Fracture 1977, Volume 4, ICF4, Waterloo, Canada, June 19 - 24, 1977

CREEP-FATIGUE INTERACTION FAILURE IN TYPE 316 STAINLESS STEEL

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INTRODUCTION

Most fatigue failures in metals and alloys at both ambient and elevated temperature occur by the initiation and subsequent propagation of a dominant surface nucleated crack. For smooth specimens and relatively undefected components in the limited life region above the endurance limit, initiation is often rapid in terms of final endurance although the initiated crack (length a_o) is very small (of order 10 μm). Endurance N_f is then an integration of crack growth rate, da/dN, from a_o to a final crack size a_f at which unstable ductile or brittle fracture occurs. Now, a_o and a_f are both structure sensitive but on a different structural scale, a_o is defined materially by dislocation substructure and is also influenced by surface environment e.g. microscale corrosion effects. a_f however is related to the material macro-defect structure, e.g. second phase particles, inclusions, which dominate fracture. The third parameter determining N_f, da/dN, may or may not be structure sensitive but is dependent on material mechanical properties (σ_y, n) and also applied stress-strain field (σ, ε_p) and environment.

Now endurance at elevated temperature, is much more sensitive to the types of cycle imposed than at ambient temperature. Variations in strain rate and the introduction of dwell periods can have order of magnitude effects on endurance. This has led to difficulties in attempting to draft design codes for elevated temperature power plant e.g. ASME Code Case 1592, on a similar basis to low temperature codes such as ASME III in relation to fatigue endurance.

The present paper shows briefly how cycle shape affects the three parameters which determine endurance at elevated temperature for one material, type 316 stainless steel. The principles involved, however, apply to a wider range of alloys.

FATIGUE CRACK GROWTH AT ELEVATED TEMPERATURE FOR CONSTANT STRAIN RATE CYCLES

It is now well established that fatigue cracks propagate incrementally by shear decohesion at the crack tip during the crack opening process [1, 2]. Hence the means by which cracks accommodate the plastic opening demanded, sometimes termed plastic blunting, is the means by which they advance. If one examines the equations for crack tip opening (δ) under elastic and fully plastic applied stress-strain fields a general equation can be derived with the form,

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$$\delta \approx \frac{Aa}{\beta E} \ln [\sec(\beta\sigma)] + \frac{Aa\,\varepsilon_p}{(n+1)} \tan(\beta\sigma) \qquad (1)$$

for small values of work hardening exponent n [3]. A is a constant (~1 for plane strain crack opening), $\beta = \pi/2T$ (where T is the material flow stress) and σ and ε_p are the applied stress and plastic strain. For small values of σ/T, equation (1) reduces to,

$$\delta \approx \frac{\beta\sigma^2 a}{2E} + \frac{\beta\sigma\varepsilon_p a}{(n+1)} \qquad (2)$$

Now the general equation for fatigue crack growth rate [3] has elastic and plastic components similar to equation (2), viz.

$$\frac{da}{dN} = \frac{A_1\Delta K^m}{Ef(T)} + A_2 \left(\frac{\Delta\sigma}{2T}\right)^2 \frac{\Delta\varepsilon_p a}{(2n + 1)} \qquad (3)$$

A comparison between equations (2) and (3) shows some similarity in form between elastic and plastic components but da/dN is usually only a fraction of the CTOD, δ.

Now in the plastic straining region of fatigue failure, where most creep-fatigue studies have been made and cycle wave shape effects investigated [4], it is the plastic component of equation (3) which dominates crack growth. For a given applied strain range, the crack growth rate is sensitive to the ratio ($\Delta\sigma/T$). In earlier work on a 20 Cr/25 Ni stainless steel, Wareing et al [5] showed that this ratio was increased by increasing temperature and decreasing strain rate at elevated temperature. The result in endurance terms was a considerable decrease, reflecting the crack growth rate effect. The increase in strain rate sensitivity of T at elevated temperature is a result of stress relaxation creep processes which operate at moderate and low strain rates.

