UNDERSTANDING FATIGUE-CREEP INTERACTIONS

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ABSTRACT

Previously reported data on the behaviour of 1 CrMoV and Type 316 stainless steel during high strain fatigue tests containing tensile dwells have been examined. Under the respective test conditions there is a marked difference in failure mode between the two alloys. To achieve an understanding of this, the rates of damage accumulation in both fatigue and creep have been computed, and the relevant microstructures determined. It is concluded that 1 CrMoV is more susceptible to creep-dominated failure due to easy cavity nucleation at grain boundaires, its relatively poor creep ductility and low cyclic strain hardening exponent. In contrast Type 316 has few grain boundary precipitates, a high ductility and strain hardening exponent, and generally experiences a fatigue dominated failure.

KEYWORDS

Damage accumulation; microstructure; 1 CrMoV, Type 316; Dominant failure mode.

INTRODUCTION

Creep-fatigue interactions have now been studied in a wide range of engineering alloys (Conway, 1969; Cheng and others, 1972; Ellison and Paterson, 1976; Wareing, 1977; Plumbridge, Priest and Ellison, 1979). Such studies have involved interactions in which fatigue and creep are applied sequentially, or simultaneously, when an element of each damaging mode exists in every cycle. The interaction between creep and fatigue has been investigated using smooth specimens and those containing notches from which crack growth may be monitored. The present study is concerned with two specific sets of results from smooth test pieces during simultaneous creep-fatigue in a 1 CrMoV ferritic steel and a Type 316 stainless steel. (Ellison and Paterson, 1976; Wareing, 1977). These alloys represent a wide range of material properties and microstructure, the former in the fully-bainitic condition being creep brittle (creep fracture strain less than 10%), whereas the latter, in the solution treated condition, is creep ductile. However, both alloys exhibit a reduced endurance when a tensile dwell period is inserted in the strain cycle, but as discussed later there is a marked difference in the damage that causes this reduction in life. The principal objective of the present work is to provide an understanding of why this difference arises, so that when extrapolation of laboratory data is undertaken, the equations governing appropriate damage mechanism will be used.

SUMMARY OF PREVIOUS TESTS ON 1 CrMoV and 316

Experimental details of the tests on 1 CrMoV are given elsewhere (Ellison and Patterson, 1976). In this paper we are only concerned with the effects of dwell periods inserted at peak tensile strains, i.e. cycles as shown in Fig. 1b. For 1 CrMoV it was found that endurance decreased with increasing dwell time. Metallogrphic examination (Plumbridge, Priest and Ellison, 1979) revealed that the failure process was 'creep dominated' and that fatigue cracks at the surface were small and not connected to the creep damage, which had built up during successive dwells Fig. 1b. The creep damage was predominantly in the form of grain boundary cavitation. In this study failure was designated as five percent fall off in load from the stress range saturation plateau. These observations suggest that there is little, if any, interaction between fatigue and creep damage micromechanisms.

