Predicting Creep Crack Initiation in Austenitic and Ferritic Steels using the Creep Toughness Parameter and Time Dependent Failure Assessment Diagram

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Abstract

Methods for evaluating the creep toughness parameter, K_c^{mat} , are reviewed and K_c^{mat} data are determined for a ferritic P22 steel from CCG tests on compact tension, C(T), specimens of homogenous parent material (PM) and heterogeneous specimen weldments at 565 °C and compared to similar tests on austenitic 316H stainless steel at 550 °C. Appropriate relations describing the time dependency of K_c^{mat} are determined accounting for data scatter. Considerable differences are observed in the form of the K_c^{mat} data and the time dependent failure assessment diagrams (TDFADs) for both the 316H and P22 steel. The TDFAD for P22 shows a strong time dependency but is insensitive to time for 316H. Creep crack initiation (CCI) times predictions are obtained using the TDFAD approach and compared to experimental results from C(T) specimens and feature components. The TDFAD based on PM properties can be used to obtain conservative prediction of the CCI on the weldments. Conservative predictions are almost always obtained when lower bound K_c^{mat} values are employed. Long term test are generally more relevant to industrial component lifetimes. The different trends between long and short term CCI time and growth data indicate that further long term test are required to further validate the procedure to predict the lifetimes of high temperature component lifetimes.

Keywords

TDFAD, Creep Toughness, Creep Crack Initiation (CCI), 316H Steel, P22 Steel, Weldments.

Nomenclature

$a, \Delta a$	= Crack length, Crack extension
$\dot{a}$	= Crack growth rate
A	= Creep strain rate coefficient
A_p	= Primary creep strain coefficient
B_n	= Specimen net section thickness between side-grooves
B_r	= Coefficient in the stress rupture time law
C^*	= Steady state creep characterising fracture mechanics parameter
CCG	= Creep crack growth
CCI	= Creep crack initiation
D	= Constant coefficient in creep crack growth rate correlation with C^*
D_i	= Constant coefficient in creep crack initiation time correlation with C^*
E	= Elastic (Young's) modulus
E'	= Effective elastic modulus
H	= Correlating coefficient in the K_c^{mat} vs. time relation
HAZ	= Heat affected zone
J	= Elastic-Plastic fracture mechanics parameter

j	= The power-law exponent in the K_{mat}^c vs. time relation
J_{IC}	= Material Mode I plane strain elastic-plastic fracture toughness
K	= Linear-Elastic stress intensity factor
K_{IC}	= Material Mode I plane strain fracture toughness
K_r	= TDFAD parameter measuring proximity to fracture
K_{mat}^c	= Creep toughness parameter
LB	= Lower bound
L_r	= TDFAD parameter measuring proximity to plastic collapse or creep rupture
L_r^{max}	= Maximum value of L_r at the cut-off point in TDFAD
n	= Creep stress exponent
n_p	= Primary creep stress exponent
P	= Load
p	= Primary creep parameter
P_{LC}	= Plastic collapse load of cracked body
PM	= Parent material
ROA	= Reduction of area
$t, \Delta t$	= Time, time increment
t_i	= Creep crack initiation time
t_i^{Exp}	= Experimentally measured creep crack initiation time
$t_i^{Predicted}$	= Predicted value of the creep crack initiation time
UB	= Upper bound
U_p	= Area under the load displacement curve associated with plasticity
ν	= Poisson ratio
ν_r	= Rupture stress exponent
W	= Specimen width
WM	= Weldment
Δ^c	= Load line displacement associated with creep
ε^c	= Strain component associated with creep deformation
ε^e	= Strain component associated with elastic deformation
ε^{pl}	= Strain component associated with plastic deformation
ε^{total}	= Total strain
ε_f	= Plastic ductility
ε_f^c	= Creep failure strain (creep ductility)
ε_{ref}	= Total strain at reference stress
ε_{ref}^e	= Strain at reference stress associated with elastic deformation
ε_{ref}^p	= Strain at reference stress associated with plastic deformation
ε_{ref}^c	= Strain at reference stress associated with creep deformation
$\Delta\varepsilon_p^c, \Delta\varepsilon_s^c$	= Primary and secondary creep strain increments
ϕ_i	= Exponent in correlation of initiation time with C^*
ϕ	= Exponent in correlation of creep crack growth rate with C^*
η	= Factor relating J/C^* to the plastic/creep area under a load displacement curve
$\sigma_{0.2}$	= 0.2% proof strength
σ_{UTS}	= Ultimate tensile stress
$\sigma_{0.2}$	= 0.2% proof stress of the material

$\sigma_{0.2}^c$ = 0.2% creep strength (stress at 0.2 % inelastic strain)

σ_{flow} = Flow stress $(\sigma_{UTS} + \sigma_{0.2})/2$

1 Introduction

The lifetime of components operating in high temperature plant is limited by the mechanisms of creep crack initiation (CCI) and growth (CCG). The CCI time is the period of time prior to the onset of crack extension from an existing defect due to creep crack growth and can occupy large fraction (> 80%) of a components service lifetime ¹. The CCI time is generally defined as the time for a defined small measureable crack extension, typically 0.2 mm for laboratory specimens or 0.5 mm ². Reliable predictions of CCI time are fundamental in high temperature component lifetime assessments due to the duration of the CCI period.

The time dependant failure assessment diagram (TDFAD) approach has become increasingly recognised for the prediction of creep crack initiation (CCI) times ³. The development of the TDFAD has mainly been based on austenitic type 316H stainless steel which differs from ferritic steels in their tensile and creep material behaviour and further verification are required before the procedure may be recommended for ferritic steels ^{4, 5}. Two widely used alloys for high-temperature plant components with contrasting properties are ferritic P22 (2¼Cr1Mo) steel and austenitic type 316H stainless steel. The favourable qualities of these steels include high creep ductility and relatively high weldability ^{6, 7}. These two steels are considered here in the evaluation of the TDFAD method.