Figure 1 shows the effect of cycle strain rate on the fatigue endurance of type 316 stainless steel tested at a plastic strain rate of 1.43% at 625°C. It can be seen that the endurance is approximately halved when the strain rate drops below 10^{-4}/s. Fractographic evidence showed that the striation spacing, which at this strain level corresponds to the crack growth increment, to crack depth ratio increased by a corresponding factor of two as the cyclic strain rate decreased. This is also shown in Figure 1. It is worth noting also that at the strain rates below the transition, the crack path was increasingly intergranular but striations of the expected size were found on grain boundary facets. This indicates that at low strain rates grain boundaries represent a weaker although more constrained crack path with a lower effective flow stress, T. There was no evidence that intergranular failure occurred ahead of the current crack front. Woodford and Coffin [6] have noticed similar crack path behaviour in the A286 alloy at 593°C.

In this type of cycle involving a constant strain rate, the growth rate and hence endurance is determined by the material flow properties. An upper limit on growth rate occurs when $\Delta\sigma/2 \to T$ and work hardening is no longer effective in limiting crack opening and hence extension - the work hardening equivalent to general yield in perfectly plastic material. The crack growth rate is then given by the plastic displacement applied to the cracked section (W - a) and,

$$\frac{da}{dN} = \frac{\Delta\varepsilon_p W}{(W - a)} (W - a) \qquad (4)$$

which is independent of crack length. Integrating between a_a and a_f gives a lower bound on endurance as,

$$N_f = \frac{(a_f - a_o)}{\Delta\varepsilon_p W} \qquad (5)$$

All the work to date on type 316 stainless steel for the symmetrical constant strain rate cycle indicates that endurance variations are related primarily to crack growth rate variations and a_f and a_o are essentially constant.

CYCLES INVOLVING VARIABLE STRAIN RATES

In early work on type 304 stainless steel at 650°C, Berling and Conway [7] discovered that the introduction of some imbalance into a cycle, e.g. by means of a dwell period at maximum tensile strain, could lead to a reduction in endurance for an otherwise high strain rate cycle. They also found that the reduction in endurance was primarily due to the formation of grain boundary cracks or cavities in the bulk material into which the crack was propagating. These observations have since been confirmed by many experimenters on various high temperature alloys, particularly austenitic stainless and ferritic steels. The bulk cavitation seems to be a result of tensile creep deformation during stress relaxation in the dwell period which is reversed by rapid plastic straining elsewhere in the cycle. Several stress relaxation periods are needed to generate significant cavitation [5].

Fractography and metallographic sections have shown that the overall failure process is complex but follows a three phase pattern shown in Figure 2, which is a schematic section through a failure. The first phase is simple fatigue with insufficient cavitation damage developed to affect either growth rate or crack path. Striation observations in this phase have confirmed that the crack growth rate is just that expected for the constant higher strain rate of the main tensile plastic strain increment. In other words, no decrease in endurance would occur for the imbalanced loop if all growth were of the phase I type. However, after a time phase I is succeeded by phase II growth where cavitation ahead of the crack front predetermines the crack path as a grain boundary path and influences the growth rate. No direct measurements have been made of phase II growth rates. Finally, phase III intervenes to give premature unstable fracture of cavitated material. In terms of the three parameters mentioned as those determining endurance, in this type of failure a_o is unaffected, a_f decreased and da/dN increased, but only during phase II.

In earlier papers, Tomkins [3] and Wareing [8] examined a simple lower bound endurance based on a maximum crack growth rate given by equation (4) and a fixed reduced a_f of 0.8W. For this bound it was assumed that no phase I growth existed and that all growth was by phase II through material which was cavitated enough to satisfy the displacement criterion given by equation (4). This does represent a lower bound if a correct value for a_f is chosen but it is most pessimistic at lower strain levels. It also lacks some reality in that for the effective flow stress, $(T = T(1 - (p/L)^2)$ where p/L is the ratio of cavity size : spacing) to be reduced significantly p/L must be large which will only apply after a considerable accumulated creep strain.

Recently, Raj, Tomkins and Wareing [9] have proposed a bound based on phase I growth and a final crack size a related to cavity spacing. This bound is based on the fractographic evidence that phase I growth rates are not significantly affected by small amounts of bulk cavitation. This is consistent with the small effect on T of a modest p/L ratio. It is also observed that phase I terminates when the crack opening displacement, δ, equals the cavity spacing L. Once this criterion has been satisfied, cavity linkage can occur readily along the flow bands radiating at ±45° from the crack tip where displacement falls relatively slowly [2, 10]. Thus phase II growth is rapid and the effective a_f is given by the crack length at the end of phase I. This is given by equation (1) with $\delta = L$.

