Characterizing high temperature crack growth behaviour under mixed environmental, creep and fatigue conditions

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Abstract:

Components in high temperature plant could undergo failure due to combinations of fatigue, creep or oxidation/ corrosion depending on the loading, temperature and environmental conditions. A novel and robust approach for a progressive failure modelling is presented in this paper which for the first time attempts to combine these failure mechanisms as time or cycle dependent processes. In this study, a combined multiaxial inter/transgranular crack growth model at the meso-scale level was proposed to conveniently deal with the various failure scenarios that may exist in plant components. The simulated crack under the combinations of time dependent creep and oxidation mainly propagated along grain boundaries initiating from the notch surface, exhibiting an irregular shapes with crack branching. Whereas under fatigue/oxidation condition, the crack grew in a transgranular manner. Furthermore, the role of creep, fatigue and oxidation on the failure life was dependent on the applied duration period at peak loads. Cracks were prone to nucleate in transgranular and then propagate in intergranular. There existed competitions between creep, fatigue and oxidation damage. Finally, the failure modes due to different damage mechanisms and loading conditions in the cases of creep-fatigue-oxidation were proposed. The calculated failure modes corresponded with those observed in engineering alloys.

Keywords: Creep-oxidation; creep-oxidation-fatigue; crack growth mechanism; micro-meso modelling.

1 Introduction

For the engineering components in the aero, chemical and power industries operating at high temperatures and cyclic loads, it is possible that multiaxial creep, fatigue, oxidation or a combination of these mechanisms could be reasons for premature failures. To understand the crack growth behaviour and failure life under creep or fatigue or oxidation conditions, many researches have been performed. However, modelling a combination of these mechanisms, which may contribute to premature failure, has received less attention [1, 2]. The role of creep or oxidation or fatigue damage significantly change when the materials and loading conditions vary. Therefore, various failure behaviours are observed, i.e. the creep-ductile steels in power plants components suffer the creep-fatigue [3-6], the creep-brittle high strength alloys in aerospace applications suffer the oxidation fatigue [7, 8] and the materials under complex loading conditions are subjected to the creep-fatigue-oxidation [1, 2, 9]. In addition, even though many mechanical tests have been conducted, there is scarce literature pertaining to the crack growth behaviour, trans/inter-granular failure mechanisms and their correlation with the interaction among creep, fatigue and oxidation.

Creep-fatigue-oxidation is a much more complex phenomenon. Particularly, the modelling of these combination is difficult to perform at the fundamental level. Creep damage is usually associated with the formation of micro voids at grain boundaries or grain boundaries intersections [10, 11]. In addition, high temperature components often operate under highly oxidizing or carburizing or corrosion conditions. The damage under these conditions is similar to that caused by the environmental creep deformation and is also time dependent. There are three categories of damage rate dependencies in the oxidation processes: linear relationship between damage rates and service time [12-14]; logarithmic and parabolic relationship [15-19] describing the reduced damage rates with the increased time. In contrast, the fatigue damage usually happens at the start-up and shut-down of thermal systems and the gradients in the temperature at elevated temperatures is also cycle dependent. However, the damage trend is significantly complicated when the components are operated under combined high temperature, oxidation environment and cyclic or static loading conditions. The combination could stimulate the damage accumulation towards a final premature failure. It has been revealed [20] that the damage accumulation could be accelerated for repeated buckling of oxide under compressive loading and brittle cracking under tensile loading. The load bearing section could be reduced owing to the oxidation spallation; as a result, the subjected stress increased and subsequently the remained creep life

was greatly reduced. It has also been shown [21,11] that the degradation of the creep resistance in heat resistant steels was related to the carbon dioxides development at grain boundaries. Other creep-fatigue-oxidation interaction consequences such as embrittlement or dynamic embrittlement were also observed in Nickel alloys under high temperatures [22]. The oxidation damage was serviced as the nucleation of the cracks in creep or fatigue and then affected the damage and failure modes.

The failure modes under complex environments have been studied and considered in previous work [6-8] evaluating the remaining life under creep-fatigue interaction. However, owing to the difficulties in testing and modelling the crack initiation and crack growth data under creep-fatigue-oxidation interactions, have been limited. Furthermore, for simulating intergranular or transgranular crack growth modes caused by the various failure conditions, the classical numerical simulation procedures using a finite element fine mesh distribution could not be used as no differences in material properties can be shown to exist between grains and grain boundaries. Recently, using generic grains and independent grain-boundaries structure in the FEM models to represent an idealised microstructure in real materials [7, 8,23] has shown a significant improvement in developing realistic failure simulations. This work revealed the intergranular crack growth modes under creep-oxidation and creep-fatigue conditions are possible to model and that differences can exist in the failure times. Further novel development of this approach to take into account fatigue damage is therefore beneficial to understanding the crack growth under the interactions of creep, fatigue and oxidation modes. In this paper a model using a novel mesh generation scheme which simulates idealised grains and grain boundaries the failure mechanisms can be distinguished by the mode of failure they produce. This model allows for surface depletion and inter/transgranular cracking. Elastic/plastic/creep runs using ABAQUS [24] implemented by a user subroutine in which the creep/oxidation/fatigue strain and rate dependent based criteria model is used to derive failure times for different loading and boundary conditions.