The TDFAD approach for CCI time predictions relies on the availability of reliable creep toughness (K_{mat}^c) data. The main advantage of the TDFAD approach is that detailed calculations of crack tip parameters such as C^* are not required and that the TDFAD can indicate whether failure is controlled by crack growth, creep rupture or plastic collapse i.e. the fracture regime need not be specified in advance. Methods for evaluation or estimating the K_{mat}^c parameter have been developed and modified over recent years.

A considerable amount of creep toughness data has been determined for austenitic type 316H stainless steel in the region of 500–700° C, though mainly at 550° C ^{8, 9}. Limited K_{mat}^c data have also been determined for a range of ferritic steels ^{4, 5, 8, 10-12}, and in some cases TDFADs have also been produced. Creep toughness data often show appreciable scatter ^{8, 9} which must be considered when obtaining CCI time predictions. In the majority of cases, where TDFADs have been determined together with creep toughness data, explicit CCI times have not been obtained. Generally, the CCI data for a given crack extension have been plotted on a TDFAD for a range of times and an assessment made to determine if the TDFAD approach leads to conservative predictions, or sometimes a range of CCI times indicated. The degree of conservatism/non-conservatism however, is not indicated.

Considerably high loads are often applied to specimens in CCG tests, leading to relatively short test durations compared to the lifetimes of high temperature components operating in plants. Therefore, the application of methods based on CCG test data to predict real component lifetimes require validation ⁴. Components are often welded and the application of the TDFAD approach for weldments has been considered ¹²⁻¹⁶. However, a number of complications arise when considering weldments such as material mis-match effects which may require consideration.

In previous work ⁹, the TDFAD approach has been applied to predict CCI times in austenitic type 316H stainless steel at 550° C, and the sensitivity of predictions to creep toughness data bounds and reference stress solutions have been examined. In this work methods for K_{mat}^c evaluation and estimation are reviewed and K_{mat}^c data are determined for a ferritic P22 steel from CCG tests on compact tension, C(T), specimens of homogenous parent material and heterogeneous specimen weldments at 565 °C. The weldments consist of parent material (PM), weld metal and a heat affected zone (HAZ). The crack tip is located on the HAZ/PM boundary where cracking is often observed in practice. Appropriate relations describing the time dependency of K_{mat}^c are determined accounting for data scatter. The TDFAD diagram and its associated parameters are then compared for the austenitic type 316H and ferritic P22 steel. The influence of test duration, and material condition are also considered and CCI time predictions obtained and compared to experimental results from short and long term laboratory tests specimens and feature components.

2 The Failure Assessment Diagram (FAD) Approach

The FAD procedure considers that failure will occur by plastic collapse or brittle/ductile fracture. The proximity to failure by fracture and plastic collapse are measured by the parameters K_r and L_r , respectively, defined by

$$K_r = \frac{K}{K_{IC}} = \left(\frac{K^2}{E'J_{IC}} \right)^{\frac{1}{2}} \quad (1)$$

$$L_r = \frac{\sigma_{ref}}{\sigma_{0.2}} = \frac{P}{P_{LC}} \quad (2)$$

where K_{IC} and J_{IC} are fracture toughness values (critical values of K and J for fracture under Mode I, tensile loading), E' the effective elastic modulus (equal to E for plane stress or $E/(1 - \nu^2)$ for plane strain conditions where ν is the Poisson ratio). In Eqn (2) σ_{ref} is the reference stress ¹ of a geometry, $\sigma_{0.2}$ is the 0.2% proof stress (measure of the materials yield stress), P the applied load and P_{LC} the collapse load of the cracked geometry ¹.

The R6 Option 1 curve ^{17, 18} is material independent and defined by

$$K_r = \left(1 - 0.14L_r^2\right) \left[0.3 + 0.7 \exp\left(-0.65L_r^6\right)\right] \quad \text{for } L_r \leq L_r^{max} \quad (3)$$

$$K_r = 0 \text{ for } L_r > L_r^{\max} \quad (4)$$

The cut-off, $L_r^{\max}$, which indicates failure by plastic collapse is defined by the ratio

$$L_r^{\max} = \frac{\sigma_{flow}}{\sigma_{0.2}} \quad (5)$$

where σ_{flow} is the mean of the ultimate tensile stress, σ_{UTS} , obtained from the engineering stress-strain curve and the 0.2% proof stress i.e. $\sigma_{flow} = (\sigma_{UTS} + \sigma_{0.2})/2$

The R6 Option 2 FAD has a material specific failure assessment curve (FAC) which has been derived based on the assumption that crack growth occurs when the J -Integral attains a critical value.

2.1 The Time Dependent Failure Assessment Diagram (TDFAD)

The R6 Option 2 FAD has been extended to a time dependent failure assessment diagram (TDFAD) which addresses limited high temperature crack growth³. This is done by replacing K_{IC} , in Eqn (1) by a creep toughness corresponding to a given crack extension at a given time, denoted $K_{mat}^c(\Delta a, t)$, and $\sigma_{0.2}$ in Eqn (2) by the 0.2% inelastic (creep and plastic) strain from an isochronous stress-strain curve at a particular time and temperature, $\sigma_{0.2}^c$, also called the 0.2% creep strength. The value of $\sigma_{0.2}^c$ will decrease as time increases i.e. creep strain increases. In the TDFAD, for the case of a single primary load, the parameters K_r and L_r are therefore defined as

$$K_r = \frac{K}{K_{mat}^c} \quad (6)$$

$$L_r = \frac{\sigma_{ref}}{\sigma_{0.2}^c} \quad (7)$$

The cut off point $L_r^{\max}$ is defined on the TDFAD as

$$L_r^{\max} = \frac{\sigma_r}{\sigma_{0.2}^c} \quad (8)$$

where σ_r is the stress to cause creep rupture at the same time as $\sigma_{0.2}^c$ is evaluated. If L_r exceeds $L_r^{\max}$, failure is expected to occur by creep rupture rather than by fracture. In order to be consistent with the R6 procedure, $L_r^{\max}$ in the TDFAD should not exceed the value of $L_r^{\max}$ defined in Eqn (5). The time to rupture in a uniaxial creep test over a range of stresses can often be approximated by the power-law relation,

$$t_r = B_r \sigma^{-v_r} \quad (9)$$

where B_r and v_r are the rupture coefficient and exponent, respectively.

