the crack length with the use of a suitable calibration equation.

The calibration equation can be determined simply by correlating the initial and final output voltages with the measured initial and final crack length. Once the test was completed, the specimens were broken open and the initial $a_0$ and average failure $a_f$ crack length were measured. According to the known initial $V_0$ and final $V_f$ PD voltage, the instantaneous crack length $a$ was obtained by a linear calibration equation following ASTM E 1457 [51] as

$$a = a_0 + \frac{V_f - V_0}{V_f - V_0} (a_f - a_0)$$  \hspace{1cm} (4.7)

4.8.5 Crack Growth and Displacement Rate Calculations

The relationship between $i^\prime$ and $C^\prime$ can be determined if $i^\prime$ is known as described earlier in Section 2.4.4. According to ASTM E 1457 [51], the creep crack growth rate and displacement rate can be determined from the crack size $a$ and displacement $\Delta$ differentiated by time $t$ respectively. Recommended approaches that utilise the secant or incremental polynomial methods are given in Ref [51].

However, only the total experimental load-line displacement rate, $i^\prime$ can be measured in a test. The experimental displacement rate can be separated into an instantaneous part $i^\prime$ and a time dependent part that is directly associated with the accumulation of time dependent creep strains, $i^\prime$ [161] as

$$\dot{i} = \dot{i}_{\text{elastic}} + \dot{i}_{\text{plastic}}$$  \hspace{1cm} (4.8)

The instantaneous part can be further divided to an elastic and plastic part, $\dot{i}_{\text{elastic}}$ and $\dot{i}_{\text{plastic}}$. Plastic contribution may be neglected, and the elastic contribution was calculated in accordance with ASTM E 1457 [51] as
To ensure that the measured load line displacement is mainly due to the creep and not due to elastic deformation, the elastic part of the instantaneous load line displacement rate \( \dot{\epsilon} \) requires validation.

### 4.8.6 Initiation Time

The initiation time in terms of \( t_{0.2} \) and \( t_{0.5} \) has been described in Section 2.5.2. Recommended initiation time in ASTM E1457 [51] is \( t_{0.2} \) at \( \Delta a = 0.2 \) mm.

During this early period, the crack tip undergoes damage development and redistribution of stresses. The incubation period, prior to steady state creep, could be a substantial period of the test time [124, 162]. The initiation time can be expressed as an empirical law as [163]

\[
t_i = \frac{D_i}{C^{*\phi_i}}
\]  

(4.10)

where \( D_i \) and \( \phi_i \) are material constants related to initiation time.

### 4.8.7 Validity Criteria for \( C^* \)

Validity criteria are specified in ASTM E1457 [51] for CCG rate to be correlated with \( C^* \) parameter. Firstly, the material must be identified as ‘creep-ductile’ which is satisfied by the case as the creep load line displacement rate, calculated from Equation (4.8) and (4.9) constitutes at least half of the total load line displacement rate, \( \dot{\epsilon} \). For those data not satisfying this criterion are rejected. With the condition that \( \dot{\epsilon} \), the material can be classified as being ‘creep brittle’ in this case \( K \) is the recommended parameter.

The second validity criterion is to verify data for which the time exceeds a transition time, \( t_T \) as defined in Equation (2.54). Additionally, data points gathered prior to the time for 0.2mm crack extension, \( t_{0.2} \), should be excluded as they are considered to comprise the transient crack growth region where creep damage is building up to a steady state at the crack tip. Also, the data
acquired after the accumulated load line displacement, $\Delta LLD$ greater than 0.05 $W$ are also considered as invalid due to the additional bending moment due to the rotation of the arm.

R5 [57] suggests an additional criterion for $C^*$ correlation, this requires that the non-dimensional crack velocity, $\lambda$, is less than 0.5, where $\lambda$ is defined by

$$\lambda = \frac{i}{EC_{\text{exp}}}$$  \hspace{1cm} (4.11)

Note that the reference stress $\sigma_{\text{ref}}$ can be evaluated by Equation (2.10).

## 4.9 Tables

Table 4.1: Chemical Composition in wt% for the 316H material taken from the base metal of the header 1B1/1

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>S</th>
<th>P</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>Co</th>
<th>B</th>
<th>N</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.052</td>
<td>0.56</td>
<td>1.66</td>
<td>0.06</td>
<td>0.017</td>
<td>11.68</td>
<td>17.28</td>
<td>2.33</td>
<td>0.039</td>
<td>0.0045</td>
<td>0.036</td>
</tr>
</tbody>
</table>

Table 4.2: Test specimen matrix

<table>
<thead>
<tr>
<th>Test Type*</th>
<th>Test ID</th>
<th>Specimen Geometry</th>
<th>Temperature (°C)</th>
<th>$d$ (m m)</th>
<th>$W$ (mm)</th>
<th>$a_0 / W$</th>
</tr>
</thead>
<tbody>
<tr>
<td>T</td>
<td>RT-T-1</td>
<td>Tensile specimen</td>
<td>20</td>
<td>N/A</td>
<td>N/A</td>
<td>N/A</td>
</tr>
<tr>
<td>T</td>
<td>RT-T-2</td>
<td>Tensile specimen</td>
<td>20</td>
<td>N/A</td>
<td>N/A</td>
<td>N/A</td>
</tr>
<tr>
<td>T</td>
<td>RT-T-3</td>
<td>Tensile specimen</td>
<td>20</td>
<td>N/A</td>
<td>N/A</td>
<td>N/A</td>
</tr>
<tr>
<td>UC</td>
<td>UC-1</td>
<td>Uniaxial creep specimen</td>
<td>550</td>
<td>8</td>
<td>N/A</td>
<td>N/A</td>
</tr>
<tr>
<td>UC</td>
<td>UC-2</td>
<td>Uniaxial creep specimen</td>
<td>550</td>
<td>8</td>
<td>N/A</td>
<td>N/A</td>
</tr>
<tr>
<td>UC</td>
<td>UC-3</td>
<td>Uniaxial creep specimen</td>
<td>550</td>
<td>8</td>
<td>N/A</td>
<td>N/A</td>
</tr>
<tr>
<td>UC</td>
<td>UC-4</td>
<td>Uniaxial creep specimen</td>
<td>550</td>
<td>7</td>
<td>N/A</td>
<td>N/A</td>
</tr>
<tr>
<td>CCG</td>
<td>CT-1</td>
<td>C(T)</td>
<td>550</td>
<td>N/A</td>
<td>40</td>
<td>0.50</td>
</tr>
<tr>
<td>CCG</td>
<td>CT-2</td>
<td>C(T)</td>
<td>550</td>
<td>N/A</td>
<td>40</td>
<td>0.35</td>
</tr>
<tr>
<td>CCG</td>
<td>CT-NW-1</td>
<td>C(T)</td>
<td>550</td>
<td>N/A</td>
<td>40</td>
<td>0.50</td>
</tr>
<tr>
<td>CCG</td>
<td>SENT-1</td>
<td>SEN(T)</td>
<td>550</td>
<td>N/A</td>
<td>25</td>
<td>0.30</td>
</tr>
</tbody>
</table>

*Abbreviations: T: Tensile testing, UC: Uniaxial Creep, CCG: Creep Crack Growth Testing; XW Cross-Weld;
Table 4.3: Summary of collected data for different types of tests for comparison with this study

<table>
<thead>
<tr>
<th>Material</th>
<th>Header code</th>
<th>Cast</th>
<th>Test Type</th>
<th>UC: Initial Engineering Stress (MPa)</th>
<th>CCG: Initial $K \sqrt{m}$ (initial $a/W$)</th>
<th>Source</th>
</tr>
</thead>
<tbody>
<tr>
<td>316H SS (AR)</td>
<td>1B1/1</td>
<td>53415</td>
<td>T UC</td>
<td>257, 280, 290, 300, 335</td>
<td>N/A</td>
<td>IC [134]</td>
</tr>
<tr>
<td>316H SS (8% PC)</td>
<td>1B1/1</td>
<td>53415</td>
<td>T UC CCG in C(T)</td>
<td>257, 270, 280, 300</td>
<td>25.5(0.35) 25.0(0.50)</td>
<td>IC [134]</td>
</tr>
<tr>
<td>316H SS (AR)</td>
<td>1C2/3</td>
<td>55882</td>
<td>UC</td>
<td>280, 300</td>
<td>N/A</td>
<td>IC [147]</td>
</tr>
<tr>
<td>316H SS (AR)</td>
<td>2D2/2</td>
<td>55882</td>
<td>UC</td>
<td>290, 355</td>
<td>N/A</td>
<td>IC [147]</td>
</tr>
<tr>
<td>316H SS (8% PC)</td>
<td>2D2/2</td>
<td>55882</td>
<td>UC</td>
<td>239, 257, 280, 300</td>
<td>N/A</td>
<td>IC [147]</td>
</tr>
<tr>
<td>316H SS (AR)</td>
<td>2D2/2</td>
<td>55882</td>
<td>CCG in C(T)</td>
<td>N/A</td>
<td>42.3(0.53)</td>
<td>EDF [148]</td>
</tr>
<tr>
<td>316H SS (AR)</td>
<td>2D2/2</td>
<td>55882</td>
<td>CCG in C(T)</td>
<td>N/A</td>
<td>26.0(0.56)</td>
<td>IC[149]</td>
</tr>
<tr>
<td>316H SS (AR)</td>
<td>1C2/3</td>
<td></td>
<td>CCG in SEN(T)</td>
<td>N/A</td>
<td>N/A</td>
<td>IC [56]</td>
</tr>
<tr>
<td>316H SS (8% PC)</td>
<td>1D2/2</td>
<td>55882</td>
<td>CCG in C(T)</td>
<td>N/A</td>
<td>34.7(0.54)</td>
<td>IC[149]</td>
</tr>
<tr>
<td>316H SS (weldment)</td>
<td>-</td>
<td>-</td>
<td>CCG in C(T)</td>
<td>N/A</td>
<td>31.1(0.51) 29.9(0.52)</td>
<td>IC [150]</td>
</tr>
<tr>
<td>316H SS (XW)</td>
<td>-</td>
<td>-</td>
<td>T in DIC</td>
<td>N/A</td>
<td>N/A</td>
<td>Open University[151]</td>
</tr>
</tbody>
</table>
4.10 Figures

Figure 4.1: Illustration of the header 1B1/1, identifying the positions of the welds

Figure 4.2: Main dimensions of the welded steam nozzle and part of the steam header
Figure 4.3: A cross section of the nozzle/header branch indicating main dimensions and the position of the crack, base and weld metal.

Figure 4.4: Typical macro-graph of the cross-weld region.
Figure 4.5: The initial crack position in a C(T) specimen

Figure 4.6: The initial crack position in a SEN(T) specimen
Figure 4.7: Schematic drawing of the crack initiation position design for a C(T) specimen

Figure 4.8: Schematic illustration of the half-nozzle/header component, showing the designed extraction of testing specimens and specimen orientations

Figure 4.9: Schematic illustration of the location of three tensile specimens extracted from a weldment
Figure 4.10: The dimensions of a tensile specimen tested at room temperature

<table>
<thead>
<tr>
<th>Parameter (mm)</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>H</td>
<td>Gauge length</td>
</tr>
<tr>
<td>B</td>
<td>Specimen width</td>
</tr>
<tr>
<td>t</td>
<td>Specimen thickness</td>
</tr>
</tbody>
</table>

Figure 4.11: The location of the welds in the three tensile samples

Figure 4.12: Speckle patterns after painting in the three tensile samples
Figure 4.13: The dimensions of a uniaxial rupture specimen tested at high temperature

Figure 4.14: The location of weldments in the four uniaxial creep rupture specimens to be tested at 550°C
Figure 4.15: The dimensions of CT-1 and CT-2 extracted from an ex-service steam header

Figure 4.16: The dimension of CT-NW-1 sent from EDF energy

Figure 4.17: The dimensions of SENT-1 extracted from an ex-service steam header
Figure 4.18: An example set-up of a C(T) specimen to be performed a CCG test
Chapter 5

Tensile and Uniaxial Creep Rupture Behaviour

5.1 Introduction

Room temperature tensile tests performed by DIC and high temperature uniaxial creep rupture tests were conducted and the results are presented in this chapter to characterise the elastic-plastic or creep properties within the cross-weld region from an ex-service type 316H weld header. The results are discussed in order to examine possible variations due to the existence of welding residual stresses. Based on the previous findings in Ref [153, 154], it was found that pre-straining could have a similar effect on tensile and CCG behaviour to the effect of the welding process on pre-conditioning the HAZ. Therefore, the experimental tensile and uniaxial creep results were also compared to 8% plastic strain pre-compressed specimens as well as as-received materials.

5.2 Tensile Properties

Tensile tests were performed on type 316H stainless steel specimens consisting of base metal, weld metal and HAZ. DIC technique as described in Section 4.6 provides an accurate measurement in a full range of strain distributions. Therefore, the tensile properties within the regions of the cross-weld (XW), WM and BM can be individually obtained. In the present study, three specimens were under tensile tests at room temperature. Due to the similar behaviours from different specimens, the results in RT-T-1 are presented only. The high temperature tensile properties (545 °C) measured by DIC from uniaxial cross-weld tensile specimens on 316H stainless steel were taken from Ref [151].

5.2.1 Room Temperature Tensile Behaviour

During the tensile test, the strain distributions in the loading direction were synchronously measured by DIC. As can be seen in Figure 5.1, the strain in the same point was tracked during
the entire loading procedure until rupture. In the three tested specimens, fracture generally occurred in the weld region, close to WM and BM boundary, as presented in Figure 5.2.

The evolution of strain variation along a selected region was recorded during the tests. Figure 5.3 shows the strain variation in 25%, 50%, 75% and 100% of rupture time along a selected straight line which includes 68 strain sampling points. It can be seen that the strain in the HAZ region has a significantly drop, thus smaller than the strains in the neighbouring weld and base metal. A strain fluctuation within the weld metal can be observed, which could result from individual multi-weld passes during the welding procedure.

The strain was averaged among the values from several selected paths within the same regions. As seen in Figure 5.4, three paths have been selected to examine the material properties in BM and WM and two more paths have been included for XW strain examination. The stress vs. average strain curves containing scatter bands in XW, BM and WM regions are plotted in Figure 5.5. A linear line is fitted to the elastic region of each local stress-strain curve, which is used to determine the 0.2% proof stress and Young’s modulus. The key tensile parameters are provided in Table 5.1. In different regions, Young’s modulus has similar values while the 0.2% proof varies. Experimentally, 0.2% proof in XW region is around 50MPa higher and 20 MPa lower than BM and WM region respectively. The increase of 0.2% proof stress compared to as-received materials implies hardening effects would exist during welding.

5.2.2 High Temperature Tensile Behaviour

It was also shown in Ref [151] that at 545 °C, similar to room temperature tests, the specimen fractured in the weld metal, close to the boundary region. The strain variation along the gauge length of the specimen under an applied stress of 315 MPa was examined in a selected region. The results can be seen in Figure 5.6. Similar to room temperature strain distribution, the strains in HAZ have a lower value compared to those in BM and WM and strains in the WM fluctuate.

In Figure 5.7, local engineering and true stress-strain curves have been constructed in XW, BM and WM regions. The true stress-strain behaviour is determined from the engineering stress-strain curves in Ref [151]. It is found that the 0.2% proof stress of BM is considerably smaller than the weld metal and the HAZ. As tabulated in Table 5.1, the 0.2% proof stress of XW is 80
MPa and 18 MPa higher than BM and WM respectively. The behaviour of XW is close to WM, however, the approximately 50% increase in 0.2% proof stress in the XW reflects a stronger hardening effect at high temperature.

### 5.3 Uniaxial Creep Results

Four static load uniaxial creep tests have been performed on cross-weld uniaxial rupture specimens with initial engineering nominal stress ranging from 293 to 345MPa. The test procedure has been described in Section 4.7. Any results required from uniaxial creep rupture tests are presented in Table 5.2 and Table 5.3. Definitions of geometrical parameters, types of stress, strain and time which are mentioned in Table 5.2 and Table 5.3 were described in Section 4.7.4. It should be noted that all the discussions below are based on the true behaviour rather than engineering. The fracture positions of all specimens were in the weld metal and possibly near the boundary between the base and weld materials.

#### 5.3.1 Uniaxial Creep Rupture Behaviour

The experimental creep curves are given in Figure 5.8. Due to the stress dependency, creep deformations were accelerated with an increasing stress. Also seen in Figure 5.8, the creep failure strain can be relatively independent on the testing conditions. This observation is consistent with the previously investigations by Spindler [164], Bettinson [160] and Mehmanparast [134]. The behaviour of creep failure can be independently controlled by plastic hole growth [26] and creep cavity growth [27], which may result in an unpredictable creep ductility. According to Table 5.2, the average ductility in the range of nominal stress is obtained in both axial and ROA measurement where $\varepsilon_f$ (axial) = 5.8% and $\varepsilon_f$ (ROA) = 18.2%.

The relative proportion of the primary, secondary and tertiary stages in terms of time spent and strain accumulated are tabulated in Table 5.3 and illustrated in Figure 5.9 and Figure 5.10 respectively.

As obviously seen in Figure 5.9, secondary creep, as the most important creep stage, takes the largest proportion on the creep time duration. The percentage of time elapsed in secondary creep region increases as increasing nominal stresses. Meanwhile, time spent both in primary and
tertiary creep decreases. At the low stress, tertiary stage takes the proportion as large as secondary stage, which means the formation of fracture usually takes longer time than that at the high stress. The proportions both in primary and tertiary creep reduced at the high stress.