In this study, to understand the crack growth manner and the variation of the failure lives for materials subjected to the combination of creep, oxidation and fatigue, a continuum damage model is proposed which for the first time considers the combined effects of creep, fatigue and oxidation on the damage accumulation. The novel meso-modelling approach coupled with the finite element (FE) method is performed to allow the crack to propagate along grain boundaries or within grains to demonstrate the real crack growth behaviour in materials. Furthermore, the crack growth mechanism under different conditions are analysed.

2 Continuum damage Modelling of Creep and Environmental Cracking in Alloys

Creep and fatigue can be described independently with the creep and fatigue damage as a function of the applied load, and operating time and number of cycles respectively:

$$D^c = f(\sigma, t) \quad (1)$$

$$D^f = f(\sigma, N) \quad (2)$$

The oxidation/corrosion damage is dependent on time and is correlated with the distance in materials. The distance x_i from the surface where the diffusion or environmental damage can take place, which is determined by:

$$D^e = f(x_i, t) \quad (3)$$

The creep and oxidation failure mechanisms are usually intergranular. They operate independently. For one thing, the creep deformation requires an applied load. For another, although the external load and the applied stress assist in the damage process of the stress corrosion cracking (SCC), the environmental attack (such as oxidation) does not rely on the applied load. In contrast, the fatigue is cycle dependent and transgranular in nature. As a simple assumption, a linear accumulation of these mechanisms will give:

$$D^t = D^c + D^f + D^e \quad (4)$$

This allows a simple approach to predict the failure under environmental creep and creep/fatigue conditions to be developed. Individually the models have been applied solely to creep damage based on the NSW multiaxial ductility model [25, 26] and to low cycle fatigue and to oxidation [12, 13]. However this paper combines these mechanism into one novel modelling process allowing for a more a realistic approach to multiaxial failure predictions. The methods are described briefly in the next sections.

2.1 Fracture and Damage Mechanics under Creep

The creep damage increment is assumed to correlate with the local crack tip deformation and the multi-axial failure strain [27-29]. This rule has been used to represent the damage accumulation and crack extension under creep deformations. Based on the NSW model [26, 27] considering the effect of the stress state at the crack tip, therefore, the increment of the damage under the multi-axial creep condition can be determined by the ratios of the increment of the equivalent creep strain and the multiaxial creep ductility. This is defined as follows:

$$\Delta D^c = \frac{\Delta \bar{\epsilon}^{cr}}{\epsilon_f^*} \quad (5)$$

The multiaxial creep ductility, ϵ_f^* , can be obtained from a number of available void growth models as reported in [30, 31]. These model are dependent on the creep hardening parameter n . For most engineering materials, the value of the creep hardening parameter n in Norton's law lies within 5 to 10. In these cases, the creep index n in the full Cocks and Ashby equation [31] can be assumed to be a constant, allowing an approximation of the model to be written as:

$$\frac{\epsilon_f^*}{\epsilon_f} = \sinh(0.5) / \sinh(2h) \quad (6)$$

The values of $\epsilon_f^* / \epsilon_f$ is between 1-30 [27], in the extreme ranging between plane stress to plane strain conditions for creep brittle to creep ductile alloys.

The effect of the stress state on the creep deformation and damage accumulation processes are considered in this model. The creep strain rate is governed by the equivalent stress and void growth and initiation mechanisms. Hence, the creep damage initiation criterion employed in this study assumes that the critical equivalent creep strain $\epsilon_f(h, \dot{\epsilon}^{cr})$ is a function of stress triaxiality $h = \sigma_m / \sigma_e$ and equivalent creep strain rate $\dot{\epsilon}^{cr}$ giving creep damage as:

$$D^c = \int \frac{d\bar{\epsilon}^{cr}}{\epsilon_f(h, \bar{\epsilon}^{cr})} dt \quad (7)$$

where D^c denotes the creep damage accumulation. The damage is calculated using the elastic-plastic-creep analysis and then the crack initiation and crack propagation induced by the multiaxial creep conditions could be obtained. Substantial work has been conducted using this method to predict the damage accumulation and crack growth behavior [13, 26, 27] and therefore the details of the method will not be expanded in this paper.