The time dependent failure assessment curve is based on the assumption that crack growth occurs when the dependent time parameter, $J(t)$, attains a critical value. A failure assessment diagram for a specific time is defined by the equations

$$K_r = \left[\frac{E \varepsilon_{ref}}{L_r \sigma_{0.2}^c} + \frac{L_r^3 \sigma_{0.2}^c}{2E \varepsilon_{ref}} \right]^{-1/2} \text{ for } L_r \leq L_r^{\max} \quad (10)$$

$$K_r = 0 \text{ for } L_r > L_r^{\max} \quad (11)$$

In Eqn (10) ε_{ref} is the total strain at reference stress at a given time, given by the sum of the elastic and plastic strain and the total creep strain accumulated in that time i.e.

$$\varepsilon_{ref}(t) = \varepsilon_{ref}^e + \varepsilon_{ref}^p + \varepsilon_{ref}^c(t) \quad (12)$$

Note that ε_{ref} is the true strain at true stress $\sigma_{ref} (= L_r \sigma_{0.2})$ (and not the engineering strain). Note also that at short times $\varepsilon_{ref}(t) \approx \varepsilon_{ref}^e + \varepsilon_{ref}^p$ and $\sigma_{0.2}^c \approx \sigma_{0.2}$, and Eqn (10) reduces to the R6 Option 2 curve. At long times $\varepsilon_{ref}(t) \approx \varepsilon_{ref}^c$ and Eqn (10) reduces to

$$K_r = \left[\frac{E \varepsilon_{ref}^c}{\sigma_{ref}} \right]^{-1/2} \quad (13)$$

Equation (13) can be derived for steady state creep conditions, where the creep strain rate is a constant, based on the assumption that at long times $J(t)$, evaluated, can be approximated by the product of C^* and time (i.e. $J(t) \approx C^* t$ ¹⁹) and C^* is given by the reference stress estimate¹. The TDFAD can therefore measure the proximity to failure by fast fracture, creep crack growth, plastic collapse and creep rupture. Therefore, in the TDFAD approach, a failure mode does not have to be pre-defined.

2.2 Application of TDFAD to Predict Creep Crack Initiation (CCI)

The TDFAD can be used to predict if a crack will extend a distance Δa in a given time or the time required for a specified amount of crack extension. Since CCI can be defined as the period of time required for a small increment of crack growth, Δa , the TDFAD may be used to predict CCI times. For some materials the curves may not vary greatly with time and curves for longer times can be used to provide a conservative TDFAD for an assessment at shorter times⁸.

To predict CCI an initial time estimate is made and the values of L_r and K_r , and their associated parameters, are determined for the specified initiation distance, Δa , at that time. The point (L_r , K_r) is then plotted on the TDFAD. If the point lies within the TDFAD then the crack extension is less than Δa and creep rupture is avoided in the assessment time. To determine an initiation time, t_i , a time locus of points (L_r , K_r) is constructed at various times. The time for a crack extension Δa is given by the intersection of a point on this locus (for a given time) with the failure assessment curve for the corresponding time. An iterative process is required to match the times associated with the point of intersection of the locus and the TDFAD constructed. The procedure is further detailed in^{8,9}.

2.3 Isochronous Stress-Strain Data

Isochronous stress-strain curves for the specified temperature are required in order to determine $\sigma_{0.2}^c$ and the overall TDFAD. Isochronous stress-strain data are generated here using the elastic-plastic and creep material response. The method used follows the procedure in the RCC-MR design code ²⁰ for primary-secondary creep of Type 316 stainless steel material. Thus, the primary and secondary creep strain increments, $\Delta\epsilon_p^c$ and $\Delta\epsilon_s^c$, are calculated according to

$$\Delta\epsilon_p^c = p A_p^{1/p} \sigma^{n/p} \epsilon^{(1-1/p)} \Delta t \quad (14)$$

$$\Delta\epsilon_s^c = A \sigma^n \Delta t \quad (15)$$

The creep strain increment, $\Delta\epsilon^c$, is equal to the larger of the two increments calculated from Eqns (14) and (15) i.e.

$$\Delta\epsilon^c = \begin{cases} \Delta\epsilon_p^c & \text{for } \Delta\epsilon_p^c \geq \Delta\epsilon_s^c \\ \Delta\epsilon_s^c & \text{for } \Delta\epsilon_p^c < \Delta\epsilon_s^c \end{cases} \quad (16)$$

The primary and secondary creep constants in Eqns (14) and (15) are given in Table 1 for 316H stainless steel at 550° C and P22 steel at 565° C. For a particular time, the total strain at any stress level is given by the sum of the elastic and plastic strain and the total creep strain accumulated in that time, *i.e.*

$$\epsilon^{total}(\sigma, t) = \epsilon^e(\sigma) + \epsilon^{pl}(\sigma) + \sum_t \Delta\epsilon^c(\sigma, \Delta t) \quad (17)$$

3 Evaluation of the Creep Toughness Parameter, K_{mat}^c

The K_{mat}^c parameter is evaluated from the load displacement curve generated during a CCG test using the relation ^{3, 8}.

$$K_{mat}^c = \left(K^2 + \frac{E' U_p}{B_n (W-a)} \eta + \frac{n}{n+1} \frac{E' P \Delta^c}{B_n (W-a)} \eta \right)^{1/2} \quad (18)$$

where U_p is the area under the load displacement curve associated with plasticity, W is the specimen width, B_n is the net specimen thickness between any side-grooves, a is the crack length and n the secondary creep power-law stress exponent (see Eqn (15)) and η a geometry function ($\eta = 2.2$ for C(T) specimens ²¹). Note that the K_{mat}^c relation has been modified since the work in ⁹.