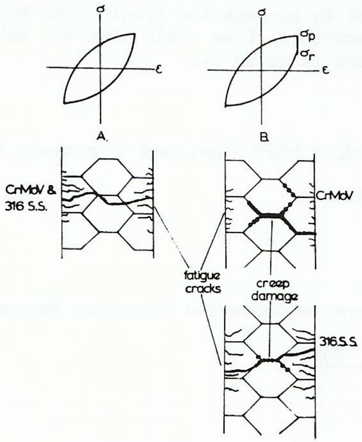


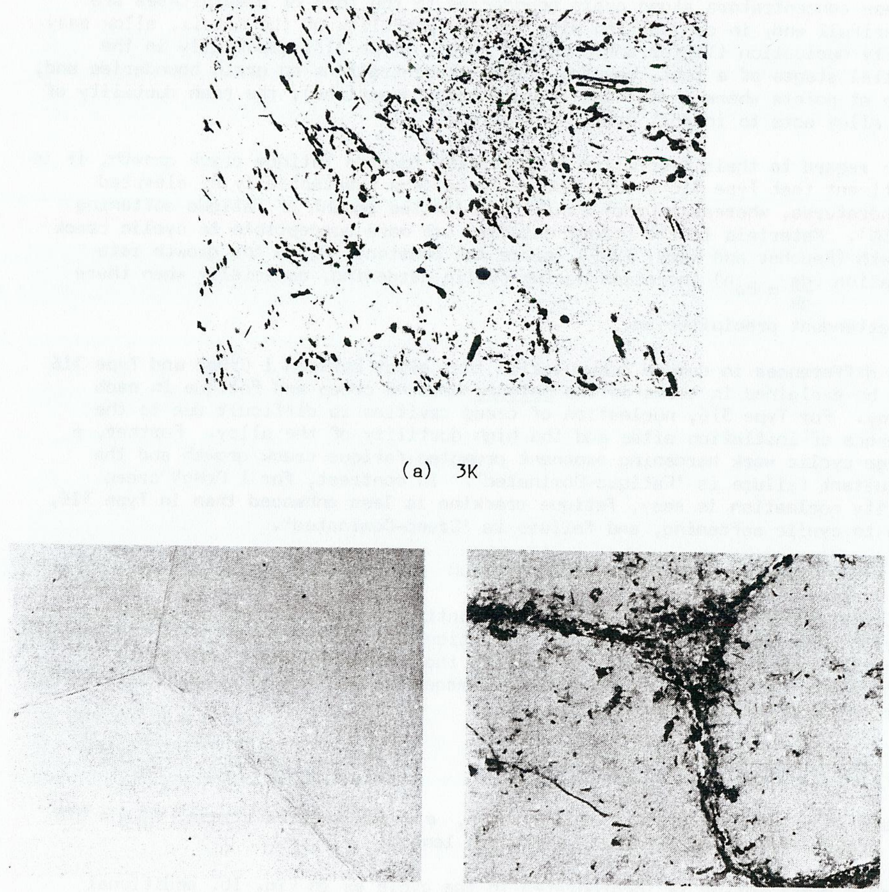
Fig. 1 a) Damage associated with a pure fatigue, 1b) Damage associated with creep-fatigue interaction in 1 CrMoV and 316.

For Type 316 stainless steel a substantially different mechanism of failure during cycling with tensile hold periods was reported by Wareing (1977). Again, endurance diminished with increasing dwell period, but failure arose due to the growth of a surface initiated fatigue crack Fig. 1b. The rate of extension of this single, major, crack had been enhanced when it encountered grain boundary cavitation damage which was formed during the dwell period. Therefore, in this case the failure was (fatigue dominated), and a strong interaction between fatigue and creep damage micromechanisms had occurred. The following paragraphs will attempt to explain these two extreme examples.

THE ROLE OF MICROSTRUCTURE AND MECHANICAL PROPERTIES

In this section the influence of microstructure and mechanical properties on damage accumulation in the two alloys during dwell periods is examined. Figure 2a is typical of the tempered bainitic 1 CrMoV steel, which contains a fine dispersion of M_4C_3 carbides, together with coarse M_7C_3 carbides in

the matrix and along prior austenite grain boundaries. Due to its high tempering temperature, the alloy is very stable and is unlikely to change microstructurally during the course of a laboratory test.



(a) 3K

(b) 3.6K

(c) 3.6K

Fig. 2 a) Tempered bainite
b) Solution treated 316
c) Solution treated and aged at 625°C 316

Type 316 in its solution treated condition contains no precipitates (Fig. 2b), but they occur on ageing at the test temperature (Weiss and Stickler, 1972). Figure 2c shows precipitation occurring after approximately 500 h at 625°C (Dean and Plumbridge, 1980).

During the dwell period stress relaxation occurs, rapidly at first, but for the majority of the dwell to an almost constant level. The deformation permitting stress relaxation consists of matrix flow and grain boundary sliding with the subsequent nucleation and growth of grain boundary cavities. Deformation during the dwell is under constrained conditions and it is thought that this will give rise to more deformation at grain boundaries, enhancing damage. In 1CrMoV, stress concentrators along grain boundaries in the form of precipitates are plentifull and, in conjunction with the low ductility of the matrix, allow easy cavity nucleation (Tipler and Hopkins, 1976). Type 316, especially in the initial stages of a test, has few stress concentrations on grain boundaries and, even at points where they exist (triple point junctions), the high ductility of the alloy acts to inhibit crack nucleation.

With regard to their respective propensities towards fatigue crack growth, it is pertinent that Type 316 steel cyclically hardens substantially at elevated temperatures, whereas 1 CrMoV exhibits a limited amount of fatigue softening (~15%). Materials in the former category are more susceptible to cyclic crack growth (Reuchet and Remy, 1979), since the constant, B, in the growth rate equation ($\frac{da}{dN} = Ba^n$) increases during cyclic hardening, especially when there is attendant precipitation.

