

The Fracture Mechanics Concept of Creep and Creep/Fatigue Crack Growth in Life Assessment

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Abstract: There is an increasing need to assess the service life of components containing defect which operate at high temperature. This paper describes the current fracture mechanics concepts that are employed to predict cracking of engineering materials at high temperatures under static and cyclic loading. The relationship between these concepts and those of high temperature life assessment methods is also discussed. A model for predicting creep crack growth initiation and growth in terms of C^* and the creep uniaxial ductility is presented and it is shown that this model gives good agreement with the experimental results. The effects of cyclic loading on crack growth behaviour are considered and fractography evidence is shown to back a simple cumulative damage concept when dealing with creep/fatigue interaction. Finally a discussion is presented which highlights the important aspect of life assessment methodology for high temperature plant.

Keywords: Fracture Mechanics, C^* parameter, uniaxial ductility, fractography, cumulative damage

1. Introduction

In both power generation plants and the chemical industries there is a need to assess the significance of defects which may exist in high temperature equipment operating in the creep range. Analysis of defects occurring under high temperature conditions using fracture mechanics methods has become an important life assessment tool for industries which have components operating in creep range. Engineering life assessment and component design utilise models based on theoretical principles which always need to be validated under practical and operational circumstances. This need arises at the design stage for setting acceptance level in association with the sensitivity of non-destructive inspection equipment and also during use for making residual lifetime and structural integrity prediction.

The mechanism of time dependent deformation [1-2] is shown to be analogous to deformation due to plasticity. Therefore elasto-plastic fracture mechanics method can be linked to high temperature fracture mechanics parameters. Techniques are shown for determining the creep fracture mechanics parameter C^* using experimental crack growth data, collapse loads and reference stress methods. Models for predicting creep

crack growth interims of C^* and creep uniaxial ductility are developed. Cumulative damage concept is used for predicting crack growth under static and cyclic loading conditions.

A number of processes dominate the creep processes [2-4] as shown in Figure 1. When secondary creep dominates, it is often possible to express secondary creep strain rate $\dot{\epsilon}_s$, in the form

$$\dot{\epsilon}_s = A \sigma^n \exp(-Q/RT) \quad (1)$$

where n and Q are material dependent parameters and R is Boltzmann's

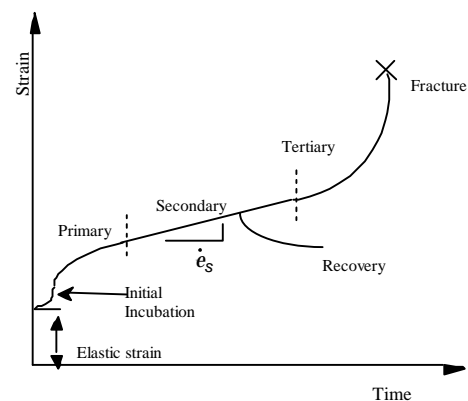


Fig. 1. Various Regions in a Creep Curve

constant. The values of n and the activation energy Q

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are sensitive to the processes controlling creep. Creep in polycrystalline materials is sensitive to grain size, alloying additions, initial condition of the material, heat treatment and testing conditions. Figure 2 shows Schematic view of the regimes of fracture in terms of geometric constraint and metallurgical factors ranging from elastic to fully ductile behaviour [2] and Figure 3 compares the creep properties of a range of engineering alloys. A creep fracture can be transgranular or intergranular [5-7] There is a general trend towards transgranular failures at short creep lives and relatively low temperatures and intergranular failures at long lifetimes and higher temperatures.

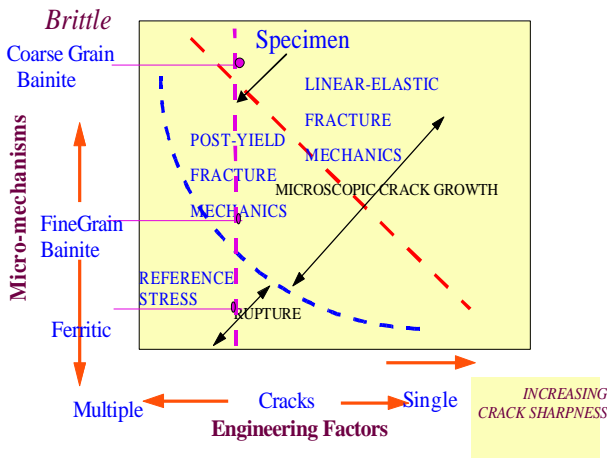


Fig. 2. Schematic View of the Regimes of Fracture in Terms of Geometric Constraint and Metallurgical Factors Ranging from Elastic to Fully Ductile Behaviour [2].