2.2 Damage Accumulation under Low Cycle Fatigue

Low cycle fatigue behavior at elevated temperature is of considerable interest in the selection, design, and safety assessment of many engineering components. This phenomenon is often named as high temperature low cycle fatigue, when the cycles to failure are usually

lower than 100,000 cycles. In addition, the applied stress in the low cycle fatigue behavior is higher than the yield stress of the materials, where a high amplitude of plastic strain occurs. This can occur under local collapse of cracked samples or global collapse of uniaxial samples. Hence, the fatigue damage evolution per cycle is usually assumed to be correlated with the plastic range [32, 33]. On the basis of the postulation that the net tensile hysteretic energy ahead of crack tip during fatigue cycle controlled the low cycle fatigue life [34] at maximum stress the following equation describes the fatigue response at the crack tip:

$$\sigma_{\max} \Delta \epsilon_p N_f^\beta = C \quad (8)$$

In this equation, $\sigma_{\max}$ denotes the highest stress level during each cycle while $\Delta \epsilon_p$ represents the generated plastic deformation during each cycle and N , C are material constants. It can be assumed that in the creep-fatigue crack growth process, the peak stress and the local plastic range during each cycle affect the damage accumulation and dominate the crack propagation. Hence, the damage model considering the effect of stress for low cycle fatigue behaviour is defined as:

$$D_f = \int \frac{(1 - \omega_f)^{-q}}{(1/\beta) C_1 (\sigma_{\max} \Delta \epsilon_p)^{-1/\beta}} dN \quad (9)$$

where C_1 , β , C and q are material parameters related to the fatigue. In the low cycle fatigue, the cyclic soften or harden phenomenon by changes of the microstructure would occur with a progressive directional accumulation of plastic strain. A monotonic stress strain relation between the stress range $\Delta \sigma$ and the plastic range $\Delta \epsilon_p$ can be used to represent this phenomenon:

$$\frac{\Delta \sigma}{2} = K' \left(\frac{\Delta \epsilon_p}{2} \right)^{n'} \quad (10)$$

where K' is the cyclic strength coefficient and n' denotes the cyclic strain hardening exponent. The parameters in Eq. 9 and 10 are usually obtained from pure fatigue results at a specified temperature.

2.3 Damage Evolution due to Environment/Oxidation

Environmental assisted cracking covers a wide range of time dependent damage mechanisms. These are described as corrosion, oxidation, nitriding, carburisation and other

kinetics linked time dependent phenomenon. For metals there are generally three kinetic laws to characterize their degradation rates. For more complex alloys, more complicated oxidation mechanisms and descaling, pitting and micro-crack models are employed [15-17]. For describing different modes of oxidation/corrosion failure modes, the prevail model used a time growth relationship with a linear, cubic, parabolic or logarithmic growing trend.

The environmental or oxidation damage accumulation for various engineering alloys, against the operating time, tends to exhibit a parabolic kinetic response. The damage accumulation is reliant on a parabolic rate constant, k_p . This factor changes when materials, temperatures and corrosive and oxidation environments vary. Thus this can be depicted by the relationship:

$$D_p^2 = k_p t + D_{p0} \quad (11)$$

where D_{p0} is a constant, D_p is the damage film or thickness (or the mass gain due to oxidation, which is proportional to the oxide or corroded film thickness or the mass loss due to corrosion) for the parabolic rate. Differentiating the above gives an environmental damage term increment dD_p as:

$$dD_p = 0.5\sqrt{k_p} t^{-0.5} dt \quad (12)$$

The rate constant, k_p , may also be predicted within a limited temperature range according to an Arrhenius type relationship given by:

$$k_p = k_0 \exp(-Q/RT) \quad (13)$$

The total extent of the time dependent environmental damage at any time is calculated [12, 13]:

$$D^e = D^{dist} \int dD_p dt = D^{dist} \int 0.5\sqrt{k_p} t^{-0.5} dt \quad (14)$$

where D^e is the environmentally induced damage evolution state variable and increases incrementally during the coupled analysis constrained by the distance of the element from the closet surface D^{dist} . The oxidizing damage parameter k_p is derived from the measured oxidation rates using a parabolic form. This model [12, 13] had satisfactorily predicted the creep-oxidation behavior for a range of steels.

2.4 Combining Creep, Fatigue and Oxidation Damage

During the FEM simulations, the combined damage is calculated by the summation of the creep, fatigue and environmental damage. Damage D^{tot} in Eq.(4) is calculated for every

element at each iteration against the local corresponding damage initiation criteria given by Ω_G and Ω_{GB} for the grain and grain boundary region, respectively. Once this criterion satisfies the specific value during the FEM analyses, the material stiffness at the grain boundaries or gains is then degraded using a scalar damage equation:

$$\sigma = (\Omega - D^{tot}) \bar{\sigma} \quad (15)$$

where $\bar{\sigma}$ is the tensor of effective stress in the undamaged state which is computed in the current increment. D^{tot} is the overall damage variable and the threshold damage Ω criteria is expressed as Ω_G and Ω_{GB} for the grains and the grain boundaries. The critical damage of D^{tot} is set to be 0.999. Ω_G is 1 while Ω_{GB} is 0.9 for demonstrating the weakness of grains.

