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Short Creep-Fatigue Crack Growth in an Advanced 9 %Cr Steel

Reference

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ABSTRACT

While the development of short cracks due to cyclic elastic loading has been relatively widely studied, in particular at room temperature, their consideration for cyclic inelastic loading at high temperatures and lower frequencies is not so common. Short creep-fatigue crack growth rates may be correlated in terms of cyclic strain range, cyclic J -integral or strain energy density factor, with appropriate allowance for associated creep damage accumulation. Candidate approaches are evaluated with reference to test results generated for an advanced 9 %Cr turbine rotor steel. This paper promotes the use of cyclic strain range and strain energy density factor relative to other candidate correlating parameters in relation to the results of a series of 30 min hold time creep-fatigue tests performed using fully instrumented uniaxial specimens with short crack starters. The focus of the testing campaign is an advanced 9 %Cr turbine rotor steel at temperatures of 600 and 625°C.

Keywords

short creep-fatigue crack growth, J -integral, strain energy density factor, 9 %Cr steel

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Nomenclature

A, B, C, D, E = stress-strain hysteresis loop area designations in Figs. 1 and 2

A_J, A_K, A_W = constants in fatigue crack growth equations (Eqns. 1, 6 and 11)

a = crack length (in m, unless stated otherwise)

a_o = initial crack starter length (0.2 mm)

a_{temp} = temporary calculated crack length in PD conversion

a_{f_temp} = temporary calculated final crack length in PD conversion

a_f = real final crack length, as measured by post-test measurement

$\dot{a}_c$ = creep crack growth rate with time

B = function in Eq 14 (defined by Eq 15)

C^* = parameter characterising stress and strain rate fields at tip of crack in material deforming due to creep

D = specimen diameter

DCPD = direct current potential drop (crack monitoring system)

E = elastic modulus

EPFM = elastic plastic fracture mechanics

g = function of geometry and cyclic strain hardening exponent used in derivation of J -integral

K_{max}, K_{min} = maximum and minimum stress intensity factor

K_{cl} = crack opening stress intensity factor

LC = long crack

LCF = low cycle fatigue

LEFM = linear elastic fracture mechanics

MSC = microstructurally short crack

m_J, m_K, m_W = exponents in fatigue crack growth equations (e.g., Eqs 1, 6, and 11)

N, N_2 % = cycle number, number of cycles to crack initiation (2 % load drop, see Fig. 5)

n = cyclic strain hardening exponent

PSC = physically short crack

Q = crack size exponent in Eq 14 (typically close to 1)

q = C^* exponent in Eq 4

q_o = fraction of the total load range for which a crack is open

r = specimen radius

t_h = hold time during each cycle

w_e = elastic energy expended per cycle

w_p = plastic energy expended per cycle

- w_d = plastic strain energy dissipated per cycle (where ' $w_d \cdot a$ ' is the strain energy density factor)
 V = instantaneous PD value
 $V_{0.2}$ = PD value at the beginning of the test (at $a_o = 0.2$ mm)
 Y = compliance function, e.g., for round bar tensile specimens with edge chordal crack (Y_{cc}) or edge semi-circular crack (Y_{scc}) as defined in Ref. [42]
 α = damage exponent in Eq 16
 β = constant in A_K function of creep-fatigue crack growth law (Eq 5)
 β_{oxide} = coefficient in A_K function of creep-fatigue crack growth law (Eq 5) representing hold time effect of oxide assisted growth on cyclic crack growth component
 β_{creep} = coefficient in A_K function of creep-fatigue crack growth law (Eqn. 5) representing hold time effect of prior creep damage on cyclic crack growth component
 $\varepsilon, \Delta\varepsilon$ = strain, strain range
 $\varepsilon_{max}, \varepsilon_{min}$ = maximum strain, minimum strain
 $\Delta\varepsilon_e, \Delta\varepsilon_p, \Delta\varepsilon_t$ = elastic strain range; plastic strain range, total strain range
 $\Delta J_e, \Delta J_p, \Delta J_t$ = elastic, plastic and total range of J integral
 ΔK_{eff} = effective stress intensity factor range
 γ = constant in creep crack growth rate equation (i.e., Eq 4)
 θ = included angle shown in Fig. 4
 $\sigma, \Delta\sigma$ = stress, stress range
 $\sigma_{max}, \sigma_{de}$ = maximum stress, stress at end of hold time (dwell period)
 φ = creep damage function in Eq 16