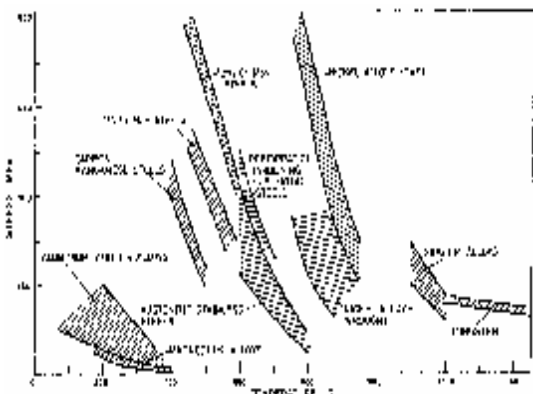


Fig. 3. Stress to Produce Creep Rupture in 100 Hours [5]

Intergranular failures are usually most relevant to practical operating situations. They can result in creep ductilities, which are much less than room temperature ductilities.

2. High Temperature Fracture Mechanics

The first analytical study of fracture mechanics was made in the 1920s [8] for cracks in an ideal brittle material. This study was based on the consideration of energy balance in a cracked body. The stress intensity factor K was proposed, as a parameter that describes the stress state around a crack tip, by Irwin [8]. The energy release rate G which corresponds uniquely to stress intensity factor K was also proposed. In ductile materials with low yield stress or in creeping materials the (Linear Elastic Fracture Mechanics) LEFM approach is not sufficient in predicting fracture of a cracked structure.

The J contour integral [9] was proposed as a fracture mechanics parameter which has been widely accepted as an effective parameter for the assessment of ductile fracture for large scale yielding cases [2].

In general the sharper the crack and the larger the grains the more brittle the material behaviour and the more likely that LEFM will be the relevant correlating parameter.

Conversely the more creep ductile the material and the more diffuse the number of cracks or damage region the likelihood that (Non-Linear Fracture Mechanics) NLEFM J at room temperature and C^* at high temperature will describe the crack tip region [10,11]. The assumption made is that creep deformation is time dependent.

The overriding assumption is that the local crack tip stress singularity will dominate and control the development of the crack under creep.

The crack tip singularity, over time, will therefore decrease with increasing ductility in creep producing the same deformation contour as in plasticity.

In the extreme the stress at diffused crack tip or damage region, where it can be assumed that no stress singularity exists, is equivalent to the remote and the analysis would employ net-section stress, ligament collapse load and reference stress S_{ref} concepts to predict failure lives. These areas are schematically shown in Figure 4.

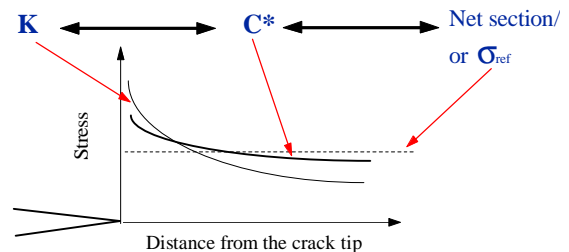


Fig. 4. Schematic Profile of the Stress Field at the Crack Tip and the Relevant fracture Mechanics Parameters Associated with them.

3. Creep Fracture Mechanics (C* integral)

For many materials, creep deformation can be considered to be composed of three regimes, namely primary, secondary and tertiary creep regimes. The secondary creep strain rate can be represented by a power law,

$$\dot{\epsilon}_s = A_s \mathcal{S}^{n_s} \quad (2)$$

where A_s , and n_s are material constants. The use of an average creep rate obtained directly from creep rupture data has

$$\dot{\epsilon}_A = \frac{e_f}{t_r} = \dot{\epsilon}_o \left(\frac{\mathcal{S}}{\mathcal{S}_o} \right)^{n_A} = A_A \mathcal{S}^{n_A} \quad (3)$$