2.5 Randomisation Criteria for Crack Growth during the Simulation

Because there is material and testing variability represented by an overall scatter in the experimental data, the mode of the damage and surface cracking conditions can be assumed to be random in nature at the micro/mesoscale level. The development of the damage is dependent on unknown material properties and compositions affected by considerable material inhomogeneity due to the variations in crystal plasticity tensile and creep properties as well as microstructural variations at the sub-grain level. In addition to this data scatter, unknown experimental factors also produce an element of randomness in the microstructurally damage development. These effects can be simply implemented in the model as a random damage and crack growth process whereby a random multiplier given by r_n within a normal statistical distribution is allowed at each element and every time step to influence the final decision to reduce its material stiffness in the damage integration process when damage reaches the chosen critical value. Employing the Monte Carlo principle within a normal distribution [13] critical damage can then be derived from:

$$D^{ran} = D^{tot} (\Omega \pm r_n) \quad (16)$$

This changes Eq. (15) to:

$$\sigma = (\Omega - D^{ran}) \bar{\sigma} \quad (17)$$

Allowing the material stiffness to be degraded using a scalar damage equation which has a Monte Carlo distribution term D^{ran} for damage accumulation. D^{ran} is the damage level, given by Eq. (16) for each element reflecting a level of scatter representative of actual

material scatter given by r_n . In the present analysis the limits of $r_n = \pm(0.1 * \Omega)$ giving a normally distributed random deviation for each element [13].

For assuring more closely simulating real testing condition, the FEM simulation is repeated using the same input variables and the same mesh to produce a different damage and crack profile. In this way, each time step calculation is self-adjusting and the stress equilibrium is continuously maintained in the analysis. This method makes it particularly useful to compare the effects of damage and crack growth development at the grain level where grain size, orientation and grain boundary angles and local material variability affect cracking rates.

3 Mesh Design and FE Model Development

In this paper, a notched bar specimen is employed. The geometry and dimension of the specimens are shown in Figure 1, which are designed according to Reference [35]. The specimen diameter D is 5 mm and the net diameter d considering the notch is kept constant at 3 mm. The notched bar specimens having a radius, R of 1.2 mm (blunt), 0.6 mm (medium) and 0.25 mm (sharp), which correspond to the acuity level of 2.5, 5.0 and 12.0, respectively. The notch acuity ratio α defined as the ratio of the diameter of minimum cross section d and the radius of notch root r is employed to represent the notch constraint level:

$$\alpha = \frac{d}{R} \quad (18)$$

Figure 1: Dimension and size of the employed notch bar specimen.

On account of the geometrical and loading symmetric conditions, only a quarter of notched bar specimen was modelled as an axisymmetric model, as shown in Figure 2. The nodes across the mid-section of the model were constrained at x-direction (y-symmetry boundary condition) and the uniform stress, $\sigma_{applied}$, was applied on the top surface of the model and designed according to the net section stress of the notched bar specimen:

$$\sigma_{net} = \frac{D^2}{d^2} \sigma_{applied} \quad (19)$$

where σ_{net} denotes net section stress, D and d are notch diameter and net section diameter. The continuum axis-symmetric quadrilateral elements with reduced integration scheme were applied in the analyses.

Since the creep damage and crack usually was found in the notched region, the generated microstructures were only at this region to save the computational time. An idealized microstructure with random grains and finite thickness grain boundaries has been generated using the Voronoi tessellation technique [36-38]. The micro structure model replicates the prior austenite grains structures in of P92 steels. The generated random grains are illustrated in Figure 2, where the average grain size is designed as 50 μm owing to the measured average prior austenite grain in P92 steels [39]. The generated grain size ranges between about 50 - 150 μm . Although, in reality, there are no physical grain boundary regions, grain boundary elements have been defined to accommodate intergranular damage. Having considered different thickness values for grain boundaries, the optimum size found to prevent element distortion was 1 μm . This size is consistent with a realistic size of the damaged zone located at the grain boundaries in P92 steels.

In P92 steels, there are further anomalies with no clear gain boundaries and different inclusions and discontinuities in the matrix. Simply, all of these factors allow for the local stress concentrations and the proposed idealized mesh, though not fully representative of the complex microstructures, still represents local stress concentrations with varying stress state due to the random shapes and sizes of the grains. The density for these stress concentrations can be changed by either reducing grain size or adding grain to grain variability in the material properties. Thus for simplicity and in view of the microstructural complexity a finite width of 1 μm as pseudo grain boundary boundaries has been chosen. This is supported in previous work [40] by the fact that the size of creep voids containing micro cracks observed on/along grain boundaries in 9%Cr heat resistant steels is approximately 1 μm . Furthermore, it has been shown that by considering different thickness values for grain boundary damage regions that placing a limit of 1 μm width of the grain boundary is important to avoid element distortion [12, 41] .

For the FEM analysis, the microstructure region in the notched bar specimen is further refined to include the idealized microstructure. The minimum element size within the grain structured region close to the crack tip is 5 μm and 1 μm in the grain boundaries. In contrast,

gradually coarse meshes are used in the adjacent zones. In total, about 25000 to 30000 continuum axis-symmetric quadrilateral elements are generated in FE models with different notch roots and more than 90% elements are focused in the random grain regions.