Introduction

Short creep-fatigue crack growth properties have been determined for a number of engineering alloys in recent years (e.g., Refs. [1–9]), albeit not so extensively as those relating to long cracks essentially propagating under linear elastic loading conditions. The consideration of creep-fatigue crack development within a cyclic plastic zone is now an integral part of published high temperature defect assessment procedures (e.g., Ref. [10]), and guidance for the generation of crack growth properties for such defect integrity calculations is now available [11]. The following paper reviews candidate approaches for modelling short creep-fatigue crack growth rates with reference to test results generated for an advanced 9 %Cr turbine rotor steel.

During operation, high temperature turbine rotors are subjected to a combination of thermal transients, self-weight bending, and centrifugal loading. Of these, thermal transients are usually the most influential with respect to near surface defect development, due to repetitive changes in temperature gradient during service duty, e.g., Refs. [12,13].

During turbine start-up, compressive stresses may be generated at features on the rotor periphery, which can become residual tensile due to reversed plasticity when the temperature ultimately attains a steady stable distribution throughout the section. During steady operation, the tensile residual stresses relax while the thermal strain remains unchanged. During turbine shut-down, the rotor surface cools down more rapidly than the interior, resulting in the generation of additional tensile stresses at the rotor surface which reduce to close to zero when the turbine is at standstill at ambient temperature [14–16]. This type of operating cycle is typically represented in the laboratory by means of isothermal strain controlled low cycle fatigue tests with hold time(s) at peak strain in tension and/or compression [17,18].

In order to improve the effectiveness of design assessment and subsequent inspection management, it is necessary to fully incorporate the influences of creep, oxidation, and microstructural degradation into the modelling of high temperature fatigue crack initiation and growth [19,20].

Creep-Fatigue Crack Propagation

LONG CRACK GROWTH

Low frequency fatigue crack growth becomes increasingly influenced by oxidation and creep interaction effects at high temperature [21]. While the frequency is relatively high, and the loading linear elastic, fatigue crack growth rate may be represented by the effective stress intensity factor range [14,22,23], i.e.,

$$(1) \quad da/dN = A_K (\Delta K_{\text{eff}})^{m_K}$$

$$(2) \quad \text{where } \Delta K_{\text{eff}} = K_{\text{max}} - K_{\text{cl}} = q_0 \Delta K = q_0 (K_{\text{max}} - K_{\text{min}})$$

As the frequency is reduced or hold time(s) increased, for cycle shapes involving one or more dwell periods, long creep-fatigue crack growth rates may be expressed as the sum of crack growth rates due to fatigue (e.g., during transient loading) and crack growth rates due to creep (e.g., during hold time(s) or steady loading), i.e.,

$$(3) \quad \begin{aligned} (da/dN)_{\text{total}} &= (da/dN)_{\text{fatigue}} + (da/dN)_{\text{creep}} \\ (da/dN)_{\text{total}} &= (da/dN)_{\text{fatigue}} + \int_0^{t_h} \dot{a}_c dt \end{aligned}$$

where fatigue crack growth rate is given by, for example, Eq 1, and creep crack growth rate per unit time is expressed as a function of the C^* parameter, i.e.,

$$(4) \quad \dot{a}_c = \gamma (C^*)^q$$

In practice, the total (creep-fatigue) crack growth rate is not simply the sum of Eqs 1 and 4. Creep crack growth during the hold time is preceded by the generation of creep and strain enhanced oxidation damage at the crack tip. These typically enhance the rate of fatigue crack growth and influence the magnitude of the fatigue crack growth constant (to become a function replacing A_K as originally given in Eq 1) [14], i.e.,

$$(5) \quad (da/dN)_{\text{total}} = \left(\beta + \left(\beta_{\text{oxide}} + \beta_{\text{creep}} \right) t_h \right) (\Delta K_{\text{eff}})^{m_K} + \int_0^{t_h} \gamma (C^*)^q dt$$