In the absence of specimen load-displacement data a method has been proposed to estimate K_{mat}^c from CCI and CCG data. Under steady state conditions the CCG rate, $\dot{a}$, may be described by the expression ²

$$\dot{a} = DC^{*\phi} \quad (19)$$

where D and ϕ are temperature dependent crack growth constants. The CCI time may also be described by the C^* parameter according to the power-law relationship

$$t_i = D_i C^{*-\phi_i} \quad (20)$$

where D_i and ϕ_i are temperature dependent CCI constants. Assuming widespread creep conditions, a constant secondary creep strain rate and using the reference stress estimate of C^{*-1} , it can be shown that K_{mat}^c may be estimated from CCI data according to the relation²²

$$K_c^{mat} \approx \left(E D_i^{1/\phi_i} t^{(1-1/\phi_i)} \right)^{1/2} \quad (21)$$

Alternatively, if crack growth is assumed a continuous process commencing at zero time, then K_{mat}^c may be estimated from CCG data²² using

$$K_c^{mat} \approx \left(E \left(\frac{\Delta a}{D} \right)^{1/\phi} t^{1-1/\phi} \right)^{1/2} \quad (22)$$

It can therefore be seen from Eqn (22) that K_{mat}^c is expected to decrease with time according to

$$K_c^{mat} \propto \left(t^{1-1/\phi} \right)^{1/2} \quad (23)$$

Creep crack initiation and growth models²³⁻²⁵ predict that $\phi_i = \phi = n/(n+1)$ thus, substituting for ϕ in Eqn (23) it may be written

$$K_c^{mat} \propto t^{-1/2n} \quad (24)$$

It is therefore expected that K_c^{mat} follows the power-law relationship

$$K_c^{mat}(\Delta a, t) = H t_i^{-j} \quad (25)$$

where H is the correlating coefficient and, j is the power-law exponent.

4 Creep Deformation and CCG Behaviour of P22 and 316H Steel

Firstly, the high temperature tensile and creep deformation and rupture behaviour of the two materials are compared to reveal their differences, which will be reflected in the form of their TDFADs. The true-stress vs. true-strain curves of both materials are shown in Figure 1 up to the point in the tests where a failure mechanism intervened. The tensile properties are given in Table 1. The shape of the tensile behaviour of P22 and 316H are clearly different. The 0.2% proof stress of P22 is almost 70% higher than that of 316H and exhibits relatively little hardening. On the contrary, 316H has a high degree of work hardening and has a 10% higher plastic ductility, ε_f , than P22 steel, based on engineering strain (Eng) values.

The secondary creep strain rate and rupture-time vs. stress relations for the two steels are compared in Figure 2(a) and (b), respectively. Note that the stress values for 316H are true-stress values accounting for plastic strain on loading. However, in the absence of loading data, and since the applied stress are significantly less than

the $\sigma_{0.2}$ of the steel (thus little plastic strain is expected on specimen loading), the engineering stress has been used for the P22 tests. In Figure 2(a) the stress to cause a given strain rate or rupture time is around a factor of three higher for 316H than P22. The creep strain rates for a given stress are therefore significantly higher, and rupture times lower, for P22 steel in comparison to 316H at these temperatures. The uniaxial creep failure strain (creep ductility), ε_f^c , of P22 is over twice that of 316H based on both axial and reduction of area (ROA) measurements (see Table 1).

4.1 Isochronous Stress Strain Curves

The materials tensile data and creep laws have been combined to generate isochronous stress-strain curves as specified in Section 2.3 using the material properties given in Table 1. The resultant curves are very different for the 316H and P22 steel as shown in Figure 3(a) and (b), respectively, for times up to 100,000 hours. The P22 steel's curves are strongly time dependent even at short times. In fact the P22 curve for a time of one hour almost overlays the 316H curve at 100,000 hours. The corresponding 0.2% inelastic (creep and plastic) strain, $\sigma_{0.2}^c$, are illustrated and compared in Figure 4. Little change in $\sigma_{0.2}^c$ for 316H steel is observed for the first 1000 hrs whilst creep strains dominate in the P22 steel. Even at very short times it can be seen that for P22 $\sigma_{0.2}^c$ reduces by 50% due to creep strain accumulation in the first hour alone.

5 Creep Fracture Behaviour

5.1 CCG Behaviour

Creep crack growth test data on the compact tension specimen, C(T), from ²⁶ for P22 and ^{27, 28} for 316H steel have been re-analysed in accord with recent changes in data analysis procedures ^{2, 3}. A sizable data set is available for 316H with test durations ranging between 100 and 18,000 hours. The CCG tests on P22 steel were relatively short ranging between 300 and 4,400 hrs. The significance of plasticity on loading of the 316H steel specimens and its influence on CCG behaviour has been described in ^{9, 29, 30}. The P22 test load-up data was not available for analysis. However, it is expected that there is little plasticity on loading of P22 specimens since the ratio $\sigma_{ref}/\sigma_{0.2}$, on load up is less than 0.5 for all specimens and is on average 0.31 and 0.45 assuming plane strain and plane stress conditions, respectively. For the 316H specimens this ratio was close to or exceeded unity assuming plane strain conditions, except for the long term tests where the ratio was approximately 0.78.