Figure 2: Schematic of the employed micro structure meshes for notched specimen.

Table 1: material constants at 600 °C for P92 steel used in the model.

Elastic-Plastic properties $\sigma = K(\varepsilon)^N$	E	135 GPa
	N	0.1
	K	500.85
	σ_Y	300 MPa
Creep properties $\dot{\varepsilon}^{cr} = A\sigma^n$	A	5.0E-45
	n	17.73
Fatigue properties	K'	366 MPa
	n'	0.081
	C_1	1778
	q	0.67
	β	0.62
Oxidation properties	k_p	1.23 $\mu\text{m}/h$

4 Material Properties

The material tensile, creep, fatigue and oxidation properties for P92 steel in this paper used for the FEM modelling is presented in Table 1. Figure 3 depicts a creep strain rate dependent failure strain for P92 steel, which is defined:

$$\varepsilon_f = \frac{\varepsilon_{fmax} + \varepsilon_{fmin} \left(\dot{\varepsilon}^{cr} / \dot{\varepsilon}^{crm} \right)^{-\alpha}}{\left(\dot{\varepsilon}_c / \dot{\varepsilon}_c^m \right)^{-\alpha} + 1} = \frac{0.23 + 0.01 \left(\dot{\varepsilon}^{cr} / 6.12E-5 \right)^{-0.6}}{\left(\dot{\varepsilon}^{cr} / 6.12E-5 \right)^{-0.6} + 1} \quad (20)$$

where ε_{fmax} and ε_{fmin} are the upper and lower shelf ductility. $\dot{\varepsilon}^{crm}$ denotes the creep strain rate at which the ductility is equal to the average of the upper ductility and the lower ductility: $(\varepsilon_{fmax} + \varepsilon_{fmin})/2$. α is a constant and can be obtained by nonlinear data fitting. The properties are generic of P92 steels and batch to batch variability will undoubtedly affect the test results. As this microstructure modelling approach allows the crack to grow along grain boundaries or within grains based on loading and failure mechanism, it will be possible to identify numerically the reason why failure occurs when there is interaction between the different mechanisms.

Figure 3: A creep strain rate dependent failure strain for P92 steel at 600 °C.

5 Results and Discussion

5.1 Simulation of creep and creep/oxidation

Numerous FEM simulations were carried out to check for the models response under creep/fatigue /oxidation loading. The results of the analysis are presented here. The important point is the failure modes under multiaxial conditions due to three failure mechanisms. It is well known that under fatigue condition, cycle dependent transgranular failure mainly occurs [42] while under oxidation and creep conditions, time dependent processes mainly develop into intergranular cracking [11] with the oxidation/corrosion occurring at the surface [43, 44]. Therefore, in this paper no further experimental results are presented in order to verify the numerical modelling approach.

Simulated crack growth profile under the combination of creep and oxidation is shown in Figure 4, where a mainly intergranular crack growth profile with additional crack branching is observed. Moreover, by incorporating the Monte Carlo principle, the damage and the crack would be varied slightly at different positions and produce a different crack growth and damage accumulation profile after every run, which could represent the scatter in oxidation data. Furthermore, it can be noted that the crack mainly propagates along the grain boundaries, exhibiting intergranular failure. The high energy level of the non-equilibrium state and the free volume of random high-angle grain boundaries cause grain boundaries to act as fast diffusion paths for the metallic alloying elements at high temperature. Cr-Mo heat resistant steels had been reported to be prone to oxidation at high temperatures, which leads to the lower values of surface roughness [45]. As creep deformation fully dominates crack development, the crack is likely to propagate along the grain boundaries [42]. This indicates that the creep-oxidation condition does not change the crack growth failure mechanism in comparison with pure creep regime.

Figure 4: Simulated crack growth profile under creep-oxidation conditions.

The effect of the combination of creep and oxidation on the failure life in notched bar specimen is shown in Figure 5, where the experimental failure life under pure creep condition is provided for comparison. A typical notch radius of 1.2 mm is adopted. A similar trend is observed in all specimens. The simulated creep failure life is in consistent with the

experimental data. Furthermore, it can be clearly noted that the oxidation significantly affects the failure life of the notched bar specimen, particularly at low loads and longer terms.

With the decreasing in the applied stress, the reduction in the failure life resulting from oxidation is increased. This is because that the oxidation damage is dependent on the time and allows surface crack initiation to develop. As the failure time is short, the oxidation damage is small. As a result, it has slight effect on the failure behaviour at short times. With the failure time increasing at low stress, the oxidation damage level steeply increases. Once a critical oxide concentration is exceeded, the nucleation and growth of micro-cracks on the surface of specimen is stimulated. This can serve as the nucleation sites of subsequent intergranular creep damage developing as voids or cracks [46]. When investigating the influence of oxidation on creep rupture life for T23 steel, Sawada et al. [47] had shown that the formation of oxide scales was found to contribute significantly to the long term creep strength degradation, due to reducing the creep failure ductility. Moreover, the oxidation can produce the loss of the available load-bearing cross-section and lead to the stresses acting on the component increase. For accurately demonstrating the role of oxidation on creep life, Nakashiro et al. [48] corrected the creep rupture data by calculating the thickness of the oxidation and the true stress in 2.25Cr-1Mo steel. The magnitude of this effect will, of course, depend on the initial wall thickness [49]. Thus, the failure life under creep-oxidation conditions is reduced much more for thinner sections.