Now in practice L is related in type 316 stainless steel to grain boundary carbide particle spacing, as these particles provide the main cavity nucleation sites. Two batches of type 316 steel were tested with an unbalanced cycle involving a tensile dwell period. These batches had an average particle spacing of 8 μm (Batch I) and 2 μm (Batch II). Figure 3 shows how the endurances of these two batches varied and how they related to predictions of endurance based on phase I growth and the $\delta = L$ criterion for a_f. Also included in Figure 3 are some data of Brinkman and Korth [11], which showed particularly severe hold time effects. The particle spacing for the material used in the latter tests is estimated at 1 μm on the basis of this criterion.

It is thought that continued examination of failure patterns in such unbalanced cycle, creep-fatigue situations will enable more rational rules to be developed for design based on true failure criteria.

SUMMARY

A study of some elevated temperature cycles on type 316 stainless steel has shown that an understanding of the three main parameters involved in failure development can lead to rational explanation of endurance variations and provide a more realistic basis for engineering design in cyclic situations.

ACKNOWLEDGEMENTS

The authors are indebted to T. Topping and H. G. Vaughan for experimental assistance.

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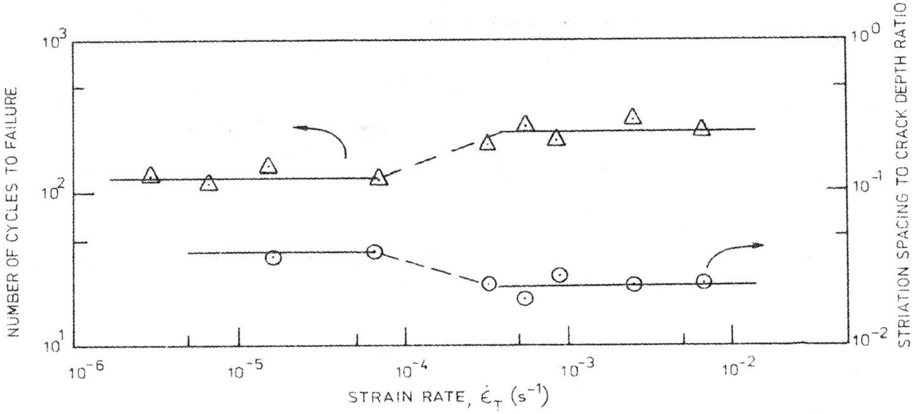

Figure 1 Relationship between fatigue endurance, striation spacing to crack depth ratio and cyclic strain rate for type 316 stainless steel at 625°C and plastic strain range 1.43%.

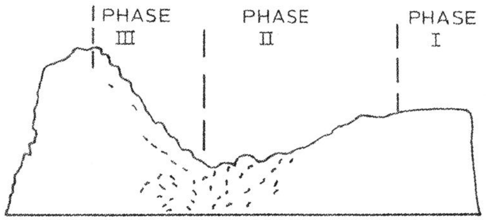

Figure 2 Schematic representation of the three phases of crack growth which occur during creep-fatigue failure.

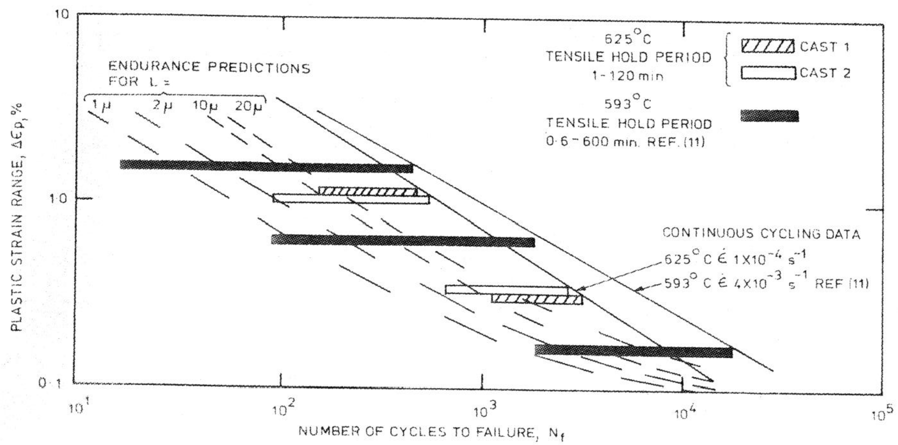

Figure 3 Comparison between predicted and actual creep-fatigue endurances of type 316 stainless steel