Mean line power-law regression fits have been made to the CCG data as shown in Figure 5. Upper bound (UB) and lower bound (LB) fits are obtained by offsetting the mean line by ± 2 standard deviations (s.d.) of the regression fit, assuming a constant slope. Separate fits have been made to the long and short term tests of 316H in Figure 5(a) due to the distinct CCG behaviour observed in the long term tests, where plasticity is limited and thus high specimen constraint effects maintained ³⁰. The CCG rate constants for these fits (see Eqn (5)) are given in Table 2. Similar CCG behaviour

is exhibited by both steels for $C^* > 1 \times 10^{-5}$ MPamh⁻¹, however significant tails are observed for P22 though the validity criterion ² are met.

5.2 CCI Behaviour

The CCI times for 0.2 mm and 0.5 mm of crack extension are plotted against the experimentally determined C^* parameter at the corresponding time in Figure 6 and Figure 7 for 316H and P22 steel, respectively. Regression fits have been made to the data and the constants in Eqn (20) determined, accounting for the degree of data scatter (see Table 3). The CCI time data for 316H (Figure 6) appear to form reasonable correlations with the C^* parameter, though the power-law correlation exponent, ϕ_i , is significantly less than $n/(n+1)$ as predicted from CCI models ^{25, 29}. The value of ϕ_i obtained from the regression fit for P22 data for $\Delta a = 0.2$ mm is greater than unity indicating that the experimentally determined C^* parameter is not an appropriate correlating parameter in this case since, as indicated by the tails shown in Figure 5, steady state CCG conditions have not been achieved ². Also, since the data set for P22 is relatively small it can be difficult to establish the data trend. More reasonable correlating parameter are obtained for the P22 data at $\Delta a = 0.5$ mm, though again ϕ_i is significantly less than $n/(n+1)$. A significant degree of data scatter is observed for both materials, quantified by the UB/LB values of D_i , which are generally a factor of 5 greater or less than the mean value, respectively.

5.3 Creep Toughness Data, K_c^{mat}

The K_c^{mat} values at $\Delta a = 0.2$ and 0.5 mm of crack extension have been calculated using Eqn (18) for both materials in homogenous parent material (PM) and weldment (WM) conditions. A regression fit has been made to the data to obtain the values of H and j in Eqn (25) and again UB/LB fits made by offsetting the mean by ± 2 s.d. of the data set (see Table 4 and Table 5). In addition, the best line fit has been made to the data assuming a slope of $1/(2n)$ as predicted in Eqn (24). Note that sensible trendlines have not been obtained using Eqns (21) and (22) in conjunction with the CCG and CCI regression data detailed in Table 2 and Table 3, respectively, and thus are not shown in the following figures.

5.3.1 Homogenous Parent Material (PM) K_c^{mat} Data

The creep toughness data of 316H PM at 565 °C has previously been presented in ^{9, 31-33}. However, this data has been re-analysed here for consistency, and to be in accord with recent modifications in the K_c^{mat} evaluation procedure. The slope j of the regression fits to the 316H data in Figure 8, is a factor of 5 steeper than predicted using Eqn (24). The best fit line to the data using Eqn (24) provides a reasonable fit to the data for less than 500 hrs and 800 hrs for $\Delta a = 0.2$ and 0.5 mm, respectively. For times greater than these, a change in the data trend can be observed.

The corresponding K_c^{mat} data for PM P22 steel are presented in Figure 9. Creep toughness data for P22 steel at 550 °C has been presented in ¹². K_c^{mat} is expected to be insensitive to such small variations in temperature ⁸, thus the data in ¹² at 550 °C

has been combined with the data analysed here at 565 °C. The influence of temperature on K_c^{mat} is further discussed for materials of similar compositions to those analysed here in ¹⁰.

The regression fit values for $\Delta a = 0.2$ mm were non-sensible as the negative j value in Table 4 indicates an increase in K_c^{mat} with time. For $\Delta a = 0.5$ mm, however, $j \approx 1/(2n)$ as predicted in Eqn (24). For the P22 data therefore the mean lines shown in Figure 9 are the best fits of Eqn (24) to the data set, and the UB/LB factors are assumed to be equal to that obtained from a regression fit to the data.

5.3.2 Comparison of Parent and Weldment Data

The creep toughness data of 316 steel compact tension specimen weldments have been determined and compared to PM data in ¹⁶ where full details of the weldment geometry can be found. This weldment data is compared to additional and re-analysed PM data in Figure 10. Due to the limited amount of weldment data available the slope of the weldment trend line fitted to the data has been assumed to be equal to that of the regression fit to the PM. A relatively good fit is observed, the coefficients for which are given in Table 4. For a given time, the mean K_c^{mat} value of the 316 weldment specimens is a factor of approximately 2 less than that for a PM.

Creep toughness data from C(T) weldment specimens of P22 steel are compared to PM specimens at 565 °C and 550 °C for both $\Delta a = 0.2$ and 0.5 mm in Figure 11(a) and (b), respectively. The PM and weldment data at both temperatures fall within the same scatter bands. A best fit line assuming a slope $j = 1/(2n)$, as in Eqn (24), has been made to the PM and weldment data sets, where n has been taken to be equal to the PM value. This assumption is justified by the fact that the value of n determined from tests performed on cross-weld uniaxial specimens of P22 at 550 °C, which are considered to most represent the behaviour of C(T) weldments, were found equal to the PM test values ²⁶. Note that sensible regression fits were not obtained for this data set. On average there is a 10% difference between the PM and weldment value of K_c^{mat} for a given time and bound, whereas there is around a 40% and 80% difference between the UB or LB and mean value of K_c^{mat} for a given time, for the PM and weldment data, respectively, as can be deduced from Table 5. Therefore, within the extent of data scatter, little difference in creep toughness behaviour is observed between the two material conditions, for the P22 steel.