Figure 5: Predictions for combination of creep and oxidation on failure life in notched bar data compared to notched data creep rupture at 600 °C [35].

Figure 6 shows the variation of failure life for specimens with different notch radii. At the same net section stress, the failure life of sharp notch bar specimen is higher than the other two specimens, due to the notch strengthening behaviour for P92 steel [28, 47, 48]. It is interesting to note that as the applied stress is high, the failure life of specimens with sharp and blunt notch is almost equivalent. When the applied stress becomes low, the role of the oxidation becomes more influential, leading to a different tendency with the convergence of the failure lives.

Figure 6: Effect of notch radius on the failure life [35].

The combination of creep and oxidation also changes the crack initiation positions. Figure 7 shows the crack initiation contour for specimens with different notch radius at the same net section stress of 190 MPa. It can be noted that without considering the oxidation, the crack is likely to initiate at a distance from the notch root, which is consistent with those in literatures [47, 48]. The crack initiation position is located at the near the position owning the highest stress triaxiality across the notch throat, which stimulates the creep damage accumulation [28, 49]. When the notched specimens are subjected to pure creep loadings, the crack initiation position is varied with the notch radius. The different notch shape would produce the different distribution of the stress triaxiality. As a result, as shown in Figure 7, the crack initiation position under pure condition is varied as the notch radius changes [48].

When the oxidation damage is considered, the crack initiation behaviour shows a very different trend. The cracks are likely to be initiated at the notch root regardless the notch radius is. Since the oxidation damage is focused on the outer surface of specimen, the damage in elements near the surface reach critical damage due to the enhanced oxidation and not due

to creep. Figure 8 shows an example of the microstructures of material after steam oxidation [43]. The oxidation in P92 steel is focused on the surface and leads to the reduction of loading section.

Figure 7: Variation of crack initiation behaviour under pure creep and creep-oxidation conditions for specimens with different notch radii at the same net section of 190MPa (a) $R=1.2$ mm, (b) $R=0.6$ mm and (c) $R=0.25$ mm.

Figure 8: Microstructure of oxidation layer in P92 steel [43] .

Figure 9 compares the crack initiation position for notched specimens under creep and creep-oxidation conditions. Different from the pure creep condition, the notched specimens under creep-oxidation are all initiated at the notch root as the applied stress is low. When the applied net section stress is higher than 270MPa, the crack initiation positions coincide with the specimen under pure creep condition. Because the higher net section stress leads a shorter creep failure life, a lower oxidation damage rate is anticipated.

Figure 9: Crack initiation position against applied net section stress for specimens with different notch radii.

5.2 Simulation under creep/fatigue/oxidation

Under the combination of creep, fatigue and oxidation, several distinct processes are simultaneously observed. The microstructural evolution and environmental interactions at high temperature allows for fatigue damage and softening effect to occur. These are caused by cycling, accumulation of creep damage during holding periods [50]. It is important to identify their distinct roles. Figure 10 shows the role of the creep-oxidation-fatigue on the failure life of notched bar specimen. In this case a trapezoid load wave was employed, where the load and unload period is 2s and the duration period at the peak load varies from 30s, 60s and 600s to reveal the role of the creep-fatigue interaction. The load ratio between the peak load and the minimum load per cycle is set to be 0.1, which means that there is only tensile

loading considered in these cases. Compared with Figure 6, the application of fatigue significantly reduces the failure life. This phenomena was confirmed by Benjamin et al. [54] and U. Führer [55], studying the high temperature creep-fatigue-oxidation interactions in 9-12%Cr martensitic steels and revealing that oxidation phenomena strongly decreased the creep-fatigue lifetime compared with the results in vacuum. Moreover, it can be noted that with the duration period reducing, the failure life is greatly reduced. Because a short duration period would improve the fatigue damage, this increases the damage accumulation and thus reduces the failure life. In addition, under different hold times, the failure life of specimens with blunt notch is equivalent to that of specimen with medium notch. At the same condition, the shortest failure life occurs in the specimen with sharp notch. Because fatigue is shown to be sensitive to the stress concentration.

Figure 10: Effect of duration period on the failure life of notched bar with different notch radius.

Figure 11: Crack growth profile for specimens under different hold duration periods.

Figure 12(a): Observed crack growth mode under creep-fatigue-oxidation (a) transgranular cracking near the surface [40] and (b) combination of inter/transgranular cracking [56].