The percentage of creep strain measured in each stage (Figure 5.10) has the similar distribution as time duration. Decreasing nominal stress would result in a significantly decrease in the percentage of secondary creep while percentage of tertiary creep increases. The proportion of primary under the studied stress remains in a low level.

**5.3.2 Creep Properties**

For each specimen, secondary creep strain rate, $\dot{\varepsilon}_s$, average creep strain rate $\dot{\varepsilon}_A$ and time to rupture, $t_R$, are provided in Table 5.3. The secondary and average creep strain rate are plotted against true stress in log-log form as shown in Figure 5.11 and Figure 5.12 respectively, where the slope and intercept of the resulting regression lines can give the values of creep exponent ($n$ or $n_A$) and time dependent constant ($A$ or $A_A$). A similar method was used to fit the rupture properties according to Equation (2.16), true stress is plotted against the rupture time shown in Figure 5.13 from which the slope and intercept can be expressed in a form of $-1/\nu_k$ and $\log B_r/\nu_k$ respectively.

Two fitting lines fits to the data have been made, the red dash line stands for the best fitting straight line (regression line) and the black solid straight line is based on an assumption that the slope obtained in XW specimens has the same value as 8% pre-compressed specimens based on Ref [134] (discussed in Section 5.4.3). The two linear fitting lines are very close; therefore, both of them can reflect the real behaviours of creep in different true stresses. The values of creep power-law constants obtained in both methods are provided in these figures and summarised in Table 5.4.

**5.4 Comparison to PC and AR Specimens**

Previous study by Mehmanparast, A. et al. [134, 153, 154] has considered the pre-compression effect since it is thought to have a similar influence on tensile 0.2% proof stress and creep ductility to the effect of welding process on pre-conditioning the HAZ. For this reason, the
results from tensile and uniaxial creep rupture tests on the XW region were compared with pre-compressed (PC) to 8% plastic strain at room temperature as well as as-received (AR) materials.

### 5.4.1 Tensile Properties

The tensile properties on the PC and AR materials are provided in Table 5.1. According to room temperature tensile results, XW material has a hardening effect due to a 20% increase in 0.2% proof stress (385 MPa), however 0.2% proof stress in PC is 256 MPa which seems lower than that of AR of 313 MPa. As discussed in Ref [134], it is due to a combined isotropic/kinematic hardening behaviour exists in the PC material at room temperature [134].

High temperature 0.2% proof stress in XW is 275 MPa, which is similar to PC materials (259 MPa). Compared with that in BM (185 MPa), 50% increases in the XW and PC can be observed, the increasing rate is higher than that in room temperature. Therefore, high temperature XW and PC materials can result in a similar and stronger hardening effect, leading to a significant increase of 0.2% proof stress.

During pre-compression and welding process, it actually increases the number of dislocations in the crystal lattice of the material. When an environment filled with dislocations, the plastic deformation is hindered. The material will continue to perform an elastic deformation beyond the normal yield stress [165].

### 5.4.2 Creep Ductility

As can be seen in Figure 5.14 and Figure 5.15, the variation of axial and ROA creep ductility against true stress in XW specimens extracted from header A is compared to the PC and AR results available from header A,B,C [134]. As discussed in [134], for AR materials, the creep ductility measured by axial and ROA may be considered as stress independent within normalised stress range of $0.2 \leq \sigma / \sigma_{0.2} \leq 1.70$. The creep ductility in PC specimen on the normalised stress range of $1.06 \leq \sigma / \sigma_{0.2} \leq 1.15$ is somewhat dependent on the stress, especially measured by ROA.

The reasons have been discussed in Ref [134, 166] since the estimated trends in the uniaxial creep ductility at 550°C consist of an upper shelf of 13.6% for $\sigma / \sigma_{0.2} > 1.32$ and a lower shelf
of 0.9% for $\sigma / \sigma_{0.2} < 1.06$ with a stress dependent transition region in between. The relationship between $\sigma / \sigma_{0.2}$ and ductility for the PC, XW and AR materials are shown in Figure 5.16.

For the studied XW specimens, the applied stress is within the range of $1.06 < \sigma / \sigma_{0.2} < 1.32$, therefore, the creep ductility is within the transition region which depends on the stress. The creep ductility in the XW and PC materials is well described by the estimated trend of the AR material. Therefore, for a given value of normalised stress, the creep ductility would provide similar trends among the XW, PC and AR materials as hardening effects allow to be considered. The short term of XW and PC materials therefore may estimate creep ductility of long term AR materials.

For 316H stainless weld header, uniaxial creep ductility under a long term, low stress service condition is assumed to be in the lower bound. The extrapolated figure of 0.9% for the lower shelf bound ductility at 550 °C has been found to be consistent with long term experimental data (>50000h) for the same materials [148]. Therefore the lower bound creep ductility behaviour and its sensitivity to creep crack growth has been applied in the FE analysis as discussed in Chapter 7.

### 5.4.3 Creep Properties

As seen in Figure 5.17, Figure 5.18, the secondary creep strain rate $\dot{\varepsilon}_s$ and average creep strain rate $\dot{\varepsilon}$, are plotted against the true stress respectively. Figure 5.19 shows the relationship between rupture time, $t_a$, and the true stress. The comparison was made among the XW, AR and PC uniaxial specimens extracted from the same or different headers. As examined in Section 5.3.2, the straight fitting lines from regression lines or based on a given slope obtained from PC specimens in header A are close. In order to make a direct and clear comparison among different specimens, the results obtained from Ref [134] have been recalculated based on the assumption that the slope from the different resources are same as those of obtained from PC specimens in header A. The creep power-law constants determined from the fitting lines with same slopes are summarised in Table 5.4 and also provided in Figure 5.17, Figure 5.18 and Figure 5.19 respectively.
According to Figure 5.17, the secondary creep strain rate $\dot{\varepsilon}$ behaviour of the XW specimen has been studied firstly and compared to the AR and PC specimens for a range of true stress. $\dot{\varepsilon}$ in the XW specimens is slightly higher than $\dot{\varepsilon}$ in the AR specimens from the same header. It may imply that the existence of welding residual stress in the XW region accelerates the creep deformation. $\dot{\varepsilon}$ is also slightly higher than the PC specimens, though the variations are still not significant.

The average creep strain rate, $\dot{\varepsilon}$ behaviour on the XW, AR and PC specimens from different sources are presented in Figure 5.18 for a range of true stresses, the average creep strain rate is determined by axial creep displacement measurement and is related to the creep damage properties. All the data collected in the header A has been compared and it is found that the $\dot{\varepsilon}$ examined from the XW, AR and PC can be plotted with overlapping fitting lines. Therefore, it may be implied that the $\dot{\varepsilon}$ is somewhat insensitive to the material’s mechanical and microstructural changes due to welding and pre-straining effect, therefore, the $\dot{\varepsilon}$ can be well characterised under the studied range of stress.

Finally, the plot of true stress vs. time to rupture $t_R$ for the XW, AR and PC specimens are shown in Figure 5.19. Clearly for a given applied stress, a significant reduction $t_R$ has been obtained in the XW specimens compared to the AR material and the uniaxial creep rupture trend for the XW is similar to that of obtained for the PC material even for different headers. $t_R$ of the XW is reduced up to average three times compared with that of the AR. It implies that the creep-induced cavities at grain boundaries may easily be initiated in the XW region due to the existence of residual stress. The growth and coalescence of cavities may accelerate the failure procedure and result in a reduction of $t_R$.

In conclusion, the XW specimens have similar $\dot{\varepsilon}$ and $\dot{\varepsilon}$ behaviour as the AR specimens, however $t_R$ reduces to an average one third of that in the AR specimens. 8% plastic strain at room temperatures has been introduced in the PC specimen, this specimens has proved to be similar to the creep rupture behaviour of the XW specimens in the studied header.
5.4.4 Summary

Experiments were performed on cross-weld specimens extracted from an ex-service header to determine the tensile and creep rupture behaviour of 316H stainless steel welded components in order to consider the effects of welding residual stresses. The results from these tests were compared to those available on as-received and 8% plastic strain pre-compressed specimens extracted from the same ex-service steam header as well as from other sources. According to the comparisons and analysis, some of main findings are summarised as follows:

1. The XW and PC materials have similar room and high temperature elastic-plastic properties as they can harden the materials. Compared to the AR materials, the XW materials have 20% and 50% increase of 0.2% proof stress at room and high temperature respectively.

2. In tensile tests, plastic deformation occurs earlier in the AR than XW and PC specimens due to the hardening effects on the XW and PC materials.

3. The creep ductility in the XW and PC materials can be well described by the estimated trend of the AR specimen. In 316H stainless steel components serviced in a low stress and long time, the estimated creep ductility at 550 °C is under a lower shelf bound. The XW specimens have a similar average creep strain rate, close but slightly higher secondary creep strain rate compared to the AR specimens, however the time to rupture in XW is an average three times shorter for a given value of stress. This may due to creep-induced cavities initiated in weldments that accelerate the failure procedure.

4. Cross-weld specimens show similar creep behaviour in terms of secondary creep strain rate, average creep strain rate, time to rupture, when compared to specimen with pre-compressed 8% plastic strain at room temperature.
### 5.5 Tables

Table 5.1: Tensile properties at room temperature and high temperature

<table>
<thead>
<tr>
<th>Material</th>
<th>Temperature</th>
<th>$E$ (GPa)</th>
<th>$\sigma_{0.2}$ (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>XW</td>
<td>205</td>
<td>385</td>
<td></td>
</tr>
<tr>
<td>BM</td>
<td>205</td>
<td>331</td>
<td></td>
</tr>
<tr>
<td>WM</td>
<td>Room</td>
<td>205</td>
<td>404</td>
</tr>
<tr>
<td>AR [134]</td>
<td>205</td>
<td>313</td>
<td></td>
</tr>
<tr>
<td>PC [134]</td>
<td>205</td>
<td>256</td>
<td></td>
</tr>
<tr>
<td>XW [167]</td>
<td>155</td>
<td>275</td>
<td></td>
</tr>
<tr>
<td>BM [167]</td>
<td>545°C</td>
<td>140</td>
<td>185</td>
</tr>
<tr>
<td>WM [167]</td>
<td>150</td>
<td>257</td>
<td></td>
</tr>
<tr>
<td>AR [134]</td>
<td>550°C</td>
<td>140</td>
<td>177</td>
</tr>
<tr>
<td>PC [134]</td>
<td>140</td>
<td>259</td>
<td></td>
</tr>
</tbody>
</table>

Table 5.2: Uniaxial creep rupture test results

<table>
<thead>
<tr>
<th>Test ID</th>
<th>Material Condition</th>
<th>$d$ (mm)</th>
<th>$l_0$ (mm)</th>
<th>$\sigma_{nom}$ (MPa)</th>
<th>$\sigma_{true}$ (MPa)</th>
<th>$\epsilon_{eng}^{true}$ Axial %</th>
<th>$\epsilon_{eng,load}$ Axial %</th>
<th>$\epsilon_{true,load}$ Axial %</th>
<th>$\epsilon_{true,load}$ ROA %</th>
<th>$\epsilon_{true,load}$ ROA %</th>
</tr>
</thead>
<tbody>
<tr>
<td>UC-1</td>
<td>XW</td>
<td>8.05</td>
<td>35.45</td>
<td>290</td>
<td>293</td>
<td>1.12</td>
<td>1.11</td>
<td>0.34</td>
<td>4.62</td>
<td>4.52</td>
</tr>
<tr>
<td>UC-2</td>
<td>XW</td>
<td>8.07</td>
<td>35.90</td>
<td>325</td>
<td>338</td>
<td>4.06</td>
<td>3.98</td>
<td>2.83</td>
<td>8.18</td>
<td>7.94</td>
</tr>
<tr>
<td>UC-3</td>
<td>XW</td>
<td>8.06</td>
<td>35.95</td>
<td>345</td>
<td>361</td>
<td>4.54</td>
<td>4.44</td>
<td>3.37</td>
<td>5.42</td>
<td>5.28</td>
</tr>
<tr>
<td>UC-4</td>
<td>XW</td>
<td>7.01</td>
<td>35.91</td>
<td>300</td>
<td>306</td>
<td>2.27</td>
<td>2.24</td>
<td>1.26</td>
<td>5.50</td>
<td>5.39</td>
</tr>
</tbody>
</table>

Table 5.3: Uniaxial creep rupture test results continued to Table 5.2, including relative percentage of strain accumulated and time spent in each creep stage

<table>
<thead>
<tr>
<th>Test ID</th>
<th>Material Condition</th>
<th>$\sigma_{nom}$ (MPa)</th>
<th>$\epsilon_f^{true}$ Axial %</th>
<th>$t_r$ (h)</th>
<th>$\dot{\epsilon}$ (h$^{-1}$)</th>
<th>$\dot{\epsilon}^{true}$ Axial %</th>
<th>$\dot{\epsilon}^{true}$ ROA %</th>
<th>$t_{Pri}$ %</th>
<th>$t_{Sec}$ %</th>
<th>$t_{Ter}$ %</th>
</tr>
</thead>
<tbody>
<tr>
<td>UC-1</td>
<td>XW</td>
<td>290</td>
<td>4.62</td>
<td>1147.5</td>
<td>0.23×10$^{-4}$</td>
<td>0.40×10$^{-4}$</td>
<td>17.2</td>
<td>26.5</td>
<td>56.3</td>
<td>25.2</td>
</tr>
<tr>
<td>UC-2</td>
<td>XW</td>
<td>325</td>
<td>8.18</td>
<td>369.3</td>
<td>1.20×10$^{-4}$</td>
<td>2.21×10$^{-4}$</td>
<td>22.3</td>
<td>34.7</td>
<td>43.0</td>
<td>21.8</td>
</tr>
<tr>
<td>UC-3</td>
<td>XW</td>
<td>345</td>
<td>5.42</td>
<td>137.7</td>
<td>3.54×10$^{-4}$</td>
<td>3.93×10$^{-4}$</td>
<td>11.2</td>
<td>46.3</td>
<td>42.5</td>
<td>20.3</td>
</tr>
<tr>
<td>UC-4</td>
<td>XW</td>
<td>300</td>
<td>5.50</td>
<td>585.4</td>
<td>0.67×10$^{-4}$</td>
<td>0.94×10$^{-4}$</td>
<td>7.1</td>
<td>69.4</td>
<td>23.5</td>
<td>11.1</td>
</tr>
</tbody>
</table>
Table 5.4: Summary of power-law constants for each type of specimens

<table>
<thead>
<tr>
<th>Material Condition</th>
<th>Fitting</th>
<th>$A$ (MPa·h$^{-1}$)</th>
<th>$n$</th>
<th>$A_d$ (MPa·h$^{-1}$)</th>
<th>$n_d$</th>
<th>$B_r$ (MPa·h)</th>
<th>$v_r$</th>
</tr>
</thead>
<tbody>
<tr>
<td>XW-Header A</td>
<td>Regression line</td>
<td>3.75×10^{-34}</td>
<td>11.70</td>
<td>1.02×10^{-30}</td>
<td>10.41</td>
<td>4.17×10^{-25}</td>
<td>9.16</td>
</tr>
<tr>
<td>XW-Header A</td>
<td>Fixed slope</td>
<td>3.96×10^{-32}</td>
<td>10.9</td>
<td>1.00×10^{-39}</td>
<td>14.0</td>
<td>5.13×10^{-22}</td>
<td>8.0</td>
</tr>
<tr>
<td>AR-Header A</td>
<td>Fixed slope</td>
<td>2.68×10^{-32}</td>
<td>10.9</td>
<td>9.82×10^{-40}</td>
<td>14.0</td>
<td>6.61×10^{-22}</td>
<td>8.0</td>
</tr>
<tr>
<td>AR-Header B and C</td>
<td>Fixed slope</td>
<td>6.60×10^{-33}</td>
<td>10.9</td>
<td>2.09×10^{-40}</td>
<td>14.0</td>
<td>2.86×10^{-23}</td>
<td>8.0</td>
</tr>
<tr>
<td>8% PC-Header A</td>
<td>Fixed slope</td>
<td>2.22×10^{-32}</td>
<td>10.9</td>
<td>8.97×10^{-39}</td>
<td>14.0</td>
<td>5.45×10^{-22}</td>
<td>8.0</td>
</tr>
<tr>
<td>8% PC-Header C</td>
<td>Fixed slope</td>
<td>2.70×10^{-32}</td>
<td>10.9</td>
<td>1.18×10^{-39}</td>
<td>14.0</td>
<td>5.50×10^{-22}</td>
<td>8.0</td>
</tr>
</tbody>
</table>

5.6 Figures

Percentage of rupture time:
30% 40% 50% 60% 70% 80% 90% 100%

$\varepsilon_{22}(\text{eng})$ (%)
0 5 10 15 20 25 30 35 40 45 50

Testing time:
80s 107s 134s 160s 187s 214s 240s 267s

Figure 5.1: The engineering strain distribution in loading direction in different percentage of rupture time from DIC measurement
RT-T-1:

RT-T-2:

RT-T-3:

Figure 5.2: Fractural positions in the three tested specimens

Figure 5.3: Engineering strain distributions in different test stages along a selected region from room temperature DIC measurement
Figure 5.4: Selected paths for strain examination within (a) BM (b) WM (c) cross-weld region

Figure 5.5: Average tensile response from DIC measurement in different regions at room temperature based on (a) engineering (b) true stress and strain
Figure 5.5(continued): Average tensile response from DIC measurement in different regions at room temperature based on (a) engineering (b) true stress and strain.