SHORT CRACK GROWTH

Fatigue crack development can initially be divided into three stages, namely micro-structurally short crack (MSC) growth, physically short crack (PSC) growth, and long crack (LC) growth [24,25]. The first stage, MSC, is sometimes also regarded as the nucleation stage. A pre-defined length which is typically of the order of the grain size is often used to characterise the end of the nucleation stage. Thereafter, the fatigue crack propagates into the subsurface, usually along a path perpendicular to the loading direction. This stage is of major importance since a large fraction of component life can be spent in propagating a PSC into an LC [4,26,27] when the crack is no longer contained within the cyclic plastic zone. Crack sizes greater than the cyclic plastic zone size are typically regarded as long cracks (LCs). Modified linear elastic fracture mechanics (LEFM) or elastic-plastic fracture mechanics (EPFM) approaches are available to describe crack growth behaviour within the PSC regime. Partial closure during cyclic loading is considered by some to be the main feature of PSC growth and, in these circumstances, effective instead of nominal values of stress intensity factor are used, e.g., as given in Eq 1 [14,22,23].

A commonly adopted crack growth correlating parameter for the PSC regime is the energetic path-independent contour integral J [28], although the Tomkins model [1] has also been successfully employed when the fracture mode is purely transgranular (e.g., Ref. [29]). For short cracks, the region of plasticity surrounding the crack tip is relatively large compared to the crack length, LEFM loses its validity [30], and the J -integral can be adopted to represent the strain energy release rate. For LCF crack growth, propagation rate may be represented as a function of cyclic J integral, i.e.,

$$(6) \quad da/dN = A_J (\Delta J_t)^{m_J} = A_J (\Delta J_e + \Delta J_p)^{m_J} \quad \text{m/cycle} \quad (\Delta J_t \text{ in MPa}\cdot\text{m})$$

In such circumstances, ΔJ corresponds to the part of the cycle for which the crack is open [27]. Accurate J -integral determinations require FE calculation (e.g., Shih and Hutchinson [31]), in particular for component analysis. For a number of laboratory specimens, there are approximate solutions, e.g., for round bar specimens with a semi-circular surface defect [26] (originally derived from Ref. [31]), i.e.,

$$(7) \quad \Delta J_t = 3.2 w_e a + 5.0 w_p a$$

where w_e and w_p are elastic and plastic energies expended per cycle, i.e.,

$$(8) \quad w_e = (\Delta \sigma)^2 / 2E$$

FIG. 1

Representation of nominal stress and strain during cycling of a uniaxial testpiece and of expended energies used in the determination of ΔJ_t .

$$(9) \quad w_p = \Delta\sigma \cdot \Delta\varepsilon_p / (1 + n)$$

and where n is the cyclic strain hardening exponent. In terms of the cyclic stress-strain hysteresis loop shown in **Fig. 1**, w_e is area E and w_p is area A + D. It is also possible to express short crack growth rates simply as a function of cyclic plastic J -integral, where ΔJ_p is defined by the second term in Eq 7, such that:

$$(10) \quad da/dN = A_J (\Delta J_p)^{m_J} \quad \text{m/cycle} \quad (\Delta J_p \text{ in MPa.m})$$