The differences in damage accumulation that exist between 1 CrMoV and Type 316 can be explained in terms of the balance between creep and fatigue in each alloy. For Type 316, nucleation of creep cavities is difficult due to the absence of initiation sites and the high ductility of the alloy. Further, a large cyclic work hardening exponent promotes fatigue crack growth and the resultant failure is 'Fatigue-Dominated'. In contrast, for 1 CrMoV creep cavity nucleation is easy, fatigue cracking is less enhanced than in Type 316, due to cyclic softening, and failure is 'Creep-Dominated'.

ANALYSIS OF DAMAGE ACCUMULATION

In this section an attempt is made to quantify the rate of damage accumulation during the cycle shown in Fig. 1. Attention will be confined to the actual data reported earlier in this work to explain those experimental findings. For continuous cycling (Fig. 1a), fatigue damage, in the form of a crack, acrues at a rate given by (Tomkins, 1968),

$$\frac{da}{dN} = \Delta\epsilon_p \left(\sec\left(\frac{\pi\sigma}{2\sigma_{UTS}}\right) - 1 \right) a \quad (1)$$

where $\Delta\epsilon$ · applied plastic strain range, σ= peak tensile stress, σ_{UTS} = UTS of previously cycled material, a = crack length.

When a dwell period is incorporated in the cycle as in Fig. 1b, additional damage due to the creep processes described earlier takes place and, if the accumulation of this damage exceeds the ductility of the material, then fracture occurs. Wareing (1977) has suggested that when there is cavitation due to hold periods and failure occurs by extension of a fatigue crack, then a upper bound to crack growth rate (the lower bound on endurance) is given by

$$\frac{da}{dN} = \Delta\epsilon_p W \quad (2)$$

where W = width of the specimen. Equations (1) and (2) represent the limits of fatigue crack growth rates during cycling.

Creep damage during the dwell may be formed by several mechanisms depending upon the test conditions but in the present analysis we consider only grain boundary cavitation. Growth of cavities can occur by either unconstrained diffusion, constrained diffusion or deformation controlled growth mechanisms. In the case of the latter two mechanisms, growth rate is proportional to strain rate. For simplicity, we have therefore assumed that growth is by unconstrained diffusion mechanism, since a representative strain rate is difficult to determine during the dwell period. A number of theoretical models for this growth mechanism exist (Speight and Harris, 1967; Speight and Beere; 1975), the difference between them being mostly due to the chosen boundary conditions. For the purpose of this analysis, the growth rate given by Speight and Harris (1967) model has been used. This gives

$$\frac{dr}{dt} = \frac{D_{gb}\delta(\sigma - 2\gamma/r)\Omega}{2kTr^2(\ln(\lambda/2r)-\frac{1}{2})} \quad (3)$$

where D_{gb} = grain boundary diffusion coefficient, σ= applied stress (minimum relaxed), k = Boltzmann's constant, r = cavity radius, λ= cavity spacing, T = temperature, γ= surface energy, δ= boundary width. For creep-fatigue studies it is more convenient to express the growth rate in terms of dr/dN instead of dr/dt. This is possible if we state that the total time of the fatigue-creep interaction test is given by,

$$t_t = N(t_h + t_c) \quad (4)$$

where N = number of cycles, t_h = hold time, and t_c = cycle time.

Therefore, from Equation (4)

$$\frac{dt}{dN} = (t_h + t_c) \quad (5)$$

Equation (3) from creep damage accumulation therefore becomes

$$\frac{dr}{dN} = \frac{(t_h + t_t) \quad D_{gb}\delta\Omega(\sigma - 2\gamma/r)}{2kTr^2} \quad (6)$$

since $(\ln(\lambda/2r) - \frac{1}{2}) \simeq 1$.