Figure 5.6 Engineering strain distribution along a selected section at 545 °C from DIC measurement.
Figure 5.7: Tensile response from DIC measurement in different regions at high temperature based on (a) engineering (b) true stress and strain
Figure 5.8: Experimental creep curves on the XW specimens

Figure 5.9: Relative percentage of time spent in each creep state in the XW specimens
Figure 5.10: Relative percentage of strain accumulated in each creep state in the XW specimens

Figure 5.11: True net section stress vs. secondary creep strain rate in the XW specimens
Figure 5.12: True net section stress vs. average creep strain rate in the XW specimens

Figure 5.13: Time to rupture vs. true net section stress in the XW specimens
Figure 5.14: Axial creep ductility vs. true stress for the XW, AR and PC materials

Figure 5.15: ROA creep ductility vs. true stress for the XW, AR and PC materials
Figure 5.16: Comparison of the creep ductility among the XW, PC and AR materials to the estimated trends for the AR material at 550 °C extracted from a number of headers.

Figure 5.17: Comparison of the secondary creep strain rate among the XW, PC and AR extracted from a number of headers.
Figure 5.18: Comparison of the average creep strain rate among the XW, PC and AR extracted from a number of headers

Figure 5.19: Comparison of the time to rupture among the XW, PC and AR extracted from a number of headers
Chapter 6

Creep Crack Initiation and Growth Behaviour

6.1 Introduction

Experimental creep crack initiation and growth tests of macroscopically homogeneous materials are usually carried out following standard test method ASTM E 1457 [51]. In this standard, a C(T) is a commonly used geometry as it introduces high constraint so that plane strain conditions are promoted near the crack tip [21]. The resulting benchmark crack growth data are applied in assessment procedures and correlated with the fracture mechanics parameter, $C^*$, to predict creep crack initiation and growth behaviour of structural components [150].

The nature of residual stresses in welded components was discussed in Section 2.7 in terms of their role in the structure, special distribution, magnitude and measurement. The effects of residual stress on the crack initiation and propagation are discussed in this Chapter. In the previous work performed at Imperial College London, Neutron diffraction measurements have been performed along the expected crack path to quantify the residual stresses distribution in the 316H stainless steel Manual Metal Arc (MMA) weldment sections. The weld induced stresses are up to three times the room temperature yield stress of the material, which could influence the short term crack initiation and long term creep crack growth behaviour[168].

Testing of creep crack growths were performed in order to obtain the knowledge on the material cracking behaviour due to the existing of residual stresses. The initial crack was also designed to replicate to the region of cracks in the header which was also near, but not on the fusion line. To maximise the information on weldments, an additional CCG test was performed with an initial crack located within the fusion line. Since there are currently no agreed testing standards for CCI and CCG for inhomogeneous materials, such as weldments, the present study aims to examine the application of procedures recommended in ASTM E 1457 for the characterisation of crack growth in weldments.
As mentioned in Chapter 3, alternative laboratory geometries apart from C(T) specimen have been proved to be suitable for investigating various crack tip constraint configurations. Therefore, the SEN(T) specimen was also under investigation, the results was then compared to standardised C(T) specimens.

According to experimental testing procedures and the methods to analyse the creep crack initiation and growth data introduced in Section 4.8, the results from the austenitic type 316H stainless steel specimens, tested at 550°C are discussed below. The main objective of this chapter is to characterise the general behaviour of the CCI and CCG data through various testing results. The limited results obtained from the present study were compared to pre-compressed, as-received and as-welded CCG data from different sources mainly performed at same laboratories at Imperial College London [56, 134, 148-150] as can be seen in Table 4.3.

### 6.2 Specimen Description

Three C(T) specimens (specimen width, $W = 40$ mm) and one SEN(T) ($W = 25$ mm) specimen have been tested. A schematic illustration of the C(T) specimen, showing the two types of initial crack positions and the regions of the BM, WM and HAZ is shown in Figure 6.1. The CCG tests on these specimens were also compared to C(T) specimens from different material conditions. Those specimens include three pre-compressed materials with 8% plastic strain at room temperature [134, 149], two specimens with an initial crack in the HAZ [150].

The specimen geometries and material conditions for each test specimen are described in Table 6.1. For C(T) specimens, two specimen widths were generally used with $W = 40, 50$ mm respectively. The initial normalised crack length was mostly around 0.50, with two exceptional cases in CT-2 and 8PC-A2. The crack in CT-2 did not propagate after a long testing time due to the lower initial $K$ and crack length. For the rest of the specimens, the loading conditions ($P$ and $K(a_0)$), time to failure ($t_f$), amount of crack extension ($\Delta a$) and initiation time for 0.2 mm crack extension ($t_{0.2}$) are detailed in Table 6.2. All creep tests were performed at 550 °C.
6.3 Crack Length and Displacement Measurements

6.3.1 Load Line Displacements

In Figure 6.2, the load line displacement (LLD) normalised by the specimen width is plotted against the test time normalised by the test duration. Three types of specimens are illustrated in different colours as (1) black: specimens performed in this study; (2) red: pre-compressed specimens [134] (3) blue: initial cracks in the HAZ [150]. All the data have similar trends as a decreasing displacement rate in the initial period, followed by a steady crack growth rate and a rapid growth rate in the end of tests. As seen in Figure 6.2, the normalised load line displacements in CT-1 and SENT-1 are lower than other types of specimens, since the crack lies within the base metal, LLD behaviour is likely to be similar to the AR specimens as AR may provide a lower LLD value than PC and HAZ specimens [153]. Usually, tests were terminated with faster growth rate, the total LLD depends on the test termination points.

6.3.2 Crack Growth

In Figure 6.3, the crack extension, $\Delta a$ is plotted against the normalised test time for the same specimens as shown in Figure 6.2. The time to attain the initiation distance as $\Delta a = 0.2$ mm is shown in this figure. In most specimens, there is a large proportion of crack initiation periods as $\Delta a < 0.2$ mm; the reason may be due to the blunting effects during the strain accumulation at the crack tip [21]. According to the validity criterion described in Section 4.8.7, the amount of crack growth that occurs for $t < t_f$ or $t < t_{0.2}$ should excluded for $C^*$ correlation. Similar to LLD, the total crack extension is dependent on the points at which test was stopped. Generally, a smooth and slow trend towards the end of tests was observed in CT-1 and SENT-1 compared with large and rapid crack length extensions in the HAZ and PC specimens. The crack extension from CT-NW-1 is consistent with XW-6 due to the similar material, geometrical and testing conditions.

6.4 Analysis of Validity Criterion

The $C^*$ validation is described in Section 4.8.7 and summarised in Table 6.3. As seen in Table 6.3, $\lambda$ and $\Delta_{LLD}/W$ criterion are satisfied by all specimens. Since transition time $t_f$ occurs
earlier than $t_{0.2}$ in most tests, only $t > t_{0.2}$ requires considerations. In Figure 6.4, the ratio of creep to total load line displacement rate is plotted against the time normalised by testing time. The valid data as $t > t_{0.2}$ or $\dot{\epsilon}_c > \dot{\epsilon}_t$ (assuming $\dot{\epsilon}_c > \dot{\epsilon}_t$) are plotted with solid lines whereas any invalid data points are shaded. A large proportion of data remains valid under the considerations of validity criterion, therefore, there is a good correlation between the crack length rate and $C^*$ in the studied specimens.

### 6.5 Creep Crack Growth Rate Results

In Figure 6.5, the CCG rate is correlated against $C^*$, calculated from $\Delta^{LLD}$ for those specimens considered above. This figure includes a set of test data in black dot from as-received 316H stainless steel C(T) specimen of different size taken from [160] and the data band in red dots from C(T) HAZ data taken from [150]. Following the $C^*$ validity criterion, only valid data have been considered.

As can be seen in Figure 6.5, CT-NW-1 lies within the data set of weldment due to the similar material and testing conditions as Ref [150], whereas the initial cracks in CT-1 and SENT-1 were within base metal C(T) data bands, indicating that CCG behaviour was similar to the base materials. The valid data points from CCG tests performed in this project were consistent with the data bands, indicating the possibility to discuss the CCG rate under different geometry and material conditions. Therefore, a wider range of data sources are employed in the following discussions.

According to Ref [148], the available creep crack growth data for type 316H steel base metal is summarised in Figure 6.6. Those data are bounded using the current upper bound creep crack growth law for type 316H steels (identified as R66 upper bound [169]). The upper bound is not-conservative at the lower value of $C^*$ but sufficient bounds the data at higher $C^*$. The data bands shown in Figure 6.6 describes the CCG data performed on a range of C(T) specimen sizes ($W = 100$, 50 and 25 mm) under a short term test for homogeneous base materials. All the short term test data shown in Figure 6.6 generally lies within the data bands, including CT-1. The long term test data with the lower value of $C^*$ tend towards the C(T) data band.
The CCG rate data in the HAZ according to Ref [150] are again plotted against $C^*$ shown in Figure 6.7. As a given $C^*$, the CCG rate in the HAZ is higher than that in the BM. According to the explanation in Ref [153], the loss of crack tip constraint through plasticity effects and the increase of creep ductility in BM specimen would result in a decrease of CCG rate.

Compared with the data shown between Figure 6.6 and Figure 6.7, the slope of the data in base metal is less than that in the HAZ. On the average trends, the CCG rate data from HAZ are approximately a factor of seven times higher than those from the short term AR material, for a given value of $C^*$[153]. It means that the creep ductility and specimen constraint effects may also bring an increase in creep crack growth rate.

In Figure 6.5, the CCG data on 8% pre-compressed C(T) specimens [134, 153, 154] is compared with the HAZ data band. The CCG behaviour of HAZ and PC data are effectively coincident. The PC data lie within the HAZ data band and the slope of CCG in PC and HAZ may be considered similar, meaning that there is a similar crack growth rate. It proved the previous work in [153] that CCG data in PC specimen may be used to estimate the HAZ behaviour.

Finally, the valid CCG data (also taken from Ref [56]) on as-received SEN(T) specimens are compared to the standardised C(T) specimen. From Figure 6.8, the data of SEN(T) for $C^* > 1 \times 10^{-5}$ MPam/h fall within the base metal C(T) data band, which indicates that CCG behaviour is not constrained on the fracture geometry. It has also proved in Ref [56] which has investigated a series of fracture mechanics geometries.

### 6.6 CCG Prediction in NSW model

The data considered were then compared with the NSW-CCG models described in Section 2.5.1. The analytical NSW model can reduce the testing procedure and the number of tests required to describe the material properties [150]. An outline of NSW models include original NSW model based on ductility exhaustion, approximate NSW (NSWA) model and the modified NSW model were used in the current work. In previous examination [134, 170], NSW RUP prediction as described in Equation (2.63) can be reduced to the form of NSW model based on the assumption that $n_d \approx V_r$, therefore, it is not considered here.
In Figure 6.9, the secondary creep stress exponents \((n)\) were employed to discuss the CCG predicted by NSW, NSWA and NSW-MOD model for the base metal and HAZ materials. The values used in NSW prediction have been provided in Table 6.4 where the data in the AR materials were according to Ref [56] but were reanalysed in the present work. The failure strain, \(\varepsilon_f\) used in all models was the uniaxial failure strain based on the reduction of area (ROA) which was measured before and after the tests.

Firstly, the available experimental CCG data on the HAZ and PC specimens were compared to the plane stress and plane strain prediction based on the steady state creep strain rate, \(\dot{\varepsilon}^s\). The CCG data for specimens with the HAZ and PC materials generally fall between plane stress and plane strain prediction lines using NSW and NSWA models as shown in the red region in Figure 6.9 (a) and (b). The prediction in NSW and NSWA provided larger scatter band between an upper band controlled by plane strain condition and lower band controlled by plane stress condition, as the difference between plane stress and plane strain CCG for the creep ductile materials was obtained to be a factor of 30. For the NSW-MOD model shown in Figure 6.9 (c), the scatter band for \(n \approx 12\) would be reduced to around a factor of 3.9. In this case, except a small range of ‘tail’ region, most of CCG data fell between the plane stress and plane strain region of the tested data for the HAZ and PC materials. The NSW-MOD model, as an improved prediction, provided a better and less conservative representation of experimental data bands for the current analysis of the HAZ and PC materials.

CCG rate predictions on the AR specimens in NSW model were then discussed. In this study, plane stress and plane strain predictions were limited to \(\dot{\varepsilon}^s\). As shown in Figure 6.9 (a-c), the data in CT-1 and SENT-1 generally fall slightly lower than the plane stress \(\dot{\varepsilon}^s\), suggesting plane stress \(\dot{\varepsilon}^s\) is a good prediction of CCG rate. In NSW-MOD model as illustrated in Figure 6.9 (c), the \(\dot{\varepsilon}^s\) in plane stress conditions is seen to provide an approximate mean fit to the tested data, which provides a good representation of experimental data bands for the current data set. The lower bound of CCG rate requires consideration of the initial CCG rate, \(\dot{\varepsilon}^0\), which is not described here.
6.7 Crack Morphology

The microstructural observation in the micro-cracking area ahead of the main crack tip in SENT-1 is shown in Figure 6.10. In the vicinity of the starter crack, a main continuous crack can be observed with a length of around 800 µm. Ahead of the main crack, a branched cracking morphology and creep cavities in the area near micro-cracks occurs and the microcracks are mostly in a direction normal to the main crack. The voids mainly nucleate and grow on the grain boundary and the well-developed cavities are then to be coalesced to form a grain-size microcrack, the coalescence of microcracks leads to the growth of creep-induced crack [171]. The branched cracking can be either transgranular or intergranular which could be generated by the presence of tensile stress state at a moving crack tip. Near the end of the cracks, some discontinuous micro-cracks in the creep damage zone are presented in Figure 6.10, which is also near the main cracks, the micro-cracks can join up and lead to a larger crack length.

6.8 Creep Crack Initiation

In Figure 6.11(a) and (b), the crack initiation times for crack growth extension of 0.2 mm and 0.5 mm, denoted as \( t_{0.2} \) and \( t_{0.5} \) have also been examined. It has been agreed that crack initiation occurs under steady state creep condition under which \( C'(t_i) \) parameter may be used to correlate the CCG behaviour. The trends of \( C'(t_i) \) on \( t_{0.2} \) and \( t_{0.5} \) have been obtained.

As seen in Figure 6.11, although there is a large degree of scatter for each type of specimen at both \( t_{0.2} \) and \( t_{0.5} \), the initiation time in CT-1 and SENT-1 may lie with the scatter band in AR materials whilst data in CT-NW-1 may fall in the region of HAZ. Due to the limited sources, further tests are still required to prove the findings. In Figure 6.11, the trends of creep crack initiation of the HAZ and AR materials are provided in black and red lines respectively, as a given \( C'(t_i) \). The creep crack growth in the HAZ is lower than the AR materials, meaning that the crack could initiate at a shorter time in the HAZ materials than the AR materials. The main reason is that defects may exist during welding and subsequent high temperature operation; those defects may result in the easily formation of large cavities, micro-cracks and creep macro-cracks.
6.9 Summary

Creep crack initiation and growth tests were carried out on C(T) and SEN(T) specimen with two types of initial crack position in weldment, one is within the HAZ and another lies within the base metal, near but not on the HAZ. The results of CCI and CCG were also compared to any available data for as-received, pre-compressed and as-welded materials. Some of important findings are pointed out below:

1. When the cracks initiate in the base metal, near but not on the weld line, the CCI and CCG behaviour is similar to the as-received materials. The welding procedure may not affect the creep crack behaviour on the surrounding materials.

2. CCG data for cracks initiate in the HAZ are well correlated in $C^*$, suggesting standardised procedure for homogeneous materials in ASTM E 1457 are consistent to weldments.

3. CCG behaviour in the HAZ and PC are similar; the CCG data for the HAZ and PC are higher than that in the AR materials due to the highly crack tip constraint and reduced creep ductility.

4. CCG data correlated by $C^*$ for different specimen geometries and sizes are similar to the data band of BM C(T) specimen, indicating $C^*$ prediction can be well applied in different geometries.

5. CCG behaviour in the HAZ, PC and AR materials can be well predicted by NSW model. NSW-MOD model provides better and less-conservative prediction than NSW and NSWA model when creep ductility in ROA is employed in calculations.