Whereas ΔJ_p is determined from w_p (A + D in **Fig. 1**), the plastic strain energy dissipated per cycle, w_d , is derived from the internal area of the cyclic stress-strain hysteresis loop [32], A + (B + C) in **Fig. 2**. Typically under these conditions, the part of energy dissipated as heat is small, which means most of the energy is expended in microstructural evolution (e.g., Ref. [33]) and crack development [34]. Irrespective of the different mechanisms of creep and fatigue, w_d can account for their effects

FIG. 2

Representation of hysteresis loop shape with hold time at peak strain in tension.

and serve as an indication of the driving force for crack extension [35] as part of a strain energy density factor ($w_d \cdot a$), i.e.,

$$(11) \quad da/dN = A_w (w_d \cdot a)^{m_w} \quad \text{m/cycle} \quad (w_d \text{ in MPa})$$

For continuous cycling hysteresis loops, w_d is only related to the shaded area A as shown in Fig. 2 [36].

$$(12) \quad w_d = \Delta\sigma\Delta\epsilon_p - 2 \int_0^{\Delta\sigma} \epsilon_p d\sigma = \frac{\Delta\sigma\Delta\epsilon_p}{1+n} (1-n)$$

When dwell time is incorporated, the total energy dissipated should be the sum of all three areas (A, B, and C) in Fig. 2 [37], i.e.,

$$(13) \quad \Delta w_d = \frac{\Delta\sigma\Delta\epsilon_p}{1+n} (1-n) + \frac{\sigma_{\max}^2 - \sigma_{\text{hold}}^2}{2E} + \frac{\Delta\sigma(\sigma_{\max} - \sigma_{\text{hold}})}{2E(1+n)}$$

SHORT CREEP-FATIGUE CRACK GROWTH

Thermal fatigue is generally a strain controlled process, and a widely used empirical expression for high strain low cycle fatigue (LCF) short crack growth when creep damage is not so significant, is given by [3]:

$$(14) \quad da/dN = Ba^Q \quad \text{m/cycle} \quad (\text{with } a \text{ in m})$$

The value of B is directly related to total strain range $\Delta\epsilon_t$ and environment, which can be determined experimentally; Q is material dependent, but is usually close to unity. As short creep-fatigue crack growth rates increase with increasing hold time, an upper bound is recommended in the R5 assessment procedure to be [4,10]

$$(15) \quad B = 2.61 \times 10^4 \Delta\epsilon_t^{2.85} \quad (\text{mm})^{1-Q}/\text{cycle}$$

In order to account for creep damage more precisely, a number of modified forms of Eqs 1, 6, and 14 considering interactions between the crack front and local microstructural features such as grain boundaries and cavities are suggested with the general form of

$$(16) \quad (da/dN)_{\text{total}} = da/dN \cdot (1/(1-\varphi))^z$$

where φ can be a function of the volume fraction of cavities [38], a function of cavity population, stress, strain, and strain rate [39], a function of cracked grain boundaries per cycle [2], or a function of threshold stress intensity factor due to crack length and the strongest microstructural barrier [40]. Moreover, φ in Eq 16 can also be creep damage fraction, expressed in terms of either ductility exhaustion or time fraction [10].

Short Creep-Fatigue Crack Growth Data for 9 %Cr Steel

EXPERIMENTAL DETAILS

In order to further investigate the mechanics and mechanisms of short creep-fatigue crack growth, a series of tests were performed with a recently developed advanced 9 %Cr high temperature turbine rotor steel.

Round bar tensile test specimens were taken from close to the periphery of a full-size production rotor forging and machined into cylindrical profiles with shoulders, threaded-ends, and shallow crack starters (**Fig. 3**). The parallel gauge length section was 20 mm long, with a diameter of 8 mm. The short straight-fronted chordal crack starter was located at mid-span with a depth of 0.2 mm and a width of 0.3 mm. Tests were performed in displacement (nominal strain) control using a Class 0.5 side contacting extensometer, with a 15 mm gauge length, and a 50 kN servo-electric testing machine. Temperature was controlled in accordance with the requirements of ASTM E2714 [41] by means of a resistance heating furnace. Three Type N thermocouples were attached to the top, middle, and bottom of the specimen gauge section to ensure a homogeneous temperature distribution throughout the test.