been proposed to account for all three stages of creep [2,10,11]. where e_f is the uniaxial failure strain, t_r is the time to rupture and \mathcal{S} is the applied stress. The variables $\dot{\epsilon}_o$, \mathcal{S}_o , A_A and n_A in Equation (3) are generally taken as material constants. These equations are used to characterise the steady state (secondary) creep stage where the hardening by dislocation interaction is balanced by recovery processes. The typical value for n is between 3 and 10 for most metals. The definition of the C^* integral is obtained by substituting strain rate and displacement rate for strain and

$$C^* = \int_{\Gamma} [W_s^* dy - T_i (d\dot{\epsilon}_i / d_x) ds] \quad (4)$$

displacement of the J integral [2,10,11,12].

$$W_s^* = \int_o^{\dot{\epsilon}_i} s_{ij} d\dot{\epsilon}_{ij} \quad (5)$$

where W^* is strain energy density change rate, where $(\dot{\epsilon}_i = du_i / dt)$ is displacement rate. The asymptotic stress and strain fields are expressed by equations (6) and (7) [2, 10-12].

$$s_{ij} = s_o \left(\frac{C^*}{I_n s_o \dot{\epsilon}_o r} \right)^{1/(n+1)} \tilde{s}_{ij}(q, n) \quad (6)$$

$$\dot{\epsilon}_{ij} = \dot{\epsilon}_o \left(\frac{C^*}{I_n s_o \dot{\epsilon}_o r} \right)^{n/(n+1)} \tilde{e}_{ij}(q, n) \quad (7)$$

Therefore the stress and strain rate fields of non-linear viscous materials are also HRR type fields with \tilde{s}_{ij} and \tilde{e}_{ij} equivalent stress and strain and I_n is the normalizing factor which depends on n and the state of stress.

4. Experimental Estimate

From the view of an energy balance, the C^* integral is the rate of change of potential energy with crack extension. For any geometry the dimensionalised equation becomes

$$C^* = \frac{n}{n+1} f(a/W) \frac{P \dot{\Delta}}{B_n W} = F \frac{P \dot{\Delta}}{B_n W} \quad (8)$$

where P is the load, $\dot{\Delta}$ is the displacement rate due to creep, B_n is the net thickness between side grooves, and W is the width of the specimen. The $f(a/W)$ is a non-dimensional geometry function of a crack length a [10-12]. For CT specimens, the $f(a/W)$ is given by analogy with fracture toughness J_{IC} testing in ASTM E813 [12,15].

$$f(a/W) = \frac{2 + 0.552(1 - a/W)}{1 - a/W} \quad (9)$$

Reference Stress Method

A reference stress method for C^* calculation is also easily obtained by

$$C^* = m s_{ref} \dot{\epsilon}_{ref} \left(\frac{K}{s_{ref}} \right)^2 \quad (10)$$

The m is the normalizing factor and K is stress intensity factor. This method is predominantly used in the codes of practice to determine C^* [13, 14] in components.

5. Crack Growth at Elevated Temperature

At elevated temperatures where creep is dominant, time-dependent crack growth is observed, creep crack growth rate \dot{a} ($= da/dt$) must be estimated using appropriate parameters. Several fracture parameters have been applied for this purpose. The most commonly used parameters are stress intensity factor K , the C^* integral and reference stress (net section stress) s_{ref} . Crack growth rate \dot{a} is usually given as follows using these parameters;

$$\dot{a} = AK^m \quad (11)$$

$$\dot{a} = DC^*{}^f \quad (12)$$

$$\dot{a} = HS_{ref}^p \quad (13)$$

where A , D , H , m , f and p are material constants which may depend on temperature and stress state. A suitable parameter to describe crack growth at elevated temperature will depend on material properties, loading

condition, and time when crack growth is observed [10,11]. Modeling of equation (12) can be performed using uniaxial creep properties of material and creep process zone ahead of crack. Nikbin et al. [10,11] used equations (2) to (8) and (12) to develop this model.

$$\dot{a}_s = \frac{(n+1)\dot{\epsilon}_o}{e_f^*} \left[\frac{C^*}{I_n S_o \dot{\epsilon}_o} \right]^{n/(n+1)} \tilde{\epsilon}_{ij} \left[r_c^{1/(n+1)} \right] \quad (14)$$

The non dimensional function $\tilde{\epsilon}_{ij}$ is normalised so that its maximum equivalent becomes unity. Hence, assuming this maximum value, the constants D and f in equation (12) become;

$$D = \frac{(n+1)\dot{\epsilon}_o}{e_f^*} \left[\frac{1}{I_n S_o \dot{\epsilon}_o} \right]^{n/n+1} \quad (15)$$

$$\text{where } f = \frac{n}{n+1}$$

In this equation I_n is the normalizing factor which depends on n and the state of stress, r_c is the size of the creep process zone and e_{fo}^* is equivalent to creep ductility considering constraint effect.