6. In the CCI behaviour correlated by $C^*$ for each type of specimen there exists a significant scatter, for a given value of $C^*$, the initiation time in the HAZ is around an order of magnitude shorter than that in the AR material since the defects are easily generated during welding and pre-processing in the HAZ.
### 6.10 Tables

#### Table 6.1: Specimen geometry

<table>
<thead>
<tr>
<th>Test ID</th>
<th>Specimen Geometry</th>
<th>Crack Condition</th>
<th>$W$ (mm)</th>
<th>$B$ (mm)</th>
<th>$B_n$ (mm)</th>
<th>$a_0 / W$</th>
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</thead>
<tbody>
<tr>
<td>CT-1</td>
<td>C(T)</td>
<td>BM near HAZ</td>
<td>40</td>
<td>20</td>
<td>14.1</td>
<td>0.53</td>
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<tr>
<td>CT-2</td>
<td>C(T)</td>
<td>BM near HAZ</td>
<td>40</td>
<td>20</td>
<td>14.0</td>
<td>0.35</td>
</tr>
<tr>
<td>CT-NW-1</td>
<td>C(T)</td>
<td>HAZ</td>
<td>37.9</td>
<td>19.0</td>
<td>14.9</td>
<td>0.50</td>
</tr>
<tr>
<td>SENT-1</td>
<td>SEN(T)</td>
<td>BM near HAZ</td>
<td>25</td>
<td>12.5</td>
<td>12.5</td>
<td>0.30</td>
</tr>
<tr>
<td>8PC-A2[134]</td>
<td>C(T)</td>
<td>8% PC</td>
<td>50</td>
<td>25</td>
<td>17.5</td>
<td>0.35</td>
</tr>
<tr>
<td>8PC-A4[134]</td>
<td>C(T)</td>
<td>8% PC</td>
<td>50</td>
<td>25</td>
<td>17.5</td>
<td>0.50</td>
</tr>
<tr>
<td>8PC-D2[149]</td>
<td>C(T)</td>
<td>8% PC</td>
<td>50</td>
<td>25</td>
<td>20.0</td>
<td>0.50</td>
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<tr>
<td>XW-5[150]</td>
<td>C(T)</td>
<td>HAZ</td>
<td>50</td>
<td>25</td>
<td>20</td>
<td>0.51</td>
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<tr>
<td>XW-6[150]</td>
<td>C(T)</td>
<td>HAZ</td>
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<td>25</td>
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<td>0.52</td>
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#### Table 6.2: Specimen test details

<table>
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<tr>
<th>Test ID</th>
<th>Load, $P$ (kN)</th>
<th>$K(a_0)$ (MPa$\sqrt{m}$)</th>
<th>$t_f$ (h)</th>
<th>$\Delta a$ (mm)</th>
<th>$a_f / W$</th>
<th>$t_T / t_f$ (%)</th>
<th>$t_{0.2} / t_f$ (%)</th>
<th>$t_{0.5} / t_f$ (%)</th>
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<tr>
<td>CT-1</td>
<td>10.3</td>
<td>32.5</td>
<td>1766</td>
<td>8.4</td>
<td>0.71</td>
<td>9.5</td>
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<td>CT-NW-1</td>
<td>10.2</td>
<td>30</td>
<td>504</td>
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<td>14.8</td>
<td>28.0</td>
<td>39.6</td>
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<td>36.9</td>
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<td>360</td>
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<td>7.2</td>
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<td>8PC-D2[149]</td>
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<td>11.9</td>
<td>0.74</td>
<td>5.5</td>
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<td>XW-5[150]</td>
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<td>31.1</td>
<td>718</td>
<td>11.8</td>
<td>0.75</td>
<td>7.8</td>
<td>37.0</td>
<td>-</td>
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<td>XW-6[150]</td>
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<td>29.9</td>
<td>736</td>
<td>17.3</td>
<td>0.87</td>
<td>3.8</td>
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Table 6.3: $C^*$ validity criterion according to ASTM E 1457 and R5

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<tr>
<th>Test ID</th>
<th>$t_f$ (h)</th>
<th>$t_{0.2}$ (h)</th>
<th>$t_T$ (h)</th>
<th>Time for $\lambda &lt; 0.5$ PE (h)</th>
<th>Time for $i$ PE (h)</th>
<th>Time for $\Delta^{LLD} / W \geq 0.05$ PE (h)</th>
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<tbody>
<tr>
<td>CT-1</td>
<td>1766</td>
<td>168</td>
<td>366</td>
<td>1766</td>
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<tr>
<td>CT-NW-1</td>
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<td>75</td>
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<td>458</td>
<td>504</td>
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<td>SENT-1</td>
<td>360</td>
<td>26</td>
<td>16</td>
<td>360</td>
<td>360</td>
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<td>8PC-A4[134]</td>
<td>194</td>
<td>8</td>
<td>9</td>
<td>194</td>
<td>177</td>
<td>194</td>
</tr>
<tr>
<td>8PC-D2[149]</td>
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<td>33</td>
<td>736</td>
<td>505</td>
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Table 6.4: NSW, NSW-RUP, NSW-A and NSW-MOD model values for 316H AR and HAZ materials at 550 °C

<table>
<thead>
<tr>
<th>Type</th>
<th>$P_\varepsilon$ / $P_\sigma$</th>
<th>$A$ (MPa·h$^{-1}$)</th>
<th>$n$</th>
<th>$A_A$ (MPa·h$^{-1}$)</th>
<th>$n_A$</th>
<th>$B_r$ (MPa·h)</th>
<th>$v_r$</th>
<th>$\varepsilon_f$ (%)</th>
<th>$r_c$ (mm)</th>
<th>$I_n$</th>
<th>$I_{nA}$</th>
<th>$\bar{h}_n$</th>
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</thead>
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<tr>
<td>AR</td>
<td>$P_\varepsilon$</td>
<td>1.60×10$^{-35}$</td>
<td>11.9</td>
<td>7.23×10$^{-32}$</td>
<td>10.6</td>
<td>9.67×10$^{-11}$</td>
<td>11.4</td>
<td>21</td>
<td>0.05</td>
<td>4.42</td>
<td>4.30</td>
<td>11.58</td>
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<tr>
<td>AR</td>
<td>$P_\sigma$</td>
<td>1.60×10$^{-35}$</td>
<td>11.9</td>
<td>7.23×10$^{-32}$</td>
<td>10.6</td>
<td>9.67×10$^{-11}$</td>
<td>11.4</td>
<td>21</td>
<td>0.05</td>
<td>3.04</td>
<td>2.94</td>
<td>1.93</td>
</tr>
<tr>
<td>HAZ</td>
<td>$P_\varepsilon$</td>
<td>3.96×10$^{-32}$</td>
<td>10.9</td>
<td>1.00×10$^{-39}$</td>
<td>14.0</td>
<td>5.13×10$^{-22}$</td>
<td>8.0</td>
<td>10.32</td>
<td>0.05</td>
<td>4.42</td>
<td>4.30</td>
<td>11.58</td>
</tr>
<tr>
<td>HAZ</td>
<td>$P_\sigma$</td>
<td>3.96×10$^{-32}$</td>
<td>10.9</td>
<td>1.00×10$^{-39}$</td>
<td>14.0</td>
<td>5.13×10$^{-22}$</td>
<td>8.0</td>
<td>10.32</td>
<td>0.05</td>
<td>3.04</td>
<td>2.94</td>
<td>1.93</td>
</tr>
</tbody>
</table>

Note that the value of $\bar{h}_n$ is based on $n = 12$
6.11 Figures

Figure 6.1: Crack initiation position for welded C(T) specimen (a) crack initiates at HAZ (b) crack initiates at parent material but near HAZ

Figure 6.2: The load line displacement normalised by the specimen width with the time normalised by the test duration for each type of specimen geometry
Figure 6.3: The crack length extension with time normalised by the test duration for each type of specimen geometry

Figure 6.4: Analysis of validity criterion of CCG data with \( C^* \) correlation for each type of specimen geometry
Figure 6.5: Creep crack growth rate correlated by $C^*$ for each type of geometries, showing the C(T) 316H stainless steel HAZ and BM data band

Figure 6.6: Creep crack growth data correlated by $C^*$ for Type 316H stainless steel obtained from [148]
Figure 6.7: Creep crack growth data correlated by $C^*$ for Type 316H stainless steel HAZ obtained from [150]

Figure 6.8: Creep crack growth data correlated by $C^*$ for SENT(T) specimens compared with C(T) BM data band [152]
Figure 6.9: CCG prediction in the HAZ and AR materials using the secondary creep stress exponents and $\varepsilon_j$ -ROA values in (a) NSW (b) NSWA and (c) NSW-MOD models
Figure 6.9 (continued): CCG prediction in the HAZ and AR materials using the secondary creep stress exponents and $\varepsilon_f$-ROA values in (a) NSW (b) NSWA and (c) NSW-MOD models.

Figure 6.10: Optical Observation of creep cracks in SENT-1
Figure 6.11: Initiation time correlated by $C^*(t_i)$ for different type of specimens for (a) 0.2mm (b) 0.5 mm crack extension
Chapter 7

Studies of Re-heat Cracking in an Ex-service AISI 316H Weld Header

7.1 Introduction

Reheat cracking, which has been introduced in Section 2.7, is known as a common issue in 316 steam headers [66-68, 72, 172]. When cracks of significant lengths are detected in the components near weld regions they are generally taken out of service and replaced. Several studies have investigated re-heat cracking in steam headers of type 316H stainless steel which have operated in excess of $10^5$ h at temperatures of around 525-550°C. It is generally considered that the existence of triaxial state of weld-induced residual stresses leads to the accumulation of creep damage [66-69, 71, 72, 74, 75, 172-175].

Reheat-cracking in austenitic materials has been found in five of British Energy’s seven AGR Power Stations [72]. The reheat cracking mechanisms have been systematically studied by combining the effects of stress relaxation, microstructure evolution and service exposure [72, 74, 75, 173, 174]. The three main mechanisms for re-heat cracking have been attributed to (i) the accumulation of the creep cavitation damage at grain boundaries [74]; (ii) a grain boundary decohesion in the presence of the fine intragranular carbide precipitates under creep strain relaxation [176]; (iii) a reduced strength in the presence of a precipitate free zone adjacent to coarse intergranular precipitations [177]. All those factors can influence the creep ductility.

Ductility is the ability of a material to deform under an applied load without fracture and is a function of the triaxial stress-state due to the constraint to inelastic deformation [71]. The stress-state dependency for AISI 316H stainless has been proved from a substantial quantity of experimental data [164, 178, 179]. Generally, reheat cracking is associated with a low level of uni-axial creep ductility, which is itself further reduced by high tri-axial stress field. Small variations of ductility would affect the crack initiation, growth and damage behaviour [71].

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For accurate integrity assessments of welded components with residual stresses, residual stress analysis is a compulsory stage in the design of structural components and in the estimation of their reliability since residual stresses are one of the main effects on welding reheat cracking. Such integrity assessments would be essential for safe operation of plant, life extension, repair and replacement strategy [66]. The investigations performed on residual stresses on welded components have been briefly discussed in Section 2.7.2.

Currently, significant developments on the residual stress measurements have been achieved, however, efforts to develop efficient methods of residual stress measurement are still highly necessary [180]. Compared with the experimental measurements, FE analysis provides more effective procedures to simulate the welding process and predict the residual stress distribution in the components as long as the model is properly validated.

The welding residual stress distribution depends on many factors, including material properties, component structures, restraint conditions, heat input, number of weld pass and welding sequence [181]. The variation of magnitude and distribution of the as-welded residual stresses over periods of high temperature exposure would result in an unpredictable condition for modelling. It was found that the stress that could introduce reheat-cracking should be in tension. Based on this, a simplified method, applying a fixed displacement primary load to simulate effective tensile residual stresses in a component that relax during creep was proposed in this work. As a benchmark of residual stress prediction, analysis was performed in a case which was described previously [66-70] which investigated the measured and predicted residual stresses on a similar header component.

In the current study, rather than fully examine the residual stresses via weld simulations, a simplified axisymmetric numerical model was used by introducing a remote load generated from a fixed displacement on simulated header shaped component, a pseudo tensile residual stress can be modelled using optimised boundary conditions. Under a creeping condition, the displacement controlled internal tensile stresses were relaxed; it will result in the decrease of load capacity in the header component and the creep accumulation in terms of crack initiation and propagation. Effectively bending loads did not affect the residual stress profile due to the asymmetric condition. The stress normal to the observed crack path was examined in an FE analysis and
compared to the available deep hole (DH) drilling measured residual stress reported in Ref [66]. The following concern was taken on the estimation of rupture life for the studied nozzle. If the rupture time is far less than the required service life, it may not be necessary to perform crack growth calculations. The estimation of rupture life can be appropriately estimated according to the fracture mechanics parameter $C^*$ based on the reference stress and the material’s creep rupture data [1]. However, for the reheat cracking due to the relaxation of welding residual stresses, ductility exhaustion methods according to the theory of continuum damage mechanics are more appropriate. In this research, ductility exhaustion was used to investigate the creep damage evolution by using a simplified Cocks and Ashby void growth model which is independent of the creep exponent $n$. It is shown that this simplified model is applicable when an internal residual stress needs to be induced without the presence of extensive plasticity. The creep damage parameter could be estimated from stress-dependent creep strain rate and creep ductility, as shown in Equation (2.83) [57]. The stress which was primarily affected by the loading profile, geometrical constraint, material inhomogeneity and ex-service degradation are briefly discussed below. In addition, the sensitivity study on lower shelf creep ductility was discussed using three different types of model. The damage history for different creep ductilities were compared with the existing 8-15 mm crack length observed in the components after 87,790 h service time. Finally, the predicted creep crack growth behaviour from header/nozzle branch model was correlated using $K$ and $C^*$. The long term CCG simulation on the header/nozzle components was also compared to any available CCG data from C(T) specimens. In addition, experimental techniques including micro-hardness tests and microscopic observations were performed to examine the mechanisms of this type of crack initiation and propagation.

7.2 A Case Study of Residual Stress Measurement and Simulation

7.2.1 Introduction

Limited to the availability of measured data in terms of residual stress distribution in the 1B1/header, the predicted FE results were compared to a similar header component with similar material properties, welding procedures, working conditions and crack positions. An archetypal
example of austenitic header Heysham I /Hartlepool superheated header S4 welds at HYA/HRA was investigated by Smith et al. and Hayhurst [66-70] who studied the stress normal to the crack face by using the available measured data under deep hole drilling and predicted FE data based on welding simulations using thermo-elastic-plastic solution. The main results are reviewed below.

7.2.2 Residual Stress Measurement

Residual stress distributions were measured prior-to-service (as-welded) and ex-service after 55,000 h exposure at 525 °C. The measurements of residual stress were made through the thickness of the components at locations A. The results of pre- and post-measured residual stress in the direction which is normal to the crack face A is shown in Figure 7.1.

In Figure 7.1, the largest measured residual stress occurred in the crack initiation position with a value of 350 MPa and decreased towards the cracking direction. The residual stresses were tensile for depths up to 25mm and turned compressive in the inner surface of the cylinder. In the post-service, the tensile stresses at the outer surface released to 100 MPa. The measured data were employed in the FE model and used to verify any assumed conditions in order to introduce an effective residual stress field into the component.

7.2.3 Residual Stress Simulation

On the basis of measured residual stresses data, residual stress distributions were predicted in an FE analysis based on the thermo-elastic-plastic solution [66-69]. A 2D axi-symmetric model was built on the flank section of the studied header. A stress analysis was performed to simulate stress relaxation up to a time of 55,000h to examine the post-service residual stresses.

The predicted residual stresses in the region of the nozzle-to-cylinder weld were analysed and compared with measured data illustrated in Figure 7.1. The comparison of predicted stress with those measured on an ex-service header by the deep hole drilling technique gave reasonable agreement, confirming that there was tensile stress near the outer surface and compressive stress near the inner surface. The predicted residual stress were significantly relaxed by service aging at an elevated temperature. However, the predicted magnitudes of the tensile stress towards the outer surface were generally higher than the measured values.
7.3 Material Specification

7.3.1 Sample Description

The description of studied steam header component has been performed in Section 4.2.2 - 4.2.4. The header was taken out of service after 87,790 hours, due to the detection of reheat cracks which had been propagated with a length of around 8~15 mm at angles of approximately \(-120^\circ\)~\(-135^\circ\) from the horizontal. Cracks with a measured crack length and direction observed from different cross sections are illustrated in Figure 7.2. The dimensions of reheat crack in Figure 7.2 (a) were used in an FE model. This crack is about 12 mm in length and has grown at angle of approximately \(\theta = -132.3^\circ\) from the horizontal.

Two cross-weld regions can be seen in Figure 4.3, one HAZ/BM boundary stays at the near-crack region, around 6 mm distance from the primary crack tip, another is located remote from the crack. Microstructural observation and micro-hardness analysis were studied on the cracking and cross-weld regions. The specimen was cut from the selected rectangular region (seen in Figure 4.3) with a dimension of 95 mm in length, 6 mm in width and 8 mm in height.

7.3.2 Material Properties

The stress-strain behaviour obtained from experimental examination as described in Chapter 5 was used in the FE analysis. True elastic-plastic properties of the base and cross-weld material obtained from tensile tests were employed and shown in Figure 5.7 (b) and key properties are given in Table 5.1. Uniaxial creep properties of base and cross-weld metal at 550 °C are provided in Table 5.4. The average creep strain rate, \(\dot{\varepsilon}\) was mainly used for stress relaxation and damage analysis. Also note that the stress distributions obtained from the \(\dot{\varepsilon}\) was compared to the secondary creep strain rate, \(\dot{\varepsilon}^*\).

In the creep damage analysis, the ductility exhaustion concept (as mentioned in Section 2.6) was used to predict the initiation and propagation of the reheat cracks. As given in Equation (2.83), creep damage parameter is a function of stress-dependent creep strain rate and creep ductility, the reheat crack initiation could be estimated when the local accumulated creep strains under the
action of the crack-tip stress field were sufficient to exhaust the (multiaxial) creep ductility of the materials.

The multiaxial creep ductility, applied in the current FE model, was based on a modified Cocks and Ashby model [182]. The comparison of the influence of stress state on creep ductility as a function of $n$ is made in Figure 7.3. When components are in service for a long time at low stress conditions, triaxiality ($\sigma_m / \sigma_e$) usually lies in a range of $\sigma_m / \sigma_e \leq 2$, under this circumstance, the normalised ductility is insensitive to $n$ as can be seen in Figure 7.3. For most engineering materials, $n$ lies within 5 to 10, therefore $n - 0.5 / n + 0.5$ in Equation (2.84) would between 0.818 and 0.905 which can be assumed as a constant, approximated as $\sqrt{3} / 2 \approx 0.866$. A simplified equation based on Equation (2.84) is derived in such a way that multiaxial creep ductility is independent of the creep exponent, $n$

$$\frac{\varepsilon^*}{\varepsilon_c} = 0.610 \left/ \sinh \left[ \sqrt{3} \frac{\sigma_m}{\sigma_e} \right] \right. \quad (7.1)$$

### 7.4 Experimental Results

#### 7.4.1 Metallography

In order to perform microstructural observation and micro-hardness analysis, a section (95 mm × 6 mm × 8 mm) was extracted from the nozzle (selected region in Figure 4.3). For ease of sample preparation, this section was subdivided into three parts (seen in Figure 7.4).