For the purpose of this study, the approximate ΔJ_t solution for a round bar tensile specimen with an edge semi-circular surface crack (Eq 7) was modified to represent a round bar tensile specimen with an edge chordal crack, i.e.,

$$(17) \quad \Delta J_t = (3.2w_e a + 5.0w_p a) \cdot (Y_{cc}/Y_{scc})^2$$

The modification involved scaling the Dowling constants in Eq 7 with the $(Y_{cc}/Y_{scc})^2$ factor where Y_{cc} and Y_{scc} are compliance functions (of a/D), respectively, for round bar tensile specimens with edge chordal cracks and edge semi-circular cracks (as defined in Ref. [42]).

In order to accurately monitor crack development for the starter notch, a constant electrical DC current was passed between the specimen shoulders. The potential drop (DCPD) across the chordal crack was measured using two output wires symmetrically spot-welded to either side of the crack starter, with a spacing of 4 mm. High quality ceramic parts were mounted between both specimen ends and the loading grips to provide good electrical insulation of the specimen from the test machine.

FIG. 3 Details of round bar tensile specimen with edge chordal crack starter.

FIG. 4 Crack front evolution from chordal crack starter; (a) schematic representation, (b) actual example.

Short crack growth rates were determined at 600 and 625°C in a series of displacement (nominal strain) controlled continuous cycle and cyclic/hold tests. Nominal strain across the specimen gauge length was controlled with a strain ratio ($\epsilon_{\min}/\epsilon_{\max}$) of -1 . Continuous cycle tests involved ramping between $\epsilon_{\max}$ and $\epsilon_{\min}$ by means of a triangular waveform with a strain rate of 0.001 s^{-1} . Cyclic/hold tests involved the same wave form, but with a 30 min hold time at either peak strain in tension ($\epsilon_{\max}$) or compression ($\epsilon_{\min}$).

Testing was interrupted before final fracture at high temperature so that it was possible to determine a final crack size for in-test DCPD calibration (e.g., **Fig. 4**). This was achieved by high cycle fatigue crack growth at room temperature to reveal an unoxidised remaining ligament fracture, and thereby the full extent of (oxidised) crack extension at high temperature. In addition to final crack size measurement, the broken specimen halves were also used to provide samples for post-test metallurgical inspection.

TEST RESULTS AND DATA ANALYSIS

Stress Response

While in principle, the experiments were short creep-fatigue crack growth tests, they were in fact conducted as strain-controlled creep-fatigue crack initiation endurance tests (in accordance with Ref. [41]). Like other precipitation strengthened alloys at high temperatures (e.g., Ref. [43]), the investigated 9 %Cr steel exhibited a cyclic softening stress response, at first relatively rapidly, and then more steadily. Initially this response is mainly due to sub-grain development as a consequence of dynamic recovery [43,44] but then, at approximately half life, the effect of damage development starts to become increasingly apparent. In particular, the rate of stress drop increases as a consequence of increasing compliance due to the increased extent of damage development. Since the starter slot adopted in this study had a finite width and root radius (**Fig. 3**), the same stress response to that for a plain specimen was

FIG. 5

Typical stress response record during short creep-fatigue crack growth tests.

anticipated, but in fewer cycles due to the effect of strain concentration at the notch root (e.g., Fig. 5).

Characteristic points in the test records were (1) the beginning of the steady softening regime, (2) the mid-life cycle number (i.e., $0.5.N_2\%$), (3) the end of the steady softening regime, and (4) the number of cycles to 2 % load drop (i.e., $N_2\%$), a commonly adopted creep-fatigue crack initiation criterion (e.g., Refs. [43,45]). The influence of total strain range, temperature, and hold time on the number of cycles to 2 % load drop in the round bar tensile specimens with short edge chordal crack starters is shown in Fig. 6.