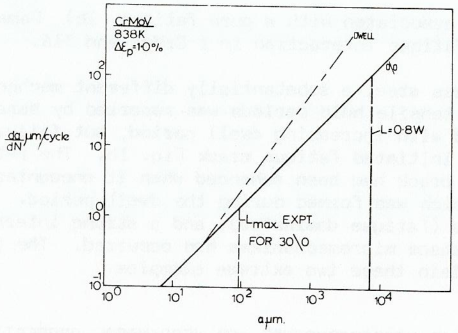
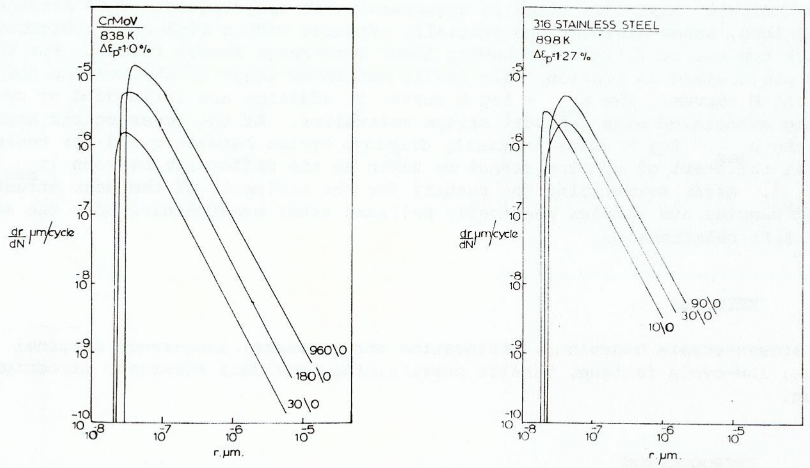


Fig. 3 Fatigue crack growth rate vs crack length

Figure 3 shows a plot of da/dN vs a for 1 Cr Mo V steel determined from equation (1), such a plot is not possible for equation (2). However, if interaction occurs, incorporation of a dwell period would increase the fatigue crack growth rate at a given crack length due to interaction with creep damage, as shown schematically in Fig. 3. The fracture criterion used for a fatigue dominated fracture is L_f = 0.8W as shown in Fig. 3, this is orders of magnitude greater than the largest fatigue crack observed in 30/0 in 1 CrMoV which was 96μm. This implies that there is no creep enhancement of fatigue crack growth in 1 CrMoV. In direct contrast a similar plot for 316 would reveal experimental fatigue crack length to be equivalent to 0.8W, the reduced life in this case being indicative of accelerated fatigue crack growth. This therefore gives rise to two effects of the dwell period on fatigue crack growth rates. In 1 CrMoV the growth rate decreases with increasing dwell period, since there is no interaction and failure is creep dominated. Whereas in 316 interaction occurs and fatigue crack growth rate is accelerated.

Fig. 4a shows the effect of hold times on cavity growth per cycle as a function of the cavity radius for 1 CrMoV steel. For a given radius the cavity growth rate increases as the dwell period increases under the appropriate test conditions. Similar growth rates are also found for 316 stainless steel as shown in Fig. 4b. Thus differences in the extent of creep damage must be associated with a difference in the ability of each alloy to nucleate creep cavities as discussed earlier.

Direct comparison of the magnitude of da/dN and dr/dN is not viable, simply due to the definition of each. From the above analysis it can be seen that as the dwell period increases then the accumulation of creep damage relative to fatigue damage increases. In 1 CrMoV it remains separate from the cyclic cracking, but in 316 it supplements fatigue crack growth. However, with long dwells and low strain ranges (as often encountered in service) it is possible, even for this alloy, that creep damage could become isolated and the failure creep nominated.



A B

Fig. 4 Effect of dwell time on cavity growth rate, for a given cavity radius.

CONCLUDING REMARKS

In this paper two previously reported simultaneous fatigue-creep interactions have been examined and the divergence in their failure modes has been explained. In 1 CrMoV, which has a low creep ductility and abundant sites on grain boundaries for cavity nucleation, creep damage exceeds that due to fatigue and the failure is creep dominated. The only interaction is that at final failure. In 316, which is creep ductile and possesses few grain boundary precipitates until these form by ageing during the test, failure is via the growth of a fatigue crack through a creep damaged material. Thus, for extrapolation or life prediction of simultaneous interactions of fatigue and creep, the appropriate modified creep equation should be used for 1 CrMoV and appropriate fatigue expression for 316.

It must be emphasised that two specific investigations have been considered in this paper. Work is continuing on the more general situation to determine the relative effects of temperature, strain range and duration of dwell period. For any alloy, creep domination is favoured by a high temperature, a long dwell and low strain range.

However, for design purposes it is necessary to identify quantitatively the interface between fatigue and creep domination to ensure that extrapolation of laboratory data is soundly based.

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