Before metallographic examination, all the three parts were ground and polished using the standardised metallographic techniques according to ASTM E3 and E1558 [183, 184]. The materials was then etched using ferric chloride (FeCl$_3$) solution, according to ASTM E340 [185]. Metallographic images of the three parts reflecting the microstructure on select rectangular regions in Figure 7.4 are shown in Figure 7.5. The whole cracked region is shown in Figure 7.5 (a); the weld/base material boundary close to the crack can be seen in Figure 7.5 (b) and Figure 7.5 (c) shows the microstructure of the weld/base material boundary remote from the crack.
The microstructure and grain boundary cavitation associated with reheat cracking in a similar header was previously examined in Ref [186]. As also observed in Figure 7.5 (a), crack propagation is clearly the result of the growth and coalescence of creep-induced cavities at grain boundaries to form micro-cracks and eventually macro-cracks [75, 186]. The subsidiary intragranular cracking within the adjacent regions of the main crack can also be observed. Due to the limited resolution of the optical microscope, the dimension of intragranular creep cavities could not be precisely measured from Figure 7.5 (a). Also, the small precipitates along the grain boundaries could not be clearly inspected.

In Figure 7.5 (b) and (c), a typical cross-weld section was photographed under optical microscope to investigate the existence of micro-cracks. Clearly, No micro-cracks were observed along the fusion lines near or away from the cracks. It means that the cracks would not initiate from boundary between BM and HAZ. Usually, the grains, particularly in the HAZ, are prone to become coarsened and the stress is more likely to be concentrated so that the mechanisms of reheat cracks are magnified. However, in the case of the unaffected zone, even though the metal structure is of the same metallurgical system, the stress is likely to be concentrated due to the geometrical constraints. In this case, the geometrical constraint effect dominates the crack initiation and growth and its direction.

According to features as shown in Figure 7.5 (c), the microstructure of the studied materials has been briefly discussed. The structure consists of standard austenitic profile formed in base metal (region A) and fine dendritic-structure in weld metal (region B). The longitudinal thermal gradient during welding has resulted in a gradient of transformed structure as a function of distance. The transition zone near the fusion line is darker than other areas (region C), and the stream-like arrays appear to emanate from prior austenite grain and extend to a distance by a rapid directional growth [187].

### 7.4.2 Vickers Hardness Measurement

A hardness survey was conducted over the surface of the welded region of the branch. In order to thoroughly investigate the hardness variation in the vicinity of the crack and fusion lines the hardness measurement was performed at 1 mm intervals. Further away from interested regions, where no steep gradients were observed measurements were obtained every 2 mm or 4 mm
intervals. For comparisons, hardness profiles were also performed at 1-2 mm intervals on a sample obtained from a newly welded sample (i.e. sample that has both been service exposed). This new weld sample (as shown in Figure 4.18) was manufactured and extracted as detailed in [188].

A contour plot of the hardness profile of the ex-service sample is show in Figure 7.6 (a), where the position of the crack and fusion line was identified. Peak hardness of 250 HV were observed in the vicinity of the fusion boundary. The weld was generally harder than the base metal, and the average hardness values for weld metal (WM) and base metal (BM) were approximately 210 HV and 170 HV, respectively.

For the new weld, Figure 7.6 (b), significantly higher hardness was observed in the WM compared to the BM. In addition, the hardness of the new weld was generally higher than the ex-service material. This indicated that both the BM and WM have softened during the long service period.

The hardness distribution along the dashed line shown in Figure 7.6 (a) is plotted in Figure 7.7. A relatively symmetric profile is seen about the weld centre. The hardness increased asymptotically between BM and WM, ranging from around 160HV to 220HV, peaking to 230 HV at the position of the heat affected zone (HAZ). The hardness of the HAZ was higher than the average value of WM which might due to the fact that heating and sequential cooling down during welding, which result in the aggregation and formation of dense grains [189]. The observed crack initiation area was in the sharp increase zone within the BM and near the HAZ regions. The crack was propagated perpendicular to the region of high hardness gradient towards the lower hardness region, even though this tended to be in the base material.

7.5 Finite Element Model

7.5.1 Stress and Creep Damage Analysis in the FE Model

A two-dimensional axisymmetric model was used to study the regions of interest in the header’s branch to the nozzle, as shown in Figure 7.8. The FE analysis was performed on ABAQUS v.12. The HAZ was not explicitly modelled, thus a bi-material system was considered. The continuum field was mainly split into four noded bi-linear axisymmetric quadrilateral with reduced
integration elements (type CAX4R) with few elements with three noded linear axisymmetric type (CAX3). The mesh was refined to 100 µm ×100 µm in the vicinity of the cracking area as shown in Figure 7.9 by creating transition partitions. The element sizes at the cross-weld region were also designed finer than other areas. The FE mesh used to model the flank section consisted of 12,248 elements, which can be seen in Figure 7.8.

The analysis was performed in two steps. Having obtained the elastic-plastic solution by applying a load, the component was allowed to creep at 550°C in the second step. The analysis time was set to 250,000 hours. Stress analysis was firstly performed at $t = 0$ and $t = 55,000$ h in order to compare the pre- and post-service stress distributions from previous data obtained from deep hole measurement and the FE prediction reported in Ref [66, 70].

The following work was carried out on creep damage analysis. Cracks in the length range of 8-15mm were found in the real components, based on this findings, the predicted crack length vs. time in the current model was plotted under different creep ductilities and compared with real situations.

Since the model highlighted an important issue based on the time taken to initiate a crack, numerically three surface conditions were therefore investigated. The first was to assume that the corner of the nozzle from where the crack grew was smooth; the second condition was to assume a small defect at the corner edge with the size of 0.2 mm or 0.5 mm in length at an angle of -132° from the horizontal, by removing 2 or 5 elements respectively from the first model (Figure 7.10). The third condition was to introduce at the surface a small region with lower uniaxial creep ductility value in order to facilitate the crack initiation (seen the blue region in Figure 7.11). This was logical as that region was very near the weld toe and exhibited higher micro hardness and lower strain. For the header, the damage analysis was specifically carried out to investigate the cracking behaviours and predict failure based on the stoppage time of 87,790h. Also for the damage analysis, mesh sensitivity studies were carried out to identify the optimum mesh size. The damage history obtained from a fine transition mesh model (Figure 7.9) were compared from a coarse mesh model which comprised of meshes up to 800 µm ×800 µm in the corner, as shown in Figure 7.12. It is clear that the fine mesh adopted converges satisfactorily.
In order to define the creep ductility for different regions, describe the stress normal to the crack face distribution and perform creep damage analysis using modified Cocks and Ashby model as shown in Equation (7.1), a user-defined subroutine was employed in the ABAQUS analysis to track the user-defined output variables. The subroutine allowed the record of the current value of accumulated creep damage (ω) base on Equation (2.83) as well as the store of the stress normal to the crack face, \( \sigma_\perp \). The calculation of \( \sigma_\perp \) in a 2D model is according to stress coordinate transformation equation as follow

\[
\sigma_\perp = \frac{\sigma_{11} + \sigma_{22}}{2} - \frac{\sigma_{11} - \sigma_{22}}{2} \cos 2\theta - \sigma_{12} \sin 2\theta
\]  

(7.2)

where \( \sigma_{11}, \sigma_{22} \) are the principle stress in the x and y direction, \( \sigma_{12} \) is shear stress, \( \theta = -132.3^\circ \).

### 7.5.2 Fracture Mechanics Parameters

The stress intensity factor, \( K \), can be used to characterise the crack growth behaviour when the creep zone is smaller than the elastic field ahead of the crack tip. In the FE model, 2D linear elastic analysis was performed to determine elastic stress intensity factor with crack length ranging from 5 and 30 mm. The elements at the crack tip were designed with collapsed type and expanded to the type of sweep away from crack-tip region. The numbers of elements were about 25,000. Figure 7.13 illustrates the localised mesh design considering the initial crack is 10mm at an angle of \( \theta = -132.5^\circ \) from the horizontal, the element at the crack tip is magnified. After positioning crack tip and its direction, \( K \) can be obtained from linear elastic FE analysis.

\( C^* \) parameter has proven to be a good prediction for the creep crack growth behaviour of 316H stainless steels. Current measurement of \( C^* \) was based on the reference stress method as described in Equation (2.49). The current study allowed a displacement controlled tension load to be applied in the model which had homogeneous and elastic-perfect plastic material properties. As load increased, the elements at the crack tip were firstly collapsed, following that, sufficient large distortion occurred. The nodal force on the corresponding elements would increase firstly and remain constant. The constant force was considered as limit load. After reaching the limit load, FE model would collapse. The reference stress could be obtained according to Equation (2.10) and it related to \( C^* \).
7.6 Stress Analysis

The case study in Section 7.2 has investigated the residual stress distribution from the measured data and the FE prediction from a similar header [66-69]. The objective for this work was to develop a residual stress profile by using loading and boundary conditions that would best represent the actual measured residual stress data that was available for the similar header. Therefore a sensitivity analysis is presented to identify the optimum loading profile and boundary conditions.

7.6.1 Loading conditions (LC)

As shown in Figure 4.3 and Figure 7.2, reheating cracks grew at an angle from the outer to inner surface of the header due to the existence of welding residual stresses. If the stress was simplified using a fixed displacement remote loading procedure, the load acting on nozzle could only be in the form of fixed displacement tension or bending or the combination of these acting on the region, as illustrated in Figure 7.8. Possible loading conditions have been examined, as described in Table 7.1.

7.6.1.1 Bending Load

Firstly, a substantial bending load of 1000 kN was placed on the nozzle with two different boundary conditions as described in Table 7.1. Stress normal to the cracking face of LC-B1 and LC-B2 was compared with measured data, as shown in Figure 7.14.

Despite the high bending stress, the maximum resultant stresses normal to the crack in LC-B1 and LC-B2 were less than 50 MPa, which were far less than the yield stress of the material (170 MPa for base metal). The bending load toward the axis of symmetry would be offset due to the asymmetric condition; therefore, the bending load was found not to be representative of the residual stress profile needed for crack propagation.

7.6.1.2 Load-controlled tension and Displacement-controlled tension

Any residual stress which results in crack initiation and propagation is tensile in nature. The ABAQUS analysis allowed both load-controlled (LCT) and displacement-controlled (DCT) tension. The load controlled method was used to introduce a direct load on the component whilst
the displacement controlled test was used to apply a displacement on a body using different boundary conditions (BCs). In the LCT, the tension surface was constrained to a reference node which could avoid any stress concentration on the loading position. Two methods were employed in the current non-linear analysis, set as LC-LCT1 and LC-DCT1. The hoop, radial, axial stress and in-plane principle stress contours were compared as shown in Figure 7.15.

The stress in all directions had the same distributions when compared with displacement of 0.25mm and load of 1300 KN. Despite the same distribution in the loading procedure, displacement controlled analysis was preferable for the following long term stress relaxation condition. During the stress relaxation, the studied header component would begin to demonstrate strain-softening behaviour and a gradual decrease in load carrying capacity with increasing strain. The release of the strain energy could result in a formation of cracks which may occur at the location that has the highest stress and the weakest bond [190]. Under the fixed displacement, the crack would continue to develop to release more storage energy. However, in the fixed load, the crack may result in a sudden failure due to the input of external energy.

**7.6.1.3 Magnitude of Displacement-Controlled Load**

In order to obtain a comparable level of stress distribution to the measured data, a sensitivity study was made for the stresses acting on the outer surface of nozzle with different vertical displacements ranging from 0.25mm to 2.0 mm, as illustrated in Table 7.1. The stress normal to the crack face was considered, because cavity growth is controlled by vacancy aggregation which is affected by the component of stress normal to the grain boundaries [191]. Figure 7.16 shows the predicted data with different vertical displacements compared with measured data on stress normal to the crack face along the crack directions.

All predicted stress distribution shown in Figure 7.16 shows a tensile residual stress within the cracking area near the outer surface of the header, and a compressive stress in the inner surface of the header. Among them, the model LC-DCT2 with a vertical displacement of 0.5 mm had a peak tensile value of 292.3 MPa, which was close to the measured data of 300 MPa. Also, the stress distribution was at a comparable level to the measured data. Therefore, LC-DCT2 provided a good representation of the residual stress profile in terms of stress normal to the cracking direction in the header.
7.6.2 Boundary Conditions

The inner side of nozzle was already fixed due to the symmetry condition. However, suitable boundary conditions need to be applied to the two surfaces denoted surf 1 and surf 2 in Figure 7.8. A study was performed to determine the most appropriate boundary conditions acting on these surfaces.

At least one surface should be fixed in y direction (U2) to avoid unlimited displacement. Generally, due to model’s high constraint and axisymmetric conditions, the restriction of rotation about the 3-direction (UR3) did not affect the stress distribution, which was validated by comparing the stress distribution with or without fixed UR3. In addition, a fixed x-direction boundary condition (U1) on surf 2 would not significantly influence the stress distribution due to the general high constraint in the x-direction. Also found that the boundary conditions restricted in both surf 1 and surf 2 was redundant, it gave similar stress distribution when compared to fix U2 on surf 2. In conclusion, only three cases, denoted BC-1 to BC-3, require consideration, as described in Table 7.2.

For the three cases considered, the stress normal to the crack face along to the crack path was compared with measured data (shown in Figure 7.1). The stress distribution normal to the crack face versus distance from outer surface is shown in Figure 7.17 (a) pre-service and (b) post-service (55,000 hrs. of creep). In this analysis the average creep strain rate was employed. Reasonable predictions were obtained of the initial stress distribution in Figure 7.17 (a), with BC-1 generally providing the best estimate. Thus, the following discussion is focused on the mechanical behaviour of BC-1. The results of the stress after creep based on the average creep strain rate in the FE analysis were shown an overestimation of the residual stress compared with deep hole measured data. The data in Ref [66] also tended to over-estimate the welding residual stress and had a greater deviation compared with the current FE analysis.

The measured data illustrated a 75% relaxation; however, the stress relaxation rate in BC-1 was about 36%. This discrepancy might be due to the fact that the creep strain rate of the component in this region might be higher than the BM values employed, due to micro-structural changes in this region and any plastic strain that might be present in the component post fabrication [192].

Contour plots of the stress distribution predicted prior to and after service exposure are shown in Figure 7.18-Figure 7.20, for BC-1 to BC-3, respectively. The stress distribution shown in Figure 7.18 shows a tensile residual stress near the outer surface of the header and an inner compressive stress. At the beginning of the creep shown in Figure 7.18 (a), the stress had a peak tensile value of 292.3 MPa and peak compressive stress of 185.6 MPa, with the peak tensile value very close to the crack initiation point. After 55,000 hours of creep, seen in Figure 7.18 (b), the peak tensile stress reduced to 184.6 MPa, which was around 36 % decrease. Similarly a 37% reduction in magnitude of the peak compressive stress was predicted. These decreases mean that stress had been relaxed partly. Note that peak tensile region appeared to follow the expected crack path with stress relaxation. Therefore, it could be predicted that the crack would initiate in the maximum tensile stress region and subsequent stress redistribution would result in crack propagation.

Generally, the stress distribution in Figure 7.19 is similar to those in Figure 7.18 where the maximum stress was also in the cracking area with a value of 291.8 MPa. However, due to an additional boundary condition in BC-2, a large tensile stress appeared at the root of the fixed surface. This stress distribution was similar to that seen in BC-1, and therefore affects the stress re-distribution along the crack path.

In BC-3, the stress distribution shown in Figure 7.20 cannot fully predict the cracks observed in Figure 4.3. Because the maximum stress acting on the tensile surface would dissipate quickly in the vertical direction. The maximum stress around the crack initiation point was 187 MPa, which was insufficient to cause crack initiation.

### 7.6.3 Creep Strain Rate

To examine the influence of the creep strain rate, results based on the secondary creep strain rate (see Table 5.4) were also employed for BC-1 and the results were compared to stress distribution using the average creep strain rate, as illustrated in Figure 7.21. The maximum stress after a secondary creep was 220 MPa compared with 185 MPa using the average creep rate, an extra 15% stress relaxation might result from the rupture stage.
7.6.4 Equivalent Creep Strain Distribution

The post-serviced equivalent creep strain distribution in case BC-1 was examined at 55,000 h and the results can be seen in Figure 7.22. For the average creep properties considered, a maximum of 0.3% creep strain deformation could be observed around the crack tip area. However, it is indicated that a higher creep strain rate properties are required to reduce the residual stresses to the magnitude measured, meaning that higher creep strains are accumulated in the component during service. The lower bound uniaxial creep ductility for base material is around 0.9% [134], and triaxiality effects will reduce this ductility further. The model presented here is considered capable of predicting re-heat cracking; therefore, creep damage analysis is discussed in the following section.