Figure 11 analyses the crack growth mode under different hold duration periods. This is carried out on the notched specimen with blunt notch. In other specimens, similar crack growth profiles are obtained. Initially, the crack grows through grains or grain boundaries. The cracks are all initiated from the outer surfaces, different from the pure creep condition (see Figure 7). Experimental evidence in the literature also reveals that the cracks initiate at outer surfaces due to oxidation [50, 51]. As an example Figure 12 shows that combination of oxidation and fatigue makes the crack initiate at the outer surfaces. In Figure 11, with the

crack growth length increasing, the intergranular creep cracking mode again dominates the crack growth behaviour. Moreover, the region with transgranular crack growth manner is reduced as the hold duration period increases from 30s to 600s. The dwell time on the failure behaviour of Alloy 718 showed the presence of both transgranular and intergranular cracks under different hold times [56], as shown in Fig.12 b of electron backscattered diffraction (EBSD) mapping on crack cross sections. It can be noted that the crack propagation was purely transgranular at the initial and then became the intergranular. This is in a good agreement with the calculated crack growth manner with different dwell times, as shown in Fig. 11. The environmental effect is enhanced due to the applied dwell time, leading to a mixed mode of crack propagation and accelerated crack growth. Moreover, some independent cracks ahead of the main crack front are observed in Fig.11. This may be caused by the complex and three-dimensional nature of the crack growth process.

Figure 13 compares the damage evolution process for specimens exposed to different hold times. The oxidation damage under the combination of creep-oxidation-fatigue is small, because the failure life is greatly reduced by fatigue. Thus, only the creep and fatigue damage components are considered in Figure 13. It can be noted that the fatigue damage component gradually decreases to the minimum value and then keeps constant. In contrast, the creep damage component exhibits a contrary trend, increasing to the peak value. Moreover, the fatigue damage values are higher than the creep damage as the crack growth length is small. This results in the transgranular crack growth behaviour near the notch root as shown in Figure 11. As the crack tip moves gradually away from the notch root, the fatigue damage is reduced and the creep damage is increased. When the crack growth length increases further (more than 0.3 mm), the creep damage level almost approaches to the maximum value and keeps constant. This suggests that the creep deformation again controls the crack growth behaviour.

Thus, the intergranular crack growth mode instead of transgranular mode is observed at near the centre of notched specimen (see Figure 11). Moreover, the level of the fatigue damage component in the specimen with short duration period is higher than the specimen with longer hold time. Specially, it is interesting to note that the creep damage level becomes higher than the fatigue damage component at crack growth length of 0.1 mm for specimen exposed to 600s hold time. But this value increases to about 0.3 mm for specimens subjected to shorter hold time. There is, therefore, a competition between creep and fatigue damage. A

higher damage level means that the role of creep would become more influential and change the crack growth mode [42, 52, 53]. As a result, the portion of the transgranular crack growth is reduced as the duration period at peak loads decreases.

Figure 13: Damage evolution process against crack growth length under creep-fatigue-oxidation environments.

The variation of the damage component evolution with the crack growth length for specimens exposed to creep-fatigue-oxidation loadings can be attributed to three reasons. Firstly, although the notch introduces a high stress concentration near the notch root, as shown in Figure 14, the higher stress is only limited in the region near the notch root. The stress near the notch root is significantly higher than the net applied stress (190 MPa) while the stress located at the centre is much smaller than the net section stress [49]. A high stress introduce a localized yielding near the notch root region. Thus, fatigue damage is higher near the notch root. With the distance away from the notch root, the stress level is reduced, which in turn reduces the fatigue damage levels. Hence, near the notch root, the fatigue damage component is much higher than the creep damage component (see Figure 13).

Figure 14: The equivalent stress distribution across the notch throat for specimens with different notch radii.

Figure 15: Equivalent stress and stress triaxiality distribution during the crack growth process (a) equivalent stress and (b) stress triaxiality.

Secondly, the stress would be redistributed during the crack propagation process. The distribution of equivalent stress and stress triaxiality ahead of crack tip is analysed in Figure 15. It should be clearly noted that after the redistribution of the stress, the region near the crack tip has the higher stress level. Specifically, the stress in the grain is slightly higher than that in grain boundaries, see Figure 15a. But as shown in Figure 15b, a much higher stress triaxiality is located at elements in the grain boundaries near the updated crack tip compared with elements located in grains. This process is due to creep damage greatly dependent on

stress triaxiality while fatigue damage is only reliant on the applied stress. Thus the presence of the higher stress triaxiality would stimulate the creep damage accumulation [54, 55] and cause the crack propagate along the grain boundaries rather than grains. The process can be further accelerated since the grain boundaries are intrinsically weaker than the grains under creep diffusion mode allowing failure at the boundaries to occur more easily.

