FATIGUE CAVITATION AT HIGH TEMPERATURE

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ABSTRACT

At low temperatures fatigue fracture occurs due to the propagation of a surface-connected crack. However, at high temperatures an intergranular damage accumulation process, i.e., cavity nucleation, growth and linkage at grain boundaries is dominant. This review summarizes recent work devoted to the topic of fatigue cavitation. The role in intergranular fatigue failure of grain boundary migration and sliding is discussed; various models of the growth of a grain boundary cavity are reviewed and further work necessary to resolve the unclarified mechanisms of fatigue failure at high temperatures is suggested.

1. INTRODUCTION

Fatigue has been considered to be the most insidious form of high temperature failure since cavitation can take place even in materials which resist creep cavitation[1]. However, the basic mechanisms at work in high temperature fatigue are still not very well understood. Attempts in the past were mainly confined to pure metals[2,3] and single phase materials[4] which are rather poor models for most engineering materials designed for high temperature applications. A number of basic questions related to fatigue cavitation still remain unresolved, and our intention here is to survey recent progress in this field. The role in fatigue cavitation of external environment and internal grain boundary environment (trace additions and residual elements) is highlighted.

2. GRAIN BOUNDARY MIGRATION AND SLIDING

A striking microstructural development in pure metals and single phase alloys at temperatures above 0.5 Tm, where Tm is the absolute melting temperature, is the formation of a diamond grain configuration (DGC), in which a majority of grain boundaries become aligned at about 45^{o} to the stress axis as a result of massive grain boundary migration (GBM)[5,6]. Though a large number of papers have dealt with this topic, the engineering significance of the formation of a DGC is rather limited. Materials designed for high temperature applications contain a fine dispersion of second phase particles, which have a strong pinning effect on GBM. (Fig. 1)

Fig. 1.Precipitate pinning grain boundary migration (Cu-Cr-Mg-Zr alloy fatigued at 400^{o}C)

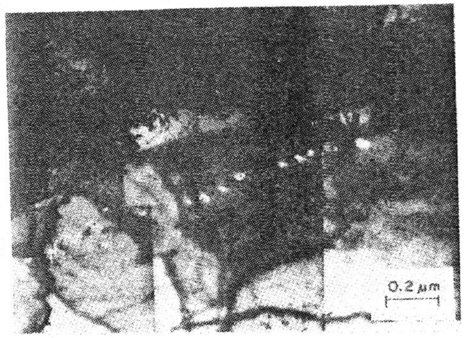

Fig. 2.Grain boundary migration leaving an array of cavities behind. (Same as Fig. 1)

It has been found[7] that in precipitate hardened materials noticeable GBM takes place only after about 50% fatigue life and the distances of migration are frequently less than 1 μm. Since at this stage grain boundary cavitation has already taken place, GBM tends to leave cavities behind (Fig. 2). These cavities, isolated within the grain, tend to shrink rather than continue to grow[8]. Also this limited GBM introduces boundary serrations (Fig. 3a,b). A similar observation was made by Saegusa and Weertman on pure copper[9]. However, in pure metals boundary serrations develop at an early stage of fatigue, thereby providing cavity nucleation sites[9,10]. In strong contrast to this, in precipitate hardened materials these serrations appear only at a later stage and are commonly free of cavities. Cavities are more frequently found at straight boundaries than at serrated ones[11]. Therefore, the massive GBM which occurs at an early stage of fatigue in single phase materials is harmful because it enhances GBS and therefore encourages grain boundary cavitation.[9,10,12] On the other hand, GBM in precipitate hardened materials which takes place only at a later stage of fatigue and is of a limited nature, seems to increase a material's fatigue resistance by isolating grain boundary cavities within grains and introducing grain boundary serrations thereby impeding grain boundary sliding and increasing the boundary's resistance to cavitation.

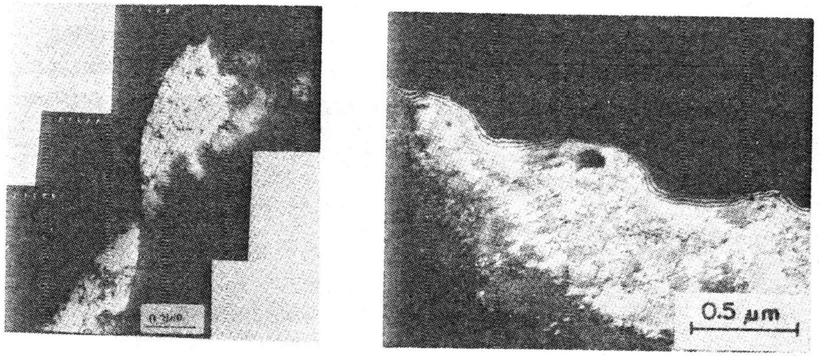

Fig. 3 Grain boundary migration resulted in boundary serrations. (a) Cu-0.8% Cr, fatigued at 400°C, $\Delta\varepsilon_p = 0.19\%$ $\nu = 10^{-2}$HZ, N_f = 64 cycles. (b) Cu-Cr-Mg-Zr alloy, 400°C, 10^{-1}HZ, $\Delta\varepsilon_p = 0.55\%$ N_f = 907 cycles.