7.7 Fracture Mechanics Analysis

Fracture mechanics analysis of the cracked component was firstly carried out. This method assumes the presence of a crack of finite size in a component and then evaluates its propagation due to creep to determine the remaining life of the components [193]. Analysis of the model with different creep ductilities and crack lengths would result in a series of values for $K$ and $C^*$. It is suggested in ASTM E 1457 [51] that $K$ is applicable for creep-brittle materials while $C^*$ is suitable for characterizing CCG behaviour in creep-ductile materials. Those two fracture mechanics parameters should be matched against corresponding values of crack growth rate, $\dot{a}$, with a consideration on the validity checks. According to the previous analysis, the creep crack length would show a relationship of the form:

\[
\dot{a} = \frac{k}{D_k \phi_k \phi} \quad (7.3)
\]

\[
\dot{a} = \frac{k}{D_k \phi_k \phi} \quad (7.4)
\]

The constant $D_k, D_k, \phi_k, \phi$ are used to fit the regression lines through the valid data.

In the widespread creep situations, where the creep crack growth data are not available, approximate NSW model as seen in Equations (2.64) can be used to estimate the creep crack growth rate. The use of NSWA model generally gives conservative predictions of crack
propagation rate [1]. Therefore, it is possible to calculate the approximate creep crack growth rate if \( C^* \) and \( \varepsilon_f^* \) are known.

\( \varepsilon_f^* \) can be obtained according to Equation (7.3), where \( \varepsilon_f = 0.9\% \) and the maximum triaxility can be extracted from FE result. Since the current analysis did not consider the triaxility, in such case, \( \varepsilon_f^* \approx 0.9\% \).

For steady state creep, the evaluation of \( C^* \) might be obtained by finite element analysis for materials with creep rates defined by a secondary creep law with a simple power dependence on stress. In practice, however, it is more recommended that a reference stress based estimate of \( C^* \) be used. The basic parameters required for using the approximate methods are the elastic stress intensity factor, \( K \) and the reference stress, \( \sigma_{\text{ref}} \), as shown in Equation (2.49).

From the linear elastic model described in Section 7.5.2, the results of stress intensity factor vs crack length under a displacement controlled tension analysis is plotted in Figure 7.24. As crack length increased, \( K \) would decrease with an increasing rate. In an elastic-linear plastic model, the FE allows the obtaining the value of \( P_{\text{lc}}/P \) from which \( \sigma_{\text{ref}} \) was obtained according to Equation (2.10), the relationship between crack length and \( \sigma_{\text{ref}} \) is plotted in Figure 7.25 in which a decreasing trend can be observed in the displacement controlled analysis. According to the simulated \( K \) and \( \sigma_{\text{ref}} \), \( C_{\text{ref}}^* \) is therefore obtained in different crack length, which is shown in Figure 7.26, predicted under either plane strain or plane stress condition. Due to the limited sources on CCG correlations in the header components, the identification and comparison have been performed by investigating a valid correlation using CCG data band in the base metal of C(T) specimen from either short or long term tests. As can be seen in Figure 7.26, the prediction in NSWA provided larger scatter band between an upper bound controlled by plane strain condition and lower bound controlled by plane stress condition, as the difference between plane stress and plane strain CCG for the creep ductile materials was obtained to be a factor of 30. The NSWA plane stress line provides a good lower bound prediction of the experimental trend for the CCG data in C(T) specimen. An over-prediction is made on the plane strain line.
Since the CCG rate in C(T) were generally fall between plane stress and plane strain prediction for the header/nozzle component, therefore it is possible to predict the CCG rate on the real component based on the laboratory testing data on C(T) specimen since CCG data correlated by $C_{ref}^*$ for different geometries and sizes are similar to the data band of 316H C(T) specimen.

### 7.8 Creep Continuum Damage Analysis

The fracture mechanics analysis has been performed to approximately correlate the crack growth history, however, when considering the stress relaxation during damage evolutions, the creep continuum damage method becomes more appropriate in predicting the creep crack growth rate in the components. As mentioned before, the creep damage parameter is a function of creep strain rate and creep ductility, both of which may be stress dependent. The stress-dependent creep strain rate could be affected by loading type, ex-service degradation, geometrical constraint and material inhomogeneity. In the current prediction, the first two variables have been considered. A sensitivity study on the effect of weld metal’s properties to the creep damage was carried out, together with a consideration of geometrical constraint.

Following that, sensitivity study to creep ductility was performed to predict the creep damage history based on the modified Cocks and Ashby model. Note that from the long/short term data, the failure ductility for 316H at 550 °C could be simplified to an approximate lower/upper shelf [134, 166], with a transition region in between, the lower shelf is approximately 0.9%, the results are different with previous research from Hayhurst [62] which suggested that failure occurs due to the pre-strain region near the residual stresses, such that creep ductility is higher under a long term and lower load condition. For predicting failure in the nozzle it is therefore more appropriate to use the lower shelf ductility. Different ductility values limited to the lower shelf were selected and discussed in three types of models.

#### 7.8.1 Material Properties

As shown in Figure 4.2, the existence of weldment allows two materials properties to be used in the FE analysis. The use of weldments usually requires an over-matched condition where the yield strength of weld metal is higher than base metal. To perform a sensitivity study on the material properties, the FE analyses were taken into account three conditions which were briefly
called as even-match, 10% over-match and 20% over-match, based on the assumption that stress of weld metal for a given strain is 0%, 10% and 20% higher than the base metal, meanwhile, the material constant $A$ in Norton’s law as shown in Equation (2.14) is 0%, 10% and 20% lower than that of base metal. The uniaxial creep ductility was kept constant at 0.9% for all three samples.

The mismatch effect on the crack initiation was firstly investigated by the FE method. Under different mismatch values, the crack initiated at identical time, 63,029h with an identical contour plot as shown in Figure 7.23.

After 180,000h, the creep damage contours for different mismatch ratios were compared as shown in Figure 7.27. In all three cases, the cracks initiated at the corner of studied geometry and grew from the outer to inner surface of the header. The growth direction would deviate from the BM/WM boundary. Under even-match, 10% over-match and 20% over-match conditions, the crack was about 13.5, 14.2 and 14.4 mm in length, and grew in a direction of -118.5, -123.2 and -126.1° respectively. Although the values were very close, it should be noted that the higher mismatch might result in a slightly long crack and lead the crack slightly close to the boundary line.

In conclusion, material inhomogeneity would not influence the crack initiation behaviour, and may not significantly affect the creep damage behaviour. The creep crack growth path would mainly depend on the geometrical constraint and occurs in a location where the stress was concentrated and did not follow the HAZ weak creep region. In real welded steam header components, the crack usually initiate at the branch between the header and nozzle due to concentration and combinations of localised welding residual stresses but the creep damage behaviour may not necessarily be influenced by the weldment.

**7.8.2 Ductility Analysis**

**7.8.2.1 Model 1: Smooth corner**

Firstly, a study looking at the influence of the mesh design on the cracking behaviour has been performed. The creep crack behaviours have been compared with two models with fine (Figure 7.9) and coarse (Figure 7.12) meshes on three ductilities ($\varepsilon_f = 0.5\%, 0.8\%$ and $0.9\%$). The
results are shown in Figure 7.28. Due to a smooth surface, the FE analysis took a significant period of time to develop a crack and a crack initiation period seemed to occur before crack growth. This period might not be realistic under reheat cracking, which tended to initiate and grow in the early stages but it did indicate the importance of surface conditions in long-term initiation of the crack. Generally, the coarse mesh would result in a larger crack initiation period; as ductility increases, the difference in crack initiation time would extend. However, after crack initiation, there were similar patterns in creep crack growth after the onset of cracking as the crack growth rate and stable crack length remains similar. Due to the similar crack propagation behaviours, the following consideration is taken into account in the fine mesh model.

Creep ductility sensitivity analysis in the fine mesh model was performed. Seven different ductility values including 0.5%, 0.8%, 0.85%, 0.9%, 0.95%, 1.0% and 1.2%, denoted as DA-1 to DA-7, were applied in this analysis. To highlight the effect of creep ductility on the time-scale of creep crack growth in the header, the variation in predicted creep crack length with time is presented in Figure 7.29 for the six cases studied. Note that the crack initiation time of DA-7 (\( \varepsilon_f = 1.2\% \)) exceeded the predictive creeping time (250,000 h). When the creep ductility was over 1.5%, crack growth did not occur after a long creeping time (10^7 h) under the present induced fixed displacement stresses. Effectively the creep damage parameter \( \omega \) would not reach 1 with higher creep ductility according to Equation (2.83).

In this model, creep ductility significantly affected crack initiation times in a way that the increase of ductility would extend the crack initiation time. During the crack propagation stage, cracks would grow in high but slightly reduced rate to a specific crack length, after which point the crack growth rate was significantly reduced and finally reached a plateau. This was consistent with the relaxation of the fixed displacement stresses that were applied to reflect an applied residual stress condition. Also shown in Figure 7.29, the estimated stable crack length would decrease with increasing creep ductilities in a clear trend.

In Figure 7.29, the results under different creep ductilities were compared with the upper/lower bounds of measured cracks in the header. In the case of DA-4 where \( \varepsilon_f = 0.9\% \), crack length was predicted grown to 13mm in the service time. However, the results were based on the assumption that the initiation time could account for a relatively large proportion of the service
time, which is not realistic. In the real situation, reheat cracking develops quickly as a consequence of the stress relief at elevated temperatures at the early stage of the service time rather than the late stage [194]. During service in the increasing temperature, the yield strength or creep strength of the material would decrease and when the welding residual stress exceeds these at the high temperature, the creep deformation would develop. If there were a sufficient ability to deform the creep deformation, the residual stress would relax. But if there were a sufficient ability to deform the plastic deformation under low ductility and high constraint condition as in the case of the header, residual stress would relax through intragranular cracking. In such a case, the damage would occur at the grain boundaries, leading to the reheat cracking.

Ignoring the crack initiation period, the creep crack behaviours under different creep ductilities are shown in Figure 7.30 and also compared with the upper/lower bounds of measured crack in the header. The crack lengths in all cases over predicted the actual crack length range, meaning that the method is conservative under low creep ductility. By extrapolating the peak values in the predicted crack lengths in Figure 7.30, a predicted ductility can be derived for the crack growth times measured in the header. The ductility is plotted against the stable crack length which is shown in Figure 7.31. Assuming the header crack grew in the same rate to 8-15 mm during the service time, after 87,790h it could be assumed that this is achieved at a higher ductility. Figure 7.31 can then be used to predict the ductility. The upper bound ductility estimated gives a value of 1.9%. Since there were numerical difficulties in achieving a successful FEM runs with 1.9% ductility, this estimation seems a good approximation. However, this requires further verification which was not possible due to the limited time.

The crack direction was also examined in the fine mesh model to provide comprehensive information on the cracking behaviour. Figure 7.32 shows the crack direction deviated from horizontal with different creep ductilities. By increasing the creep ductility, the absolute value of angle slightly decreases. The angle varies in a narrow range between 115° and 125° as the ductility is in the range of 0.50 ≤ εf ≤ 1.0. As the predicted angle is close to the real condition (−120° ~ 120°), the prediction using the simplified model can well describe the real cracking direction.
Finally, the damage contours in DA-4 over a selected cross-section for four different stages are briefly described. Figure 7.33 (a) shows that damage initiation takes place at 63,929 hrs in a distance of 4.8 mm from the outer surface. The stress distribution described in Figure 7.17 shows that the peak tensile stress occurs also in a close distance from the outer surface. Therefore, the maximum peak tensile stress normal to the crack face might result in the crack initiation. Figure 7.33 (b) shows the damage field at 73,258 hrs, in which the damage zone has grown to the outer surface. During stage (a) and (b) the damage area grew slowly due to the crack blunting effect, resulting in an unrealistic initiation time. As the crack initiated at the outer surface, the creep crack growth rate increased where the creep crack was in a more confined zone with a sharp crack and there was no significant blunting effect. Service time at 87,790 hrs is shown in Figure 7.33 (c) as crack length has grown to 13mm at an angle of -116° from the horizontal. Figure 7.33 (d) shows the damage field at 136,000 hrs where the creep crack has grown through one half of cross section.

7.8.2.2 Model 2: Existence of Small Defect

Since the earlier model using a smooth surface predicted unrealistically long initiation times, a small defect of 0.2 mm was introduced in the fine mesh mode at an angle of -132° from the horizontal, whilst a defect of 0.8mm at an angle of -132° was introduced in the coarse mesh model. As shown in Figure 7.34, crack evolution has been examined and compared with the results in three different ductilities ($\varepsilon_f = 0.5\%, 1.0\%$ and $2.0\%$) using the two models. Firstly the initiation times were virtually eradicated. The cracks in the coarse mesh model tend to start slightly earlier than that in the fine mesh model, since longer defects may facilitate the crack initiation. Once the crack was initiated, the cracks in different models grew at a similar rate, except a slightly slower rate in the coarse mesh model for $\varepsilon_f = 2.0\%$. The stable crack length in the coarse mesh was about 5% less than that in the fine mesh. The following discussion is focused on the fine mesh model due to the similar crack growth behaviours.

In the fine mesh model, two different sizes of defects with 0.2 or 0.5 mm in length were considered. Cracks behaviours were examined, without consideration of the long-time crack initiation periods, using six creep ductilities: $0.5\%, 0.9\%, 1.0\%, 1.2\%, 1.5\%$ and $2.0\%$, denoted as DA2-1 to DA2-6 separately.
Firstly, Figure 7.35 and Figure 7.36 show the comparisons of the variations in predicted creep crack length with time in model 1 and model 2 considering two initial defect sizes, 0.2mm and 0.5 mm, using two creep ductility, i.e. $\varepsilon_f = 0.5\%$ and $\varepsilon_f = 0.9\%$, respectively. As can be seen in these figures, the crack propagation in model 2 occurred significantly earlier than that in model 1, without an obvious long crack initiation period, due to the existence of the initial defect. The crack propagation in model 2 generally consisted of a short stage where the crack growth rate was increasing with time, after that, the crack grew in an almost constant rate and the rate was similar to that in model 1. This reflects the fact that the existence of initial defect would accelerate the crack initiation period but might not affect the crack propagation behaviour once the crack initiated.

A comparison was also made on the two defects sizes in model 2, showing in Figure 7.35 and Figure 7.36. An increased defect might slightly decrease the crack initiation time. However, the cracking behaviour under a constant fixed displacement was significantly dependent on the existence of initial crack rather than the size of defects, as the crack propagation behaviour was close under different defect sizes. Therefore, the following discussion only considers the defect size of 0.2mm.

Figure 7.37 shows the predicted creep crack length with creeping time for different creep ductilities which are compared to the actual conditions. The comparison made between Figure 7.30 and Figure 7.37 proved the previous finding that the crack propagation stage was similar due to the fact that the existence of defects may be insensitive to crack propagation. It was also found that the CCG rate could decrease with increasing creep ductility, and the stable crack length could also decrease. Compared with the results in model 1, cracks can still be initiated and propagated with higher ductility larger than 1.2%, however, in the numerical analysis, when the creep ductility is larger than 2.0%, the calculation would terminate due to the existence of the FE instability in the ABAQUS package, this is mainly due to the very small time increments needed in ABAQUS for iterating each step.

Finally, the contours of creep damage on DA2-2 with $\varepsilon_f = 0.9\%$ at different stages are shown in Figure 7.38. Figure 7.38 (a) shows the initiation of crack, which occurs at 4,814 hours at the root of defect due to the existence of maximum tensile stress at this point. Figure 7.38 (b) and Figure
7.38 (c) show the damage field when the crack reaches 8 mm and 15 mm, at 8,242 hours and 11,219 hours respectively. The time to cause 8-15mm cracks were between 3500 and 6500 hours, which was close and slightly shorter than the results in model 1 (4000-8000 hours), also provided an overconservative prediction. The main crack direction under a wide range of creep ductility followed the angle of defects, as expected. The crack direction was different with that in model 1 (Figure 7.32), proving that the crack direction would depend on the stress state in the crack tip vicinity.

7.8.2.3 Model 3: Existence of surface region with lower creep ductility

In order to eliminate the period of crack initiation, a small region was partitioned in the corner of nozzle to header as shown in Figure 7.11, the creep ductility in this region was set to a lower value, fixed at $\varepsilon_f = 0.5\%$ and the ductility in other regions was varied in the analyses, using the values of 0.5%, 0.9%, 1.0%, 1.5% and 2.0%, named as DA3-1 to DA3-5 separately. In Figure 7.39, the creep crack history has been compared in model 1 and model 3 for three different creep ductilities, i.e. $\varepsilon_f = 0.5\%, 0.9\%$ and 1.0%. The results of $\varepsilon_f = 0.5\%$ between the two models were same, due to the homogeneous ductility conditions. As for $\varepsilon_f = 0.9\%$ and 1.0%, the existence of regions with lower creep ductility would result in a significant reduction of crack initiation stage. The initiation time for $\varepsilon_f = 0.9\%$ is 23,000h, which is significantly lower than 76,000h in model 1 but higher than 2,000h in model 2. It means that in model 3, localised creep ductility reduction may intrigue a crack initiation, since this reduction in ductility could reflect the reduced strain due to the weld HAZ region which implies in Figure 5.6, it may easily occur at the surface. In fact this corresponds with the increased hardness measured on the header which is shown in Figure 7.6. Once the crack initiated, the crack propagation in model 3 remained similar crack growth rate and stable crack length.

Ignoring the crack initiation stage, the predicted crack length under different creep ductilities is shown in Figure 7.40. As ductility increases, both crack growth rate and stable crack length reduced, as expected. Unlike the case in model 1, the crack could be initiated when the ductility was as much as 2.0%. However, a crack did not grow for higher ductilities. Therefore, in order to estimate the stable crack length that predicts the crack length size in the header, an estimation
was made using a linear fitting line, as for the other modes, to predict the higher values of creep ductility based on the low ductility data, obtained from all the three models. The results are shown in Figure 7.41. The stable crack lengths in models 2 and 3 were generally satisfied with the results in model 1, in the lower ductility range. However, when considering the ductility over 1.0%, the estimated ductility would increase to 3.0% - 3.8% for the upper/lower bound of measured crack length when the crack length is between 8-15 mm. Compared with the results in Figure 7.31, this prediction might be more realistic.