For plane stress conditions e_{fo}^* can be taken as e_{fo} and for plain strain conditions as $e_{fo}/50$.

For most steels, $n > 1$, usually 5 to 10, so that equation (14) is relatively insensitive to the value of r_c . Considering most engineering materials, equation (14) has been simplified [10,11].

$$\dot{a}_s = \frac{3C^* \dot{\epsilon}_o^{0.85}}{e_f^*} \quad (16)$$

where \dot{a}_s is in mm/hour, e_f^* is strain as a fraction and C^* is in MJ/m²h.

6. Fatigue Crack Growth

Fatigue crack growth is usually observed as transgranular cracking at low temperature.

At elevated temperature, transgranular cracks are also observed under relatively high frequency cycles and this fatigue crack growth rate can still be characterised by elastic or elasto-plastic fracture mechanics parameters in most cases [17-19].

The crack growth rate da/dN is correlated with DK using the power law relation [17] in steady state fatigue as,

$$\frac{da}{dN} = C_f \Delta K^m \quad R = \frac{S_{\min}}{S_{\max}} \quad (17)$$

where C_f and m are material constants and m is typically around 3. Typically, da/dN is sensitive to the mean stress or the load ratio R defined by; S_{\min} and S_{\max} are the minimum stress and the maximum stress, respectively [17]. As temperature is increased, time dependent processes become more significant. Creep and environmentally assisted crack growth can take place more readily since they are aided by diffusion and rates of diffusion increase with rise in temperature. Generally the influence of frequency on crack propagation rate is more pronounced with increase in temperature and R [17-19]. Figure 5 shows a schematic description of cracking rate versus DK for cyclic cracking at elevated temperatures showing the effects of frequency, R-ratio and temperature.

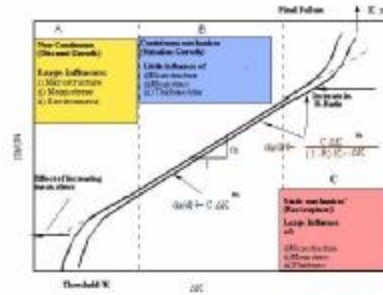


Fig. 5. Cyclic Crack Growth at Elevated Temperatures

7. Creep-Fatigue Crack Growth

Figure 6 shows example of operation scheme of a power plant the power plants have to change their operating temperature and pressure to follow the demands of electricity need and to shut down and restart for their routine maintenance, therefore, interaction between creep and fatigue is expected under cyclic loading. Some of the causes of creep-fatigue interaction might be the enhancement of fatigue crack growth due to embrittlement of grain boundaries or weakening of the matrix in grains and enhancement of creep crack growth due to acceleration of precipitation or cavitation by cyclic loading [20-22]. Nevertheless, it has been proposed that a simple linear summation rule for creep-fatigue crack growth can be applied to predict the crack growth in several engineering metals.

$$\frac{da}{dN} = \left(\frac{da}{dN} \right)_{\text{creep}} + \left(\frac{da}{dN} \right)_{\text{fatigue}} = \frac{1}{3600f} \left(\frac{da}{dt} \right)_{\text{creep}} + \left(\frac{da}{dN} \right)_{\text{fatigue}} \quad (18)$$

or

$$\frac{da}{dt} = \left(\frac{da}{dt} \right)_{\text{creep}} + \left(\frac{da}{dt} \right)_{\text{fatigue}} = \left(\frac{da}{dt} \right)_{\text{creep}} + 3600f \left(\frac{da}{dN} \right)_{\text{fatigue}} \quad (19)$$

where da/dN is crack growth per cycle (mm/cycle), da/dt is crack growth rate in mm/hour, and f is

frequency in Hz.