Crack Development

The evidence from post-test inspection indicated that crack front development remained approximately straight and parallel to the root of the starter chordal crack (e.g., Fig. 4(b)). On the basis of the model representation given in Fig. 4(a), the following electrical DCPD crack size calibration formulation could be devised, i.e.,

$$(18) \quad (1 - 0.9933(V_{0.2}/V))\pi = \theta - (\sin 2\theta)/2$$

$$(19) \quad \text{where } \theta = \arccos(1 - a_{\text{temp}}/r)$$

Temporary crack length, a_{temp} , could therefore be calculated directly from the resistance provided by the specimen net section area (i.e., Eqs 18 and 19), although there was still the need for further correction. By breaking specimens open after test, to accurately determine the final crack length (a_f), a could then be obtained with the following correction formula (with $a_0 = 0.2$, i.e., Fig. 3):

$$(20) \quad a = (a_{\text{temp}} - a_0) \cdot \frac{a_f - a_0}{a_{f_temp} - a_0} + a_0$$

The effectiveness of this approach is evident from Fig. 7, and a comparison of final crack size observations made during post-test inspection with those predicted from electrical DCPD crack measurement indications using Eqs 18 and 19.

FIG. 6

Influence of total strain range, temperature and hold time on cycles to 2 % load drop in round bar tensile specimens with short edge chordal crack starter (data points for 0.01 $\Delta\epsilon_t$ are offset for clarity).

In this way, DCPD voltages could be converted to crack length values. With the exception of one test condition (0.4 % $\Delta\epsilon_t$ at 600°C), crack lengths propagated to ~ 0.5 mm at the start of the steady softening regime (1), to ~ 0.75 mm at midlife (2), to ~ 1.75 mm at the end of the steady softening regime (3), and to ~ 2 mm at $N_{2\%}$ (4) (Fig. 8). In view of this evidence, and consistent with the finding of others (e.g., Ref. [4]), interest in the present study was limited to $a \leq 2$ mm ($a/D \leq 0.25$).

The influence of total strain range, temperature, and hold time on crack growth rate at the characteristic points is shown in Fig. 9. Increasing strain range and the introduction of a 30 min hold time are the most influential parameters on growth rate, in particular for crack sizes above ~ 1 mm, within the steady softening regime (i.e., beyond (2)). For crack sizes less than ~ 1 mm, the effect of strain range and a 30 min hold time is not so high. The influence of increasing temperature from 600 to 625°C on da/dN also appears to be relatively minor.

FIG. 7

Comparison of final crack size observations made during post-test inspection with those predicted from electrical DCPD crack measurement indications using Eqs 18 and 19.

FIG. 8

Influence of total strain range, temperature and hold time on crack length developed at characteristic points.

In addition to providing the opportunity to directly measure final crack size, post-test inspection enabled metallographic evaluation of the influence of temperature, and hold time on short crack development mechanism. At temperatures of both 600 and 625°C, the LCF short crack development mechanism was transgranular. With the introduction of a 30 min hold time, there was evidence of creep voids in the vicinity of growing cracks, and a tendency for cracks to branch and propagate with an intergranular component, in particular at the higher temperature.

Hold Time

Tests were performed to study the effect of hold time position in the cycle (i.e., either at peak strain in tension ($\epsilon_{\max}$) or peak strain in compression ($\epsilon_{\min}$)). While there is much published evidence to indicate that hold times in compression can be far more influential than hold times in tension in reducing the crack initiation endurance of

FIG. 9

Influence of total strain range, temperature and hold time on crack growth rates at characteristic points.

FIG. 10

Comparison of short creep-fatigue crack growth rates for 9 %Cr steel at 600/625°C with upperbound represented by Eq 14(15) (data points for 0.0075 and 0.01 $\Delta\epsilon_t$ are offset for clarity).

the advanced 9 %Cr steels (e.g., Refs. [46,47]), the evidence in Figs. 6, and 8-10, suggests that there is only a very small systematic effect of the position of a 30 min hold time on short creep-fatigue crack growth rates for the advanced 9 %Cr rotor steel at 600/625°C (in particular for $a \leq \sim 1$ mm, i.e., before characteristic point (2)).