7.8.2.4 Summary of Results

The ductility analysis has been considered in three different models. Model 1 developed a large and unrealistic initiation stage due to a smooth surface and the initiation time seemed dependent on the mesh design and creep ductility. However, the reheat crack initiation in real components will occur at earlier stages of service. In order to minimise the crack initiation stage, models 2 and 3 have been designed and used to analyse the cracking behaviour, the first using small pre-crack and the second hardening the surface where initiation occurs. These models eradicated or reduced the initiation times. In order to predict a realistic creep crack growth rate to correspond to the header crack size of 8-15 mm a sensitivity analysis using different ductilities were carried out. It was found that the lower values of creep ductilities of 1-2% could provide conservative service times. The higher value of ductilities could reduce the creep crack rate substantially or make the FE analysis unstable and prone to failure. It was also shown that depending on the model cracks may occur when the ductility was less than 1.2% in model 1 and less than 2.0% in models 2 and 3, but may not occur at higher ductilities since the creep damage parameter would not reach 1 for the present applied stresses. By extrapolating crack depths during stable cracking using the low shelf creep ductility data, it was possible to suggest that a ductility of 3.0-3.8% may be needed to grow the 8-15 mm cracks after 87,790 h for the header components.

7.9 CCG Correlations

The crack growth rate obtained from continuum damage analysis can be correlated with fracture mechanics parameters to compare with the CCG assessment obtained from fracture mechanics analysis performed in Section 7.7. Since the crack propagation in different models are similar,
the following considerations are only taken in model 1. As seen in Figure 7.29 crack growth with creeping time has been predicted using six values (excluding DA-7) from which CCG rate can be obtained by differentiating the plot of crack length vs. creeping time. The CCG rate was correlated with two fracture mechanics parameter in the present research, i.e. $K$ and $C^*$. Applicability of the creep strain rate in long term crack growth under high temperature at relatively low stress conditions required studies to identify the relevant correlation parameters. The correlations using $K$ and $C^*$ have been described below, also compared with sufficient CCG data performed in 316H base metal and HAZ of C(T) specimens.

7.9.1 Correlation using Stress Intensity ($K$)

$\frac{da}{dt}$ vs. $K$ is predicted with a consideration of different ductilities, the results are plotted in Figure 7.42 where $K$ may be appropriate to correlate CCG in the long term cracking in the header as the ductility at low stress long time is within the lower shelf. The studied ductilities can be appropriately estimated by an upper and lower bound as described in Figure 7.42 which may be used to predict the CCG rate.

In Figure 7.43, a direct comparison has been made between the CCG behaviour described by a lower and upper bound on the current study of nozzle/header component and the data band on the C(T) 316H base metal [134] and the C(T) HAZ [150]. Generally, CCG rate in the present study lies in the lower bound of C(T) base metal data band due to the long-term low stress condition. The data band on C(T) specimens have a large scatter, this means that creep crack growth behaviour on 316H steels may not suitably be described by the $K$ parameter as 316H is a not creep-brittle material.

7.9.2 Correlation using $C^*$

$\frac{da}{dt}$ vs $C^*$ correlation can be obtained as $\frac{da}{dt}$ was previously provided in the $K$ correlation and the $C^*$ in the different crack lengths was calculated. The relationship is shown in Figure 7.44. According to Figure 7.44, under different creep ductilities, the CCG rates correlated by $C^*$ has a similar scatter range and trends to the $K$ correlation as examined in Figure 7.42, the trends are also bounded by an upper line as $\varepsilon_f = 0.5\%$ and a lower bound as $\varepsilon_f = 1.0\%$. As reflected in Figure 7.44, scatter range only expands in the lower CCG rate which corresponds to
the stage where the crack has grown for a long time. In the crack initiation and early stage of crack growth, the scatter is within a limited range, thus a reasonable prediction using \( C^* \) correlation is appropriate, bounded with lower and upper shelf. Also seen in Figure 7.44, the reduction of creep ductility can facilitate the formation of creep damage and accelerate the cracking behaviour.

As seen in Figure 7.45, the upper and lower bound CCG data in the current simulation was compared with the NSWA prediction, consideration was also taken on the data bands on C(T) BM [160] and C(T) HAZ [150]. Compared with Figure 7.42, the CCG data band correlated by \( C^* \) had a less scatter range than \( K \), therefore, \( C^* \) gave a better prediction on C(T) BM and HAZ specimens. Except a small tail which developed in the shallow crack length, the CCG prediction on nozzle/header generally falls in the data band of C(T) BM material, due to the consistency with the trends in C(T) base metal data band, and also due to the steady state creep deformation and cracking dominated the region over long operating times. \( C^* \) is therefore likely to be the better correlation method. On the other hand, the CCG prediction was out of the range of the HAZ C(T) data band, since the cracks, though close to the HAZ region, were propagated toward the base metal.

The upper and lower shelf was bounded within the NSWA trends between plane strain and plane stress condition, the results obtained from continuum damage concept were less conservative and therefore recommended for use in the real component damage analysis.

### 7.10 Summary

In this chapter, an AISI 316H austenitic stainless steels weld header with a region of interest in the header/nozzle branch has been investigated. Reheat cracks have been found in the vicinity of the weld connect to the outlet nozzle. The re-heat crack behaviour was examined under microstructural observation and micro-hardness analysis. In addition, a simplified method without the need to perform weld simulation, was proposed in order to simulate the welding residual stress distribution in a header that relaxes during creep where re-heat cracking commonly occurs. Establishing consistency to the measured residual stress data allowed the
analysis of creep damage and the correlation of CCG rate to be carried out. The following points are highlighted.

1. The hardness profile was measured on a new weld and from a section of an ex-service header. The peak hardness measured in the HAZ and the weld metal was around 30% higher than the base metal. A comparison of hardness measurements for the new and ex-service component indicates that service exposure has softened the material. A reheat crack was found to initiate at the high hardness boundary between the weld and base metal and propagated towards the lower hardness region within the base metal.

2. The reheat crack under microstructural examination is found to be the result of the growth and coalescence of creep-induced cavities at grain boundaries which can form micro-cracks and eventually macro-cracks. The crack did not initiate at the weld fusion lines but very near it. The stress is more likely to be concentrated due to the geometrical constraint and the crack directions is mainly affected by the structural configuration of lowest resistance path rather than parent or weld properties in this 316H material.

3. In the simplified FE model, a fixed displacement model was applied and boundary conditions optimised to simulate an effective residual stress field representative of that measured on components prior to and after service. The model presented provided a good representation of the residual stress profile in the header. The results suggest that the simple model may be used to predict the creep relaxation, stress-redistribution and cracking in the component during service.

4. The fracture mechanics method using the reference stress based estimate of $C^*$ has been applied to calculate the creep crack growth rate based on the approximate NSW model. It proves the results that NSWA provide a conservative prediction on CCG rate, it may be due to the conservative creep ductility value.

5. The modelling employed for creep crack growth using the ductility exhaustion model from the multiaxial void growth model has been simplified to make it independent of the creep index $n$ for a range of steels. It has been shown that creep damage strongly depends on the stress-dependent creep strain rate and multiaxial creep ductility. Stress dependency is mainly affected by geometrical constraint, stress type and magnitude. Material inhomogeneity does not significantly affect the creep damage behaviour. Creep
ductility may affect the creep damage and cracking rate. The crack initiation stage may not take a large proportion of service time. The history of crack propagation can be predicted as cracks would grow within a constant rate at the early stage of service and grow until reaching a stable crack length. The growth rate and the stable crack length is significantly influenced by the creep ductility. Based on the lower shelf creep ductility data from the present study, the long term failure time and creep crack growth behaviour of the header may be predicted using a predicted higher creep ductility.

6. The correlation using $K$ and $C^*$ has similar trends and scatter range in the present study. $C^*$ is likely the better correlation method because it lies within the lower scatter range of the C(T) base metal data band. The prediction based on the continuum damage concept is well bounded by NSWA prediction. The consistency on C(T) base metal rather than C(T) HAZ suggests that the crack, though initiated near the HAZ, is likely to grow towards the base metal. Therefore, the weld or pre-strained creep properties are inappropriate for predicting such failures.
### 7.11 Tables

#### Table 7.1: Possible loading conditions

<table>
<thead>
<tr>
<th>Model ID</th>
<th>Load Type</th>
<th>Load</th>
<th>Boundary condition</th>
</tr>
</thead>
<tbody>
<tr>
<td>LC-B1</td>
<td>Bending</td>
<td>1000 KN</td>
<td>U1=U2=0 on surf1</td>
</tr>
<tr>
<td>LC-B2</td>
<td>Bending</td>
<td>1000 KN</td>
<td>U2=0 on surf2</td>
</tr>
<tr>
<td>LC-LCT1</td>
<td>Load-controlled tension</td>
<td>1300 KN</td>
<td>U2=0 on surf1</td>
</tr>
<tr>
<td>LC-DCT1</td>
<td>Displacement-controlled tension</td>
<td>0.25mm</td>
<td>U2=0 on surf1</td>
</tr>
<tr>
<td>LC-DCT2</td>
<td>Displacement-controlled tension</td>
<td>0.5mm</td>
<td>U2=0 on surf1</td>
</tr>
<tr>
<td>LC-DCT3</td>
<td>Displacement-controlled tension</td>
<td>1.0mm</td>
<td>U2=0 on surf1</td>
</tr>
<tr>
<td>LC-DCT4</td>
<td>Displacement-controlled tension</td>
<td>2.0mm</td>
<td>U2=0 on surf1</td>
</tr>
</tbody>
</table>

#### Table 7.2: Possible boundary conditions

<table>
<thead>
<tr>
<th>Model ID</th>
<th>Displacement Controlled Load</th>
<th>U1 on surf 1</th>
<th>U2 on surf 1</th>
<th>U2 on surf 2</th>
</tr>
</thead>
<tbody>
<tr>
<td>BC-1</td>
<td>0.5 mm</td>
<td>-</td>
<td>Fixed</td>
<td>-</td>
</tr>
<tr>
<td>BC-2</td>
<td>0.5 mm</td>
<td>Fixed</td>
<td>Fixed</td>
<td>-</td>
</tr>
<tr>
<td>BC-3</td>
<td>0.5 mm</td>
<td>-</td>
<td>-</td>
<td>Fixed</td>
</tr>
</tbody>
</table>

### 7.12 Figures

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$\sigma_{33}$:

Hoop stress

Maximum principal stress

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Figure 7.35: Crack length prediction using uncracked and pre crack models, at $\varepsilon_f = 0.5\%$.
Model 1: smooth corner
Model 2: 0.2 mm defect
Model 2: 0.5 mm defect

Crack length, mm
Crack growth time, h
$\varepsilon_f = 0.9\%$

Figure 7.36: Crack length prediction using uncracked and pre crack models, at $\varepsilon_f = 0.9\%$

DA2-1: $\varepsilon_f=0.50 \%$
DA2-2: $\varepsilon_f=0.90 \%$
DA2-3: $\varepsilon_f=1.0 \%$
DA2-4: $\varepsilon_f=1.2 \%$
DA2-5: $\varepsilon_f=1.5 \%$
DA2-6: $\varepsilon_f=2.0 \%$

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Chapter 8

Conclusions and Future Work

8.1 Conclusions

In the power generation and nuclear industries, predicting the life of aging equipment has become both a safety issue and an economic necessity. The lifetime of components operating at high temperatures in the range of 500-650°C is limited by processes such as creep crack initiation and growth. Due to the existence of welding residual stresses and inhomogeneous properties in weldments, defects may initially exist and propagate in this region of the weldment. For this reason, research work on the creep behaviour of welds in structural components is of great importance and has been carried out in this PhD program.

The first part of this work as discussed in Chapter 3 was to determine the geometry dependent fracture mechanics parameter $\eta$ factor on the six fracture mechanics geometries which contain a thin section of welds. These values can be used to derive appropriate $J$ and $C^*$ estimations for welded fracture mechanics specimens. Since the previous work was mostly limited to homogeneous materials, this research have discussed the influence of weld mismatch ratio for a range of weld width, power-law hardening exponents on the solutions of the fracture mechanics parameters for a range of crack lengths under both plane strain and plane stress conditions on both $\eta^{\text{CMOD}}$ and $\eta^{\text{LLD}}$. The results have now been implemented in ASTM E1457.

It has been found that the weld width, $h/W$ could affect the value of $\eta^{\text{CMOD}}$ and $\eta^{\text{LLD}}$. In extreme over-matched conditions for $M = 2$, average scatter of $\pm 20\%$ for $\eta^{\text{CMOD}}$ and $\pm 50\%$ for $\eta^{\text{LLD}}$ can be expected. $\eta^{\text{CMOD}}$ would be the preferred method in this case due to its less sensitivity on the effect of $h/W$. By limiting the weld width to a thin section in the case of $h/W (\text{or} ~ 2W) \leq 0.1$, the variations of $\eta^{\text{CMOD}}$ and $\eta^{\text{LLD}}$ due to $h/W$ can reduce to $15\%$ less than that of homogeneous materials.
Also found that in most cases, material’s hardening exponent and stress state (i.e. \( P_\varepsilon, P_\sigma \)) would bring at worst \( \pm 20\% \) contribution to the both \( \eta^{CMOD} \) and \( \eta^{LLD} \). Therefore, a valid \( \eta \) prediction for different mismatch ratios can be performed with a combination of different values of \( N, h/W \) and \( P_\varepsilon, P_\sigma \).

Calculated values of \( \eta^{CMOD} \) and \( \eta^{LLD} \) for specific \( M \) and \( a/W \) with upper/lower bound deviations in six studied fractural geometries were examined. These values can be used to derive appropriate \( J \) and \( C^* \) estimations for weldment specimens. To describe the influence of mismatch ratio on \( \eta \), both \( \eta^{CMOD} \) and \( \eta^{LLD} \) in this specific range of crack length can be obtained using suggested equations derived from the present numerical studies. Obviously seen in the equations is that value of \( \eta^{CMOD} \) and \( \eta^{LLD} \) for a given crack length decrease as mismatch ratio increases, but the variations of \( \eta \) due to mismatch effects are mostly within the variation of \( \pm 30\% \) compared with homogeneous materials. It means that \( \eta \) factor in a thin welded fracture mechanics geometry is close to that in homogeneous specimens.

In both homogenous and inhomogeneous materials, the use of \( \eta \) under CMOD records has proved to be an effective method to evaluate \( J \) and \( C^* \) in SEN(T), DEN(T), M(T) and the use of \( \eta \) using LLD measurements is preferred for CS(T) specimens. \( \eta^{LLD} \) is identical to \( \eta^{CMOD} \) in C(T) specimen. \( \eta^{CMOD} \) and \( \eta^{LLD} \) are suitable for shallow and deep cracked SEN(B) respectively. It should be noted that the findings show that \( \eta \) does not show a clear relationship with the input variables used and at best the values range in a scatter of about \( \pm 30\% \). This also shows that there is no need for an exact use of \( \eta \) in calculating \( J \) and \( C^* \) in this way that the estimation procedures can be effectively simplified.

The next objective in the work was to experimentally investigate and characterise the appropriate material properties and compare them with other data sets of 316H from different branches of materials. The results were then applied in modelling the crack growth behaviour welded and as-received 316H components at 550 °C. Tensile, uniaxial creep rupture and creep crack growth behaviour in cross-weld (XW) region extracted from an ex-service 316H stainless
steel weld header were studied and compared with 8% pre-compressed (PC) specimen at room temperature and as-received (AR) specimen.

Results show that XW and PC specimens have similar material hardening effects which may results in limited extent of plastic deformation compared to the AR specimen. The properties of secondary creep strain rate, $\dot{\varepsilon}_s$, average creep strain rate, $\varepsilon$ and time to rupture, $t_R$ for a range of true stress have also shown that the XW and PC materials have similar creep properties and therefore, the PC specimens may be used to predict the creep rupture behaviour of the XW specimens. Also compared with the AR specimens, the XW specimens have a similar average creep strain rate, close but slightly higher secondary creep strain rate and significant shorter rupture time for a given stress. Growth and coalescence of creep-induced cavities mostly initiated in the XW region may accelerate the failure procedure and result in a faster rupture procedure.

The creep crack growth tests were performed in the weldments as well as the region near the weld in the ex-service header. For the cracking behaviour in the base metal the CCI and CCG response was similar to AR materials, since the crack initiated near the HAZ has grown toward the base metal. When the crack initiated in the HAZ, the CCG data was found to be higher than that in the AR materials due to the higher crack tip constraint and reduced creep ductilities. For those two conditions, the NSW-MOD model provided good predictions of the steady state CCG behaviour of the HAZ and AR materials when the creep ductility was derived from reduction of area values. The difference of CCI properties between the HAZ, PC and AR has been briefly discussed. For a given value of $C^*$, the initiation time in the HAZ and PC is around 10 times shorter than those for the AR material, effectively due to the lower ductility.

The last section as described in Chapter 7 was to investigate the reheat cracking in the ex-service weld header with the region of interest in the header’s branch location near to the nozzle where the crack was found. The circumferential cracks were initiatied at the radius from the nozzle near the weld toe and grew normal to the corner of the joint of header and nozzle away from the weld. The cracking position was in the vicinity of, but not on, the fusion line. It may be argued that lower ductility were present in that region from the micro-hardness measurements where average hardness values are approximately 10-30% higher than the base metal and from the tensile tests.
measured by DIC where the HAZ remains the lowest tensile strains. However the more important point in relation to the creep ductility was that the short term test performed could not be indicative of the creep ductility at the operational loads and at the long period of service time. For this reason using available extrapolated data from the literature, it was determined that a lower shelf ductility may be more appropriate for modelling the crack in the region of interest in the header.