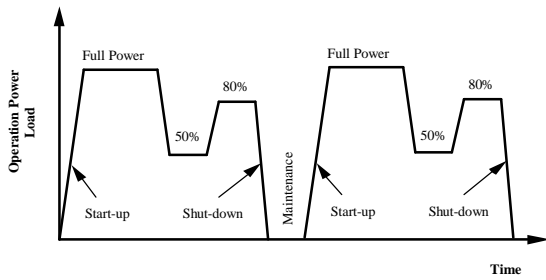


Fig. 6. Example of Operation Scheme of a Power Plant

8. Experimental Results and Discussion

Figures 7 and 8 shows, Predictions of crack growth rate versus C^* for API nickel-base superalloy tested at 700 °C, and 316 H stainless steel using the steady state model plain stress and strain condition, as it can be seen that model bounds the steady state crack growth rates.

Also the effects of geometry and specimen size on creep crack growth for the API alloy and 316 H stainless steel are shown in figure 7 and 8. It is clear that cracking rate is state of stress controlled in the creep range. Effects of the influence of frequency on crack growth/cycle in a nickel base alloy (API) at 700°C at $R = 0.7$ is shown in figure 9. In the steady state cracking region crack growth can be described by the Paris law (equation (17)) with $m \approx 2.5$.

This value is within the range expected for room temperature behaviour. Correlation of results obtained at 20 $\text{Mpa}\sqrt{\text{m}}$ for specimen thickness ($B=25\text{mm}$) is depicted in figure 9. Furthermore in figure 9 the two lines show the relative behaviour of the creep and the fatigue components of da/dN .

The slope of -1 indicates time dependent creep cracking and the horizontal line indicates fatigue control of da/dN . Therefore by adding the two components it is clear that at high frequency creep is seen to have third order effect on cracking rate and conversely at low frequencies fatigue has a third order effect.

From metallurgical and fractographic investigations performed on the alloy tested in the creep and creep fatigue range similar qualitative conclusions can be reached with respect to the mode of the creep fatigue interaction.

Figure 10 show the fractographs for the nickel-base superalloy API tested at 700 °C. There is a transition from intergranular cracking at $f=0.001$ Hz to transgranular cracking at 10 Hz. The intermediate frequencies show a mixture of intergranular and transgranular cracking modes.

These suggest that the two mechanism work in parallel and that cumulative damage concepts proposed above can well describe the total cracking behaviour. In this

region crack growth can be described by equation (18). Figure 12 shows a fractograph exhibiting the effects of creep and fatigue on the actual cracking of the CCT. In this case the frequency was varied between 10Hz and 0.001 Hz.

Where fatigue dominates the crack front profile is approximately a quadrant and the cracking rate is the same in the inner section and the surface. When frequency is reduced the crack leads in the centre and where the surface is more in plane stress crack growth rate reduces. The effect is reversed when the frequency is increased once again.

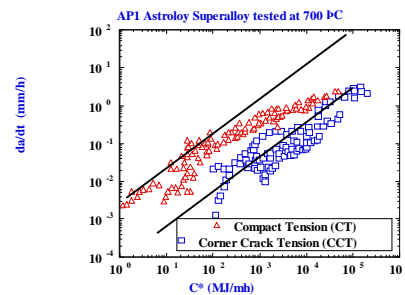


Fig. 7. Predictions of crack Growth rate Versus C^* for API Nickel-base Superalloy tested at 700 °C using the Steady State Model.

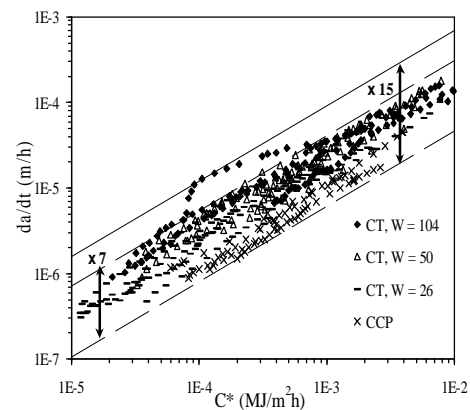


Fig. 8. Steady State Crack Growth Rate Versus C^* for 316 H Stainless steel. CT = Compact Tension and CCT = Centre Cracked Tension or M(T) Specimen.