The influence of strain range and 30 min hold time on creep-fatigue crack growth rates for the 9 %Cr steel at 600/625°C, by example for a crack size of 0.5 mm, is shown in Fig. 10, and compared with the upper-bound represented by Eq 14 (with B defined by Eq 15). Equivalent comparisons for crack sizes of 1.0 and 2.0 mm are similar. Even for cycles with a 30 min hold time, Eq 14(15) provides a very conservative estimate of short creep-fatigue crack growth rates for the 9 %Cr steel at 600/625°C. There is only a relatively small influence of hold time for the conditions investigated, and careful examination of the DCPD records indicated that no significant growth occurred during the dwell periods of the conducted tests. This indicated that any effect of the applied hold times on growth rate was due to the influence of prior creep and strain enhanced oxidation damage on fatigue crack growth rather than any contribution of creep crack growth (during the hold time) to the total propagation rate (e.g., with reference to Eq 5).

da/dN($\Delta\epsilon_t$) Correlation

The effectiveness of a number of short creep-fatigue crack growth correlating models was investigated. A commonly adopted approach is use of the empirical strain range based model represented by Eq 14(15). For the performed tests on the 9 %Cr steel, without and with hold time, Fig. 10 shows $da/dN(\Delta\epsilon_t)$ data co-ordinates, by example for a crack size of 0.5 mm. The relative dependence of crack growth rate on strain range for the 9 %Cr steel is consistent with that defined by Eq 14(15), although the resistance to crack growth is almost an order of magnitude superior to that of the upper-bound relationship (Fig. 10).

Crack growth rates are enhanced (almost equally) by a 30 min hold time at either peak strain in tension or compression. For the 9 %Cr steel, this is evidence to indicate that short creep-fatigue crack growth under these conditions is mainly

affected only by prior creep and strain enhanced oxidation damage effects on the fatigue crack growth component of cracking rather than by creep crack growth during the hold time. This effect can easily be accommodated within a time dependent version of the damage function attached to Eq 16 (e.g., Ref. [6]).

da/dN(ΔJ_t) Correlation

Figure 11 shows the crack growth rate versus ΔJ_t for all test pieces. For the case of the round bar specimens with chordal edge cracks used in this study, ΔJ_t is determined using Eq 17. Generally, most points lie within the scatter band between the two black dashed lines representing ± 2 times the mean Eq 6 formulation with parameter values of $A_J = 1.6 \times 10^{-4}$ and $m_J = 1.1$. It is notable that within the overall scatter-band there appears to be an influence of $\Delta \varepsilon_t$ on individual test m_J values. This could be due at least in part to the fact that the uncertainty associated with the determination of n in Eq 9 is relatively high when $\Delta \varepsilon_p$ is very small compared with $\Delta \varepsilon_e$. n is typically derived by direct fitting to experimental stress-strain loops (e.g., Ref. [36]). Notably, the ΔJ_t parameter does not fully account for any of the effects of hold time.

da/dN(ΔJ_p) Correlation

An alternative J -integral crack growth correlating parameter is that provided by ΔJ_p (i.e., the plastic part of Eq 17). Relative to $da/dN(\Delta J_t)$, the low to high $\Delta \varepsilon_t$ $da/dN(\Delta J_p)$ correlation appears to be more effective, although the effects of hold time are still not fully accommodated (**Fig. 12**). Generally, most points lie within the scatter band between the two black dashed lines representing ± 2 times the mean Eq 10 formulation with parameter values $A_J = 9 \times 10^{-5}$ and $m_J = 0.86$.

da/dN($w_d \cdot a$) Correlation

The strain energy density factor is a function of the plastic strain energy dissipated per cycle and the crack length, i.e., $w_d \cdot a$. The w_d parameter used to determine the strain energy density factor (i.e., $A + B + C$) in **Fig. 2** is similar to, but not identical to the w_p parameter used to determine ΔJ_p (i.e., $A + D$ in **Fig. 1**). With this in mind,

FIG. 11

Crack growth rates for 9 %Cr steel at 600/625°C expressed as function of ΔJ_t .