6 TDFAD Formation

The TDFADs for a range of times ($t = 0, 1, 10, 100, 1000, 10000$ and 100000 hrs) have been determined from Eqn (10),(11), (14)-(17), using the PM data given in Table 1, and are shown in Figure 12. Also shown in Figure 12 for comparison purposes is the R6 Option 1 curve (see Eqn (3)). Note that the R6 cut off point, L_r^{max} (see Eqn (5)) is the appropriate value to use (see Section 2.1) for times less than 10,000 hrs for the 316H material and for all times in the case of P22.

As can be seen in Figure 12(a) and described in ⁹ the 316H TDFAD is insensitive to time and deviates little from the R6 option 1 curve. The P22 TDFAD however, is in

relatively good agreement with the R6 Option 1 curve at time 0 hrs, but almost instantly deviates significantly from it. These trends are due to the relatively low and very high creep strain rates experienced in 316H and P22, respectively, (see Figure 2(a) and Figure 3) at these temperatures.

7 CCI Time Predictions

The TDFAD procedure for predicting CCI times is evaluated for both steels. A program has been developed that determines the intersection of a point (K_r , L_r) for a given time and crack extension with the TDFAD of corresponding time. Predictions for both PM and weldment data have been determined, though note the TDFADs have strictly been derived from PM data. The CCI times have been predicted for the tests on homogenous PM and weldment C(T) specimens^{16, 21, 27}. These tests have been used to derive the relationship of creep toughness on time, thus appropriate predictions are expected. In addition, CCI data are available for feature tests on P22 PM and weldment pipe components, as detailed in³⁴. These results are also predicted here using the TDFAD procedure. Predictions on C(T) specimens have been based on the plane strain reference stress solution. Details of the stress intensity factor and reference stress solution for the pipe component are given in³⁴.

The TDFADs predicted CCI time, $t_i^{Predicted}$, is compared to the experimentally determined value, t_i^{Exp} , in Figure 13-Figure 15. Included in these figures is the line for $t_i^{Predicted} = t_i^{Exp}$. Points situated below this line indicate a conservative prediction, and vice-versa. Predictions have been obtained using the mean and LB values of K_c^{mat} calculated using Eqn (25) with the regression fit values shown in Table 4 for 316H and the best fit values with an assumed slope of $j = 1/(2n)$ (see Eqn (24)) as shown in Table 5 for the P22 material. The predictions based on mean or LB K_c^{mat} values are shown as open and grey symbols, respectively.

The weldment and PM specimen predictions are shown together in Figure 13 for the 316 steel. Similar trends are observed for the two material conditions at both $\Delta a = 0.2$ and 0.5 mm. Predictions based on mean K_c^{mat} values are often, but not always, conservative whereas the use of LB K_c^{mat} values consistently gives conservative predictions. The same is true for C(T) specimen PM and weldment data for P22 material shown in Figure 14 and Figure 15, respectively, except for one point in Figure 14(b) and Figure 15(a) which is just above the line using a LB K_c^{mat} value. However, since the data set available for P22 is limited, less confidence can be assigned to the appropriateness of the TDFAD for predicting CCI for P22, especially for long term conditions since long term data are unavailable for this material.

An initiation distance $\Delta a = 0.5$ mm has been deemed suitable for the feature components³⁴, thus pipe data are not included in Figure 14(a) and Figure 15(a). Predictions for the pipe components are found to be very conservative in all cases with $t_i^{Predicted}$ predicted being less than 5% of t_i^{Exp} . The degree of uncertainty associated with t_i^{Exp} measurements and due to the approximations made in evaluating the parameters K and σ_{ref} for these components leads to higher

uncertainties in their CCI time predictions compared to that for laboratory test specimens. These additional uncertainties may contribute to the excessive conservatism in the CCI times predicted for these components. This uncertainty is however not easily quantified. The influence of the reference stress solution on the CCI predictions has previously been examined and discussed in ⁹.

8 Discussion

The appropriate isochronous stress-strain response of a location being either within the parent, weld or HAZ of a weldment will be affected by the isochronous stress-strain response of the surrounding material region. An equivalent isochronous stress-strain response may be defined, with intermediate isochronous properties, which will be a function of weldment geometry, loading, crack geometry, size and location. Note that for the weldment tests considered in this work the crack was located in the HAZ adjacent to the HAZ/PM interface. The equivalent isochronous stress-strain curve is based on the mis-match limit load and equivalent material stress-strain curve described in R6 ¹⁷. A conservative assessment is however obtained using the tensile properties of the weaker material, which would be the 316H parent material in the case of the 316 specimen weldments which consist of 316H PM and 316L weld metal ¹⁶. The mis-match reference stress for an overmatched weld is expected to be greater than that for a homogeneous specimen (see e.g. ³⁵). Thus, should a conservative prediction be obtained for an overmatched weldment using the materials PM properties and a homogeneous reference stress solution then the use of the mismatch reference stress would further increase the degree of conservatism. Tensile test data are not available for the P22 weld material at 565 °C however results in ³⁶ from a similar weld tested at 550 °C indicates that the welds yield strength is slightly (7%) less than the PM, however their general stress-strain behaviour of the weld is very similar to that of the PM. Therefore, it is considered that mismatch effects may not be significant for the P22 weldments analysis. The influence of mismatch on the isochronous stress-strain curves of Cr-Mo steel, following the R6 approach, have been examined in ¹⁵ where the influence of geometry, crack size and time on the TDFAD was found negligible. A simplified procedure for evaluating isochronous stress-strain curves has been proposed in ¹³ (neglecting plasticity and assuming the elastic response of the entire weldment is equal to the PM's response) and applied to a 1Cr0.5Mo steel weldment. A dependency of the TDFAD on time was shown in ¹³, however, the deviation of the weldments TDFAC from the PM is curve was not demonstrated.