A two-dimensional axisymmetric FE model was used to study cracking region on the nozzle/header branch. Without considering weld simulation, a simplified method using appropriate fixed displacement loads and boundary conditions was used to simulate an effective residual stress distribution that relaxed during creep where re-heat cracking commonly occurred. In the simplified model, a fixed displacement under an optimised boundary conditions can provide a good representation of the pre- and post-service residual stress profile when compared with measured residual stress.

For the creep crack growth evaluation, fracture mechanics method was firstly applied based on the $C^*$ parameter using the reference stress at the start of the crack. The reference stress for simple primary displacement controlled loading is determined by the methods of limit load, in this case the limit load is obtained from elastic-perfect plastic analysis. The creep crack growth was evaluated using the approximate NSW model which provided a conservative result on the CCG prediction. A more appropriate method based on continuum damage mechanics were then performed.

The creep damage has been estimated based on the ductility exhaustion concept where the creep damage parameter is mainly resultant from the creep strain rate and creep ductility either of which may be stress dependent. A simplified void growth model, which independent of the creep exponent $n$ for most engineering alloy was used to predict crack growth rate and direction. The stress distribution determining crack growth is primarily affected by the geometrical constraint. However the cracks may often initiate at the nozzle/header branch at a sharp angle due to the concentration of localised welding residual stress. Material inhomogeneity on the other hand, did not influence the crack initiation and propagation behaviour. Creep ductility sensitivity analysis showed that creep ductility would not affect the crack direction but would significantly affect the
creep damage history. The damage history has been well-predicted as the cracks would initiate
in a short time and then grows in a constant rate, until reaching a stable value, using the element
removal or partial surface hardening method. The increasing creep ductility would significantly
reduce the creep crack growth rate and lower the stable crack length. In the present model for the
header with an existence of cracks of 8 – 15 mm, it has been shown that the damage and creep
crack growth history could be predicted with an optimised creep ductility value which may in
the range of 3.0% – 3.8%.

Finally, $C^*$ has successfully correlated creep crack growth behaviour of the simulated
header/nozzle components as well as C(T) specimen. However $K$ may also be appropriate
especially since it is assumed that the long term cracking in the header would see lower shelf
creep ductility. $C^*$ is likely to be the better correlation method because it is consistent with the
trends examined in C(T) specimens and consistent with the time-dependent failure mechanism
that prevails at operating temperatures.

### 8.2 Future Work

Although the effect of $\eta$ in calculating $J$ and $C^*$ is small it is still useful to try to improve
numerical accuracy in all areas of fracture mechanics. Part of the scatter in the data is in fact due
to lack of details. As mentioned in Chapter 3, an alternative method to evaluate $J_p$ is to use
load separation analysis to evaluate $\eta$ for conventional fracture specimens. This theory was
initially proposed by Paris et. Al. [118] who assumes geometry-dependent $\eta$ is related to a crack
growth function and a material deformation function. The load separation criterion has been
studied for several different geometries in homogeneous material [112, 124, 146, 162, 163].
Currently, any results using load separation analysis in welded specimens are not published and
therefore required to be investigated.

The testing results discussed in Chapter 5 and 6 are based on a limit number of tests performed
on tensile, uniaxial creep rupture and creep crack growth cross-weld specimens. In order to
provide more confident results on HAZ properties from such tests, more tests need to be
performed in the future. Cross-weld specimens are assumed to be insignificant to cast-to-cast
variability due to the similarity to pre-compressed material. Due to the lack of XW specimen
from different headers containing same base and weld metal, the assumption in cast-to-cast variability is required to be proved in the future examinations.

The discussion on FE model performed in Chapter 7 is based on a 2D axi-symmetric model. 3D FEM model could be used to examine the whole range of cracking region. However this route would only be advisable once there is increased confidence in the material properties used. In addition, since creep damage is significantly sensitive to creep ductility and geometrical constraint, further understanding of the role of ductility dependency on the model is needed by modelling different shapes of headers with different material properties. Further specific tests can provide reliable creep ductility examination which will assist in the life assessment of structural components. Furthermore modelling and meshing improvements, for example simulating actual grain shapes and sizes may allow a more realistic approach to predict these cracks.
References


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149. Davies, C.M., et al., Compressive pre-strain effects on the creep and crack growth behaviour of 316H stainless steel, in proceedings of the International Conference on Pressure Vessels and Piping. 2010: Seattle, US.


164. Spindler, M.W., Creep ductility of Type 316 and 316H austenitic stainless steels between 500 and 700 C. 1996: British Energy Generation Ltd.


Appendix

A. Calculated Values of $\eta$ for Weldment Fracture Geometries

A.1 Introduction

The effects of $\eta^{LLD}$ and $\eta^{CMOD}$ have been systematically investigated on a range of centre-cracked, square-grooved weld fracture geometries. For each geometry, calculated values of $\eta^{LLD}$ and $\eta^{CMOD}$ in FEA have been obtained for specific weld mismatch ratio in a range of $0.5 \leq M \leq 2$, $h/W, 0.05 \leq h/W$ (or $2W) \leq 0.10$, hardening exponent $N 5 \leq N \leq 20$, and crack length $0.1 \leq a/W \leq 0.7$ under both plane strain and plane stress conditions. The results are tabulated in the following section.

A.2 FE Results

A.2.1 Compact Tension, C(T)

Table A.1: Calculated values of $\eta^{LLD}$ for specific crack length, weld width, hardening exponent ($N$), mismatch ratio under both plane stress and plane strain conditions in C(T)

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<th>$\eta^{LLD}$ Plane Stress</th>
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A.2.2 C-Shape Tension CS(T)

Table A.2: Calculated values of $\eta^{CMOD}$ for specific crack length, weld width, hardening exponent ($N$), mismatch ratio under both plane stress and plane strain conditions in CS(T)

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Table A.3: Calculated values of \(\eta_{\text{LLD}}^{\text{CMOD}}\) for specific crack length, weld width, hardening exponent (\(N\)), mismatch ratio under both plane stress and plane strain conditions in CS(T)

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218
### A.2.3 Single Edge Notch in tension SEN(T)

Table A.4: Calculated values of $\eta^{\text{CMOD}}$ for specific crack length, weld widths, hardening exponent ($N$), mismatch ratios under both plane stress and plane strain conditions in SEN(T)

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Table A.5: Calculated values of $\eta^{LLD}$ for specific crack length, weld widths, hardening exponent ($N$), mismatch ratios under both plane stress and plane strain conditions in SEN(T)

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\[
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222
A.2.4 Single Edge Notch in Bending SEN(B)

Table A.6: Calculated values of $\eta_{CMOD}$ for specific crack length, weld width, hardening exponent ($N$), mismatch ratio under both plane stress and plane strain conditions in SEN(B)

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For $M = 1.0$

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For $M = 1.5$

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### Table A.7: Calculated values of $\eta^{LLD}$ for specific crack length, weld width, hardening exponent ($N$), mismatch ratio under both plane stress and plane strain conditions in SEN(B)

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### A.2.5 Middle Crack in Tension, M(T)

Table A.8: Calculated values of $\eta^{CMOD}$ for specific crack length, weld widths, hardening exponent ($N$), mismatch ratios under both plane stress and plane strain conditions in M(T)

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<th>$\eta^{CMOD}$ Plane Stress</th>
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226
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Table A.9: Calculated values of \(\eta^{LLD}\) for specific crack length, weld widths, hardening exponent (\(N\)), mismatch ratios under both plane stress and plane strain conditions in M(T)

\[ M = 0.5 \]

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\[ M = 1.0 \]

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227
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### A.2.6 Double Edge Notch in Tension, DEN(T)

Table A.10: Calculated values of \( \eta^{\text{CMOD}} \) for specific crack length, weld widths, hardening exponent \((N)\), mismatch ratios under both plane stress and plane strain conditions in DEN(T)

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#### M = 1.5

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Table A.11: Calculated values of $\eta^{LLD}$ for specific crack length, weld widths, hardening exponent ($N$), mismatch ratios under both plane stress and plane strain conditions in DEN(T)

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$M = 1.0$

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### M = 1.5

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<td>0.41</td>
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</table>

### M = 2.0

<table>
<thead>
<tr>
<th>(a/W)</th>
<th>(h/2W)</th>
<th>(\eta^{LLD}) Plane Strain</th>
<th>(\eta^{LLD}) Plane Stress</th>
</tr>
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<tbody>
<tr>
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<td>N=5</td>
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<tr>
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</table>
B. Calculated Values of $\eta^{LLD}$ and $\eta^{CMOD}$ for Specific Crack Length and Mismatched Ratio with Upper/Lower Bound Deviations in a Range of Fractural Geometries

B.1 Introduction

The effects of $h/W$, $N$, $P\varepsilon$, $P\sigma$ on $\eta$ have been individually examined as shown in Appendix B, however, the main objective was to only consider the sensitivity of $\eta$ to $M$ and $a/W$. Having considered the low sensitivity for $\eta$ with respect to $N$, $h/W$ and $P\varepsilon$, $P\sigma$ in most geometries, average $\eta$ values have been obtained under various conditions. The average values of $\eta$ and the level of deviations from the mean values are tabulated for specific crack length and mismatch ratio in each geometry.

B.2 FE Results

B.2.1 Compact Tension, C(T)

Table B.1: Calculated values of $\eta^{LLD}$ for specific crack length and mismatched ratio with upper/lower bound deviations in C(T)

<table>
<thead>
<tr>
<th>$\eta^{LLD}$ $a/W$</th>
<th>C(T) base ($\pm0.10$)</th>
<th>Under-match ($M=0.5$) ($\pm0.20$)</th>
<th>Over-match ($M=1.5$) ($\pm0.18$)</th>
<th>Over-match ($M=2$) ($\pm0.14$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.35</td>
<td>1.89</td>
<td>2.24</td>
<td>1.73</td>
<td>1.63</td>
</tr>
<tr>
<td>0.40</td>
<td>2.07</td>
<td>2.35</td>
<td>1.92</td>
<td>1.81</td>
</tr>
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<td>0.45</td>
<td>2.20</td>
<td>2.61</td>
<td>2.10</td>
<td>2.00</td>
</tr>
<tr>
<td>0.50</td>
<td>2.20</td>
<td>2.61</td>
<td>2.10</td>
<td>2.00</td>
</tr>
<tr>
<td>0.60</td>
<td>2.20</td>
<td>2.60</td>
<td>2.10</td>
<td>2.00</td>
</tr>
<tr>
<td>0.70</td>
<td>2.20</td>
<td>2.60</td>
<td>2.10</td>
<td>2.00</td>
</tr>
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### B.2.2 C-Shape Tension, CS(T)

Table B.2: Calculated values of $\eta^{LLD}$ for specific crack length and mismatched ratio with upper/lower bound deviations in CS(T)

<table>
<thead>
<tr>
<th>$a/W$</th>
<th>CS(T) base</th>
<th>Under-match ($M=0.5$)</th>
<th>Over-match ($M=1.5$)</th>
<th>Over-match ($M=2$)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$\eta^{CMOD}$</td>
<td>$\eta^{LLD}$</td>
<td>$\eta^{CMOD}$</td>
<td>$\eta^{LLD}$</td>
</tr>
<tr>
<td></td>
<td>(+0.16)</td>
<td>(+0.23)</td>
<td>(+0.30)</td>
<td>(+0.24)</td>
</tr>
<tr>
<td>0.20</td>
<td>4.08</td>
<td>2.05</td>
<td>4.52</td>
<td>2.31</td>
</tr>
<tr>
<td>0.30</td>
<td>4.08</td>
<td>2.16</td>
<td>4.52</td>
<td>2.38</td>
</tr>
<tr>
<td>0.40</td>
<td>3.92</td>
<td>2.27</td>
<td>4.27</td>
<td>2.45</td>
</tr>
<tr>
<td>0.50</td>
<td>3.76</td>
<td>2.23</td>
<td>4.03</td>
<td>2.38</td>
</tr>
<tr>
<td>0.60</td>
<td>3.61</td>
<td>2.19</td>
<td>3.79</td>
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<tr>
<td>0.70</td>
<td>3.45</td>
<td>2.15</td>
<td>3.55</td>
<td>2.23</td>
</tr>
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### B.2.3 Single Edge Notch in tension, SEN(T)

Table B.3: Calculated values of $\eta^{LLD}$ for specific crack length and mismatched ratio with upper/lower bound deviations in SEN(T)

<table>
<thead>
<tr>
<th>$a/W$</th>
<th>SEN(T) base</th>
<th>Under-match ($M=0.5$)</th>
<th>Over-match ($M=1.5$)</th>
<th>Over-match ($M=2$)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$\eta^{CMOD}$</td>
<td>$\eta^{LLD}$</td>
<td>$\eta^{CMOD}$</td>
<td>$\eta^{LLD}$</td>
</tr>
<tr>
<td></td>
<td>(+0.05)</td>
<td>(+0.40)</td>
<td>(+0.06)</td>
<td>(+0.51)</td>
</tr>
<tr>
<td>0.10</td>
<td>1.00</td>
<td>0.80</td>
<td>1.06</td>
<td>1.22</td>
</tr>
<tr>
<td>0.20</td>
<td>1.00</td>
<td>1.24</td>
<td>1.06</td>
<td>1.65</td>
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<tr>
<td>0.30</td>
<td>1.00</td>
<td>1.68</td>
<td>1.06</td>
<td>2.09</td>
</tr>
<tr>
<td>0.40</td>
<td>1.00</td>
<td>2.11</td>
<td>1.06</td>
<td>2.53</td>
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<tr>
<td>0.50</td>
<td>1.00</td>
<td>2.55</td>
<td>1.06</td>
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<td>2.55</td>
<td>1.06</td>
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<td>1.00</td>
<td>2.55</td>
<td>1.06</td>
<td>2.60</td>
</tr>
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### B.2.4 Single Edge Notch in Bending, SEN(B)

Table B.4: Calculated values of $\eta^{LLD}$ for specific crack length and mismatched ratio with upper/lower bound deviations in SEN(B)

<table>
<thead>
<tr>
<th>$a/W$</th>
<th>SEN(B) base</th>
<th>Under-match ($M=0.5$)</th>
<th>Over-match ($M=1.5$)</th>
<th>Over-match ($M=2$)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$\eta^{CMOD}$</td>
<td>$\eta^{LLD}$</td>
<td>$\eta^{CMOD}$</td>
<td>$\eta^{LLD}$</td>
</tr>
<tr>
<td></td>
<td>(+0.07)</td>
<td>(+0.20)</td>
<td>(+0.05)</td>
<td>(+0.36)</td>
</tr>
<tr>
<td>0.10</td>
<td>1.08</td>
<td>1.00</td>
<td>1.13</td>
<td>1.13</td>
</tr>
<tr>
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<td>0.86</td>
<td>1.32</td>
<td>0.90</td>
<td>1.45</td>
</tr>
<tr>
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<td>0.84</td>
<td>1.76</td>
</tr>
<tr>
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<td>0.75</td>
<td>1.95</td>
<td>0.79</td>
<td>2.08</td>
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</table>
### B.2.5 Middle Crack in Tension, M(T)

Table B.5: Calculated values of $\eta^{LLD}$ for specific crack length and mismatched ratio with upper/lower bound deviations in M(T)

<table>
<thead>
<tr>
<th>$a/W$</th>
<th>M(T) base</th>
<th>Under-match ($M=0.5$)</th>
<th>Over-match ($M=1.5$)</th>
<th>Over-match ($M=2$)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
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<td>$\eta^{LLD}$</td>
<td>$\eta^{CMOD}$</td>
<td>$\eta^{LLD}$</td>
</tr>
<tr>
<td>0.20</td>
<td>(±0.07)</td>
<td>(±0.30)</td>
<td>(±0.14)</td>
<td>(±0.25)</td>
</tr>
<tr>
<td>0.30</td>
<td>0.95</td>
<td>0.55</td>
<td>1.13</td>
<td>1.08</td>
</tr>
<tr>
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<td>1.08</td>
</tr>
<tr>
<td>0.50</td>
<td>0.92</td>
<td>0.68</td>
<td>1.10</td>
<td>1.08</td>
</tr>
<tr>
<td>0.60</td>
<td>0.91</td>
<td>0.75</td>
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<tr>
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### B.2.6 Double Edge Notch in Tension, DEN(T)

Table B.6: Calculated values of $\eta^{LLD}$ for specific crack length and mismatched ratio with upper/lower bound deviations in DEN(T)

<table>
<thead>
<tr>
<th>$a/W$</th>
<th>DEN(T) base</th>
<th>Under-match ($M=0.5$)</th>
<th>Over-match ($M=1.5$)</th>
<th>Over-match ($M=2$)</th>
</tr>
</thead>
<tbody>
<tr>
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<td>$\eta^{LLD}$</td>
<td>$\eta^{CMOD}$</td>
<td>$\eta^{LLD}$</td>
</tr>
<tr>
<td>0.20</td>
<td>(±0.15)</td>
<td>(±0.23)</td>
<td>(±0.08)</td>
<td>(±0.25)</td>
</tr>
<tr>
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</tr>
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<tr>
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<td>0.95</td>
<td>0.70</td>
</tr>
<tr>
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<td>0.95</td>
<td>0.76</td>
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