FIG. 12

Crack growth rates for 9 %Cr steel at 600/625°C expressed as a function of ΔJ_p .

the $da/dN(w_d \cdot a)$ correlation shown in **Fig. 13** appears to be similar but marginally more effective to that for $da/dN(\Delta J_p)$. The black dashed lines in **Fig. 13** represent ± 2 times the mean Eq 11 formulation with parameter values $A_w = 7 \times 10^{-4}$ and exponent $m_w = 0.86$. Importantly, the strain energy density factor appears to be able to more closely account for the effect of hold time. Clearly the effectiveness of this has to be confirmed with the results of longer hold time tests.

Discussion

SHORT CREEP-FATIGUE CRACK GROWTH CORRELATING PARAMETERS

A number of stress/strain based parameters used to correlate short creep-fatigue crack growth rates for a 9 %Cr steel at 600/625°C have been examined using the

FIG. 13

Crack growth rates for 9 %Cr steel at 600/625°C expressed as a function of $w_d \cdot a$.

results from elastic–plastic strain controlled tests involving round bar tensile test specimens with shallow chordal crack starters. At the simplest, and perhaps most practical level is the $da/dN(\Delta\epsilon_c)$ correlation adopted in Ref. [10], e.g., Fig. 10. In contrast to this largely empirical approach, it is possible also to adopt debatably more physically based energy parameters. Derivation of, for example, the $da/dN(\Delta J_I)$, $da/dN(\Delta J_p)$ and $da/dN(w_d \cdot a)$ models relies on a knowledge of cyclic stress-strain hysteresis loop details from laboratory tests. Examples of the effectiveness of these correlations are shown respectively in Figs. 11, 12, and 13. While it is acknowledged that the effectiveness of such parameters when applied to other geometries has still to be verified, the evidence indicates an advantage associated with use of the strain energy density factor relative to other parameters.

EFFECT OF PRIOR OXIDATION/CREEP DAMAGE

In the past, detailed long creep-fatigue crack growth studies revealed that total crack growth rate could be determined by summing cyclic crack growth rate with the crack growth rate due to creep during the hold time (i.e., Eq 3). More importantly, it was shown that the cyclic crack growth rate component can be enhanced by the effects of creep and strain enhanced oxidation damage generated at the crack tip during previous hold times (i.e., Eq 5), e.g., Ref. [14]. For short creep-fatigue crack growth rate, accommodating the enhancing effects of prior crack tip creep and oxidation damage can be achieved through the incorporation of a generic time dependent damage function of the type included in Eq 16. The adoption of such a function with other short creep-fatigue crack growth correlating parameters is also feasible, although the evidence from this study suggests that this may not be necessary in the case of the strain energy density factor (Fig. 13). Of course, this must be confirmed by testing the effectiveness of the parameter with the results of longer hold time tests. Moreover, it is likely that the strain energy density factor will only be capable of accounting for prior crack tip deformation effects. It will still be necessary to introduce a damage function to account for the generation of physical crack tip damage on the fatigue crack growth rate component of short creep-fatigue crack development (e.g., creep cavities). This is the subject of current research.

Concluding Remarks

Candidate approaches to model short creep-fatigue crack growth rate in an advanced 9 %Cr high temperature turbine rotor steel have been evaluated using the results from elastic–plastic strain-controlled test specimens with shallow chordal crack starters. Ideally, such model approaches should be capable of accounting for growth rates due to creep and high strain fatigue transients as well as any associated interaction effects. Two promising approaches involve model formulations based either on total strain range or strain energy density factor.

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