Finally, fatigue damage is shown to be cycle dependent. Figure 16 provides the crack growth against the normalized time t/t_f (t_f is the failure life) for specimens with different duration periods. It can be noted that as long as the crack length increases more than 0.2 mm, the crack growth rate would increase rapidly. This means that there is no time for the fatigue damage accumulation. Hence, creep damage controls the crack growth as the crack length becomes higher.

Figure 16: Crack growth length against normalized time for specimens with different hold time.

Figure 17 summarises the crack growth mechanism for specimen under creep, oxidation and fatigue conditions. When only the pure creep condition is considered, the crack propagates along the grain boundaries and always initiates at a distance away from crack tip, exhibiting typical intergranular failure (see Figure 4). Specially, it is always observed that two crack initiation sites exist: towards centre plane and the notch root. As shown in Figure 17b, the crack initiation and crack growth direction are all changed as the oxidation damage is

incorporated and the crack initiates at the notch root. Coalescence of the creep cavities would have led to the propagation of crack from the notch root region towards the central region of the notch plane. Moreover, the crack dominated by oxidation damage may propagate along the grain boundaries or inside grains. But this region only accounts for a smaller portion. Besides, the intergranular growth manner is observed. Figure 17c shows the crack growth manner under pure fatigue condition, which exhibits the typical transgranular crack growth profile independent of the microstructure. Moreover, the crack growth is not very smooth. As long as the crack growth length is larger than half of diameter in specimen, the crack growth becomes straight, exhibiting a brittle failure. This is due to the reduction of ligament, which stimulates the damage accumulation across the specimen. When the duration period at peak load is applied, a mixed of transgranular and intergranular crack growth behaviour is observed (see Figure 17d and e). The transgranular growth firstly occurs due to the fatigue damage controlling the failure and subsequently the creep deformation dominates the crack growth, showing the intergranular failure. By conducting the interrupted tests, it was found [62] that the phenomenon was similar and that the crack growth transited from transgranular into intergranular mode. In addition, an oxide in the crack tip region was observed. As the duration period increases, the portion of the transgranular crack growth behaviour is gradually reduced. The cracks will propagate along the damaged grain boundaries with no reason to deviate from a simple straight path. When analysing the microstructures under creep-fatigue and pure fatigue test conditions by back scattered scanning electron microscope (SEM it was revealed [63] that although the cracks were prone to nucleate in a transgranular manner, it tended to propagate intergranularly along precavitated grain boundaries. This was due to the competition in the creep-fatigue interaction.

Figure 17: Illustration of crack growth mechanism under different environmental conditions (a) pure creep, (b) creep-oxidation, (c) pure fatigue, (d) creep-fatigue-oxidation with short duration period and (e) creep-fatigue-oxidation with long duration period.

6 Conclusions

It can be confidently stated that it is the first time that a new model combining the failure mechanisms due to creep/fatigue and oxidation corrosion in a fracture mechanics approach has been presented in the literature. The robust method adopted, based on the fundamental understanding of the failure mechanisms combined with a crack initiation and growth progressive failure approach will profoundly alter the predictive structural integrity applications in the field both at the R&D in materials and mechanical engineering and at industrial level due to its objective approach to the problem. The specific conclusions of the present work can be summarised as follows

(1) The oxidation damage significantly reduces the failure life and changes the crack initiation behaviour for P92 notched bar specimens. It leads the crack to initiate at the notch root, rather than at a distance from the notch root under pure creep conditions. Moreover, the simulated crack under the combination of creep and oxidation mainly propagates along the grain boundaries, exhibiting an irregular shape. This is consistent with the typical failure behaviour under pure creep conditions.

(2) The reduction of the failure life under creep-fatigue-oxidation conditions is much higher than under the creep-oxidation condition and is varied with the applied duration period at peak loads. The crack initiation position is still located at the notch root in comparison with the creep-oxidation conditions. But the crack growth manner is more complex than that in creep-oxidation regime.

(3) The fatigue damage component gradually decreases to a minimum value while the creep damage component exhibits an increasing trend. Moreover, the fatigue damage values

are high as the crack growth length is small and the high fatigue damage is located in the region near notch root. The creep damage level becomes higher than the fatigue damage component at crack growth length of 0.1 mm for specimen exposed to 600s hold time. But the crack growth length increases to about 0.3 mm for specimens subjected to a shorter hold time. This greatly affects the corresponding crack growth manner.

(4) The modelling of crack growth mechanisms for P92 notched specimen under creep, oxidation and fatigue conditions has been highlighted in this work and use to show that the novel approach is consistent with real material behaviour. Pure creep load and creep-oxidation produced intergranular crack growth behaviour while the cyclic load causes the crack propagate inside grains, exhibiting the transgranular failure. However, when the duration period at peak loads is applied, a mixed of transgranular and intergranular crack growth behaviour is presented, where the portion of the transgranular crack growth behaviour is small and reliant on the duration period.

Acknowledgement:

This research work was financially supported by Project of National Natural Science Foundation of China (Grant No. 51505328). In addition support from EDF Energy High Temperature Centre at Imperial College is acknowledged in the same manner.

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