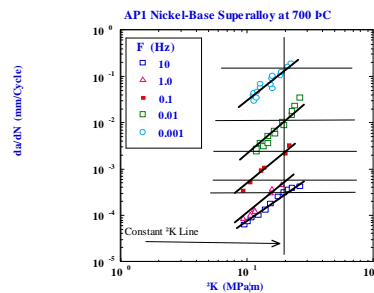


Fig. 9. Frequency Dependence of Fatigue Cracking at high Temperatures for API Nickel-base Superalloy Tested at 700 °C and $R=0.7$

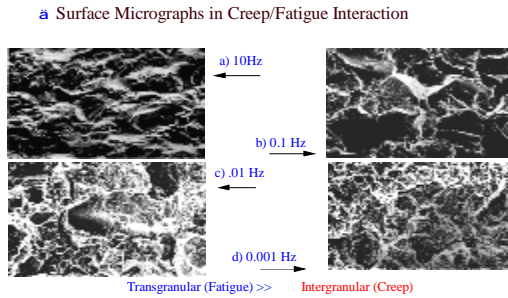


Fig.10. Effects of Frequency on Mode of Failure for API Astroloy Nickel-base Superalloy Tested at 700 °C.

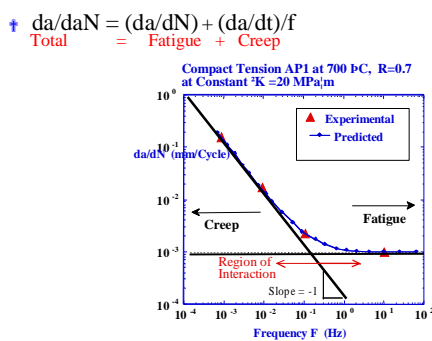


Fig. 11. Fatigue Crack Growth Sensitivity to Frequency in an API Nickel-base Superalloy Tested at 700 °C.

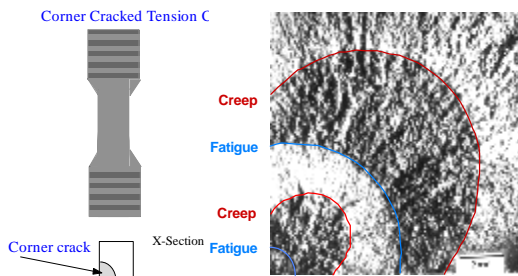


Fig. 12. (a) Corner Crack Tension Specimen API Superalloy Tested at 700 °C (b) and a fractograph Showing Influence of Creep and Fatigue on the Shape and Texture of the Crack Front.

9. Relevance to life Assessment Methodology

For new plant design procedure are required to avoid excessive creep deformation and fracture in electric power generation equipment, aircraft gas turbine engine, chemical process plant, supersonic transport and space vehicle applications.

By extending plant life and to avoid premature retirement on the basis of reaching the design life which may have been obtained previously using conservative procedures is an economic reality for most

industrial operators. Many electric power generating and chemical process plants are designed to last for 30 years or more assuming specific operating conditions. It is possible if they are used under less severe conditions to those anticipated or a re-evaluation using the modern understanding of creep and fracture is performed it may justify a further extension of the plant life.

Most high temperature design codes have been developed from those that have been produced for room temperature applications. The high temperature design codes in current use have been developed principally for application to defect free equipment. Flaws can be detected using non-destructive examinations; consequently there is a requirement for establishing tolerable defect size. Also equipment can experience a wide range of types of loading, for example, stress and temperature can be cycled in-phase or out-of-phase, a stress can be applied or removed at different rate and dwell periods can be introduced during which stress or strain are maintained constant. In these circumstances, it is important that the processes governing failure are properly understood so that the appropriate predictive models and equations can be employed.

10. Conclusion

It has been shown that high-temperature crack growth occurs by either cyclic-controlled or time-dependent processes. Over the limited range where both mechanisms are significant, a simple cumulative damage law can be employed to predict behaviour. Interpretations have been developed in terms of linear elastic and non linear fracture mechanics concepts. Linear elastic fracture mechanics descriptions are expected to be adequate when fatigue and environmental processes dominate. When creep mechanisms control stress redistribution takes place in the vicinity of the crack tip and use of the creep fracture mechanics parameter C^* should be employed for characterising the creep component of cracking.

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