Since the TDFAD is being used to predict CCI it is expected that creep fracture would be the failure mode predicted. This is true for all cases examined for the 316H PM and weldments and the majority of cases for the P22 PM and weldment analyses using the lower bound K_c^{mat} , which leads to higher K_r values. However, for the vast majority of cases failure by plastic collapse is predicted for the P22 PM and weldment analyses when the mean K_c^{mat} values are employed, which is considered unrealistic. This is due to the strong time dependence of $\sigma_{0.2}^c$ causing a rapid increase of L_r to

values greater than $L_r^{\max}$. The time dependency of the rupture stress σ_r is small in relation to $\sigma_{0.2}^c$, thus $L_r^{\max}$ calculated from Eqn (8), indicating creep rupture, does not fall below that of Eqn (5), indicating plastic collapse.

There are considerable differences between the TDFADs for both steels. It has been demonstrated here and previously elsewhere (see e.g. ⁸), that due to the insensitivity of the 316H steels TDFAC to time a single curve for a given time or even the R6 Option 1 curve may suffice. However, the significant time dependency of the P22 steel shown here demonstrates that this is not the case for this ferritic steel. This was also noted for the similar steel examined in ³⁷. An influence of tensile curve fitting on the accuracy of the results was noted in ⁵. The Ramberg-Osgood material model widely employed to describe a materials tensile response is known to cause inaccuracies especially for high strain hardening austenitic steels ³⁸. In this work however such issues are negated by employing the materials experimentally measured tensile test data.

The K_c^{mat} estimates obtained from using CCI and CCG data (see Eqn (21) and (22)) have been unsuccessful for the cases examined here, though they have provided satisfactory results elsewhere ^{8, 10}. The best fit line from Eqn (24) has provided a reasonable fit to all the P22 K_c^{mat} vs. time data and to the short term 316H data, but can not describe the longer term 316H steel data. The apparent change in slope between the long term and short term K_c^{mat} vs. time data may indicate a change in material fracture behaviour, as suggested in ³¹. However, within the extent of data scatter no firm conclusions can be made.

The influence of using the LB bound or mean K_c^{mat} value has been examined here, and significant differences are obtained in both cases, including a change of predicted failure mode for the P22 steel. The influence of the reference stress solution on the CCI predictions has previously been examined in ⁹. The plane stress reference stress is greater than the plane strain reference stress and therefore leads to higher degrees of conservatism. Using the plane stress reference stress solution with a LB K_c^{mat} value will therefore lead to conservative predictions for all cases considered here.

Long term CCI and CCG tests are more representative of component operating conditions. The difference between the CCG rate and K_c^{mat} trends for long and short term data signify that more long term tests are required to verify the TDFAD prediction method for use in high temperature components.

9 Conclusions

The creep toughness parameter, K_c^{mat} , for a crack extension of 0.2 mm and 0.5 mm has been determined for an austenitic 316H steel at 550 °C and ferritic P22 steel at 565 °C from creep crack growth (CCG) tests on the compact tension, C(T), specimen geometry. Both homogenous parent material (PM) and weldment specimen tests have been considered. K_c^{mat} shows a relatively weak dependency on time for the P22 steel, however a clear decrease in K_c^{mat} with time is observed for

the 316H material, especially at long times. Significant data scatter has been observed in the K_c^{mat} data for both materials, which has been quantified. For the P22 steel the influence of K_c^{mat} on time had to be predicted using models since, due to the extent of data scatter within the relatively small data set, a sensible regression line fit could not be achieved. For a given time, the mean K_c^{mat} values of the 316 weldment specimens are found to be a factor of 2 less than that of the PM specimens. Within the extent of data scatter, little difference in creep toughness behaviour is observed between the two material conditions for the P22 steel. Isochronous stress-strain curves and time dependent failure assessment diagrams (TDFAD) have been shown for a range of times for both the 316H and P22 PM steels. Due to the relatively low creep strain rates in the 316H steel at 550 °C, these curves are relatively insensitive to time and little change is observed in the 316H materials 0.2% creep strength $\sigma_{0.2}^c$ for times less than 10,000 hrs. On the contrary the P22 curves rapidly deviate from their tensile curves (at $t = 0$ hrs) and $\sigma_{0.2}^c$ halves within the first hour of creep. A program has been developed to obtain explicit creep crack Initiation (CCI) time predictions using the TDFAD method. CCI time predictions have been obtained for the CGG tests on the PM and weldment C(T) specimens and in addition for CCG data on PM and welded P22 pipe feature tests using the TDFAD derived from PM data. Reasonable predictions are generally obtained using the mean bound value of K_c^{mat} , though not always conservative assuming plane strain conditions. Conservative predictions are almost always obtained when LB K_c^{mat} values are employed and may be ascertained by assuming plane stress conditions. Although considerable differences are observed in the form of the K_c^{mat} data and TDFADs of both austenitic 316H steel at 550 °C and ferritic P22 steel at 565 °C reasonable and conservative CCI time predictions can be achieved for both materials following the TDFAD approach. It is however essential that the influence of time on the TDFAD is considered for the P22 steel. Further long term tests are recommended for the validation of the TDFAD method to predict the CCI times of in-service components.

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11 Tables

Table 1: Material properties for homogeneous parent material 316H stainless steel at 550 °C and P22 steel at 565 °C

	316H	P22 (PM)
$A \text{ (MPa}^{1/n} \text{ h}^{-1})$	1.47×10^{-34}	3.21×10^{-27}
n	11.58	10.68
n_p	7.45	2.47
$A_p \left(\text{MPa}^{-n_p} t^{-p} \right)$	2.60×10^{-23}	1.0×10^{-8}
p	0.746	0.3
$E \text{ (MPa)}$	140,000	140,135
$\sigma_{0.2} \text{ (MPa)}$	170	284
$\sigma_{UTS} \text{ (MPa)}$	442	365
$B_r \text{ (MPa}^{v_r} \text{ h)}$	5.27×10^{31}	1.43×10^{22}
v_r	11.3	9.11
$\sigma_{flow} \text{ (MPa)}$	306	327
$\varepsilon_f \text{ (%) (Eng)}$	37	27
$\varepsilon_f^c \text{ (%) (Axial)}$	8	31
$\varepsilon_f^c \text{ (%) (ROA)}$	21	65

Table 2: Regression fit constants to CCG rate vs. C^* data (Eqn (19)) from long term (LT) and short term (ST) tests on 316H and tests on P22 steel.

	316H (LT)	316H (ST)	P22
D Mean	9.25	3.45	4.52
D UB	16.91	8.26	16.87
D LB	5.06	1.44	1.21
ϕ	0.73	0.75	0.75

Table 3: Regression fit constants to CCI time *vs.* C^* data (Eqn (20)) from tests on homogenous parent material 316H and P22 steel for crack extensions, Δa , of 0.2 and 0.5 mm.

	316H		P22	
Δa (mm)	0.2	0.5	0.2	0.5
D_i Mean	0.7	3.9	9.3×10^{-4}	22.1
D_i UB	5.0	17.9	4.0×10^{-3}	4.5
D_i LB	0.1	0.8	2.2×10^{-4}	0.93
ϕ_i	0.55	0.47	1.1	0.44

Table 4: Mean, upper bound (UB) and lower bound (LB) regression fit constants to K_c^{mat} vs. time data (Eqn (25)) of homogeneous parent material (PM) and weldment 316H and P22 PM specimens.

	316H		316H		P22	
	PM		Weldment		PM	
Δa (mm)	0.2	0.5	0.2	0.5	0.2	0.5
H Mean	242.4	224.5	117.8	173.0	29.8	73.3
H UB	449.7	611.5	211.8	289.0	51.5	136.8
H LB	130.6	224.5	65.5	103.5	17.3	39.3
j	0.20	0.23	0.20	0.23	-0.06	0.04

Table 5: Mean, upper bound (UB) and lower bound (LB) constants for the K_c^{mat} v.s. time relationship of Eqn (25) for homogenous parent material (PM) specimens of 316H and of P22 steel, and weldment specimens of P22 steel assuming $j = 1/(2n)$

	316H		P22			
	PM		PM		Weldment	
Δa (mm)	0.2	0.5	0.2	0.5	0.2	0.5
H Mean	119.8	135.1	51.35	71.96	47.2	62.1
H UB	222.2	223.0	88.68	134.3	69.5	88.3
H LB	64.5	81.9	29.73	38.56	32.0	43.7
$j = 1/(2n)$	0.043		0.047		0.047	

12 Figures

Figure 1: Comparison of the high temperature tensile behaviour of 316H stainless steel at 550 °C and ferritic P22 steel at 565 °C.

Figure 2: Comparison of (a) secondary creep strain rates and (b) creep rupture time with stress for 316H at 550 °C and P22 at 565 °C.

Figure 3: The isochronous stress strain curves at a range of times for (a) 316H stainless steel at 550 °C and (b) ferritic P22 steel at 565 °C.

Figure 4: Evolution of the 0.2% creep strength, $\sigma_{0.2}^c$, with time of the 316H stainless steel at 550 °C and P22 steel at 565 °C.

Figure 5: CCG rate correlations with the C^* parameter for (a) 316H stainless steel at 550 °C and (b) P22 steel at 565 °C including data bounds.

Figure 6: CCI time correlations with the C^* parameter at (a) $\Delta a = 0.2$ mm and (b) $\Delta a = 0.5$ mm for 316H stainless steel at 550 °C, including data bounds.

Figure 7: CCI time correlation with the C^* parameter at (a) $\Delta a = 0.2$ mm and (b) $\Delta a = 0.5$ mm for P22 steel at 565 °C, including data bounds.

Figure 8: Creep toughness vs. time data for 316H stainless steel at 550 °C at (a) $\Delta a = 0.2$ mm and (b) $\Delta a = 0.5$ mm, including mean and upper/lower bound regression fits and the best fit using Eqn (24).

Figure 9: Creep toughness *vs.* time data for ferritic P22 steel PM at (a) 0.2 mm (b) 0.5 mm, including mean, upper/lower bound fitted lines.

Figure 10: Comparison of the creep toughness vs. time data for 316H stainless steel at (a) 0.2 mm (b) 0.5 mm, for homogenous PM and weldment C(T) specimens.

Figure 11: Comparison of the creep toughness vs. time data for P22 steel at 565 °C at (a) 0.2 mm (b) 0.5 mm, for homogenous PM and weldment C(T) specimens.

Figure 12: Comparison of the TDFADs at various times with the R6 Option1 curve for homogenous PM (a) 316H stainless steel at 550 °C and (b) P22 steel at 565 °C.

Figure 13: Comparison of the predicted and experimentally determined CCI times from tests on homogenous PM and weldment C(T) specimens of 316 steel at 550 °C for (a) 0.2 mm and (b) 0.5 mm, using mean and lower bound K_c^{mat} values.

Figure 14: Comparison of the predicted and experimentally determined CCI times from tests on homogenous PM C(T) specimens and pipe components of P22 steel at 565 °C for (a) 0.2 mm and (b) 0.5 mm, using mean and lower bound K_c^{mat} values.

Figure 15: Comparison of the predicted and experimentally determined CCI times from tests on C(T) weldment specimens and welded pipe components of P22 steel at 565 °C for (a) 0.2 mm and (b) 0.5 mm, using mean and lower bound K_c^{mat} values.