Another phenomenon which is closely related to intergranular fracture in high temperature fatigue is grain boundary sliding (GBS). The importance of GBS in high temperature fatigue lies in the role it plays in intergranular fatigue failure. Evans and Skelton[13] reported that the average GBS offset at fracture was nearly independent of stress and microstructure. Similar observations were reported for Pb,[14] α - Zr and Zincaloy-2[15]. A recent investigation on a Cu-Cr-Mg-Zr alloy indicated that if grain boundary sliding was eliminated due to a retained cold worked structure, grain boundary cavitation was virtually absent; on the other hand, pronounced GBS and abundant cavitation occurred under similar test conditions when an equiaxial grain structure was produced by a heat treatment[16]. This indicates that GBS is a prerequisite for grain boundary cavitation.

Despite the fact that grain boundary sliding (GBS) plays a very important role in the development of a DGC and intergranular fatigue failure, virtually nothing is known about the process of reversed GBS. It has been shown that GBS does not represent an independent deformation mechanism, rather it must be accommodated by other deformation modes.[17] Fig. 4 shows the progressive increase of GBS offsets in fatigue tests of α-Fe,[12] Cu and Zr.[18] It is apparent that under different test conditions, the responses were quite different. When the plastic strain range was large, there was a sudden rapid increase in the GBS offsets at the start of the test. After that the increase of GBS offsets was approximately linear with the number of cycles. On the other hand, when the applied plastic strain range was small, the amount of sliding was a linear function of the number of cycles. The workers[12] attributed the rapid increase in the GBS offsets of α-Fe during the first few cycles to the pronounced GBM occurring within the first 5-10 cycles, but if this was the case, the same phenomenon would be expected for Cu and Zr as well! The difference in the responses probably stems from a difference in the dominant accommodation mechanisms for GBS. In the former dislocation glide in the matrix of the related grains was probably the dominant accommodation process, while in the later a diffusional accommodation process was probably dominant.

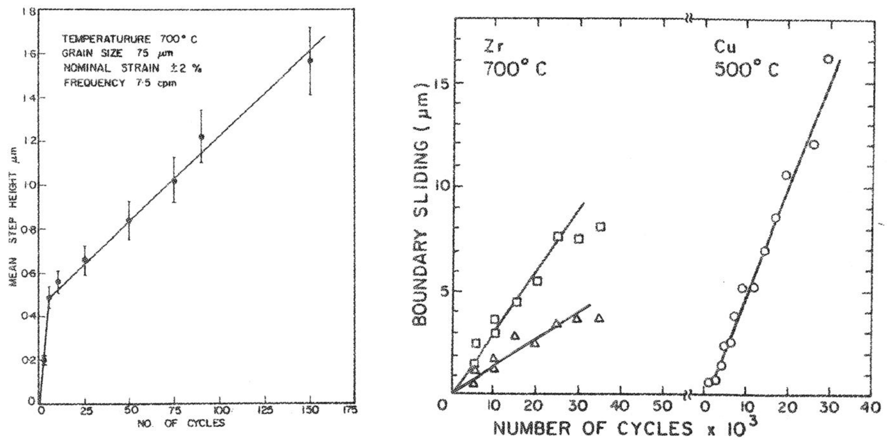

Fig. 4 (a) Vertical component of GBS versus number of cycle for α-Fe (from Ref. 12) (b) Vertical (Sv) and horizontal (Sw) component of GBS for Zr at 700°C and vertical component (Sv) for Cu at 500°C(from Ref.15)

Surface offsets due to GBS during fatigue of Cu-Cr based alloys at $400^{\circ}C$ in vacuum are shown in Fig. 5a,b. It is of interest to note that grain boundary precipitates were squeezed out due to GBS. This suggests that grain boundary precipitates strongly restrict GBS during fatigue.

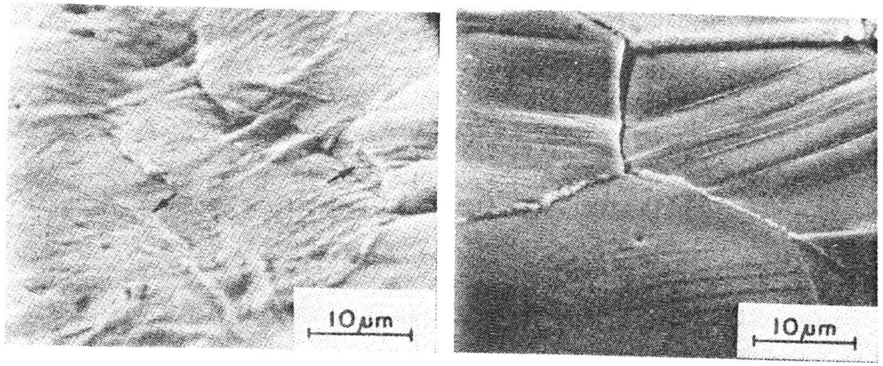

Fig. 5 Surface offsets due to GBS (a) Freshly exposed surface (Cu-Cr-Mg, $400^{\circ}C$, 10^{-3}HZ, $\Delta\varepsilon_p$ = 2%, N_f = 64 cycles)(b) Grain-boundary precipitates squeezed out (Cu-Cr-Mg-Zr, $400^{\circ}C$, $\Delta\varepsilon_p$ = 1.46% N_f = 269 cycles)

3. INTERGRANULAR FATIGUE FAILURE

It is known that the mode of fatigue fracture varies with temperatures from surface initiation and subsequent inward transgranular propagation at low temperatures to intergranular cracking and cavitation at high temperatures (i.e., at and above 0.4 Tm). The nucleation and growth of grain boundary cavities and wedge cracks were the main topics of three review articles.[15,19,20] Rather than repeat the material covered by these reviews, this paper attempts to give a critical assessment of the degree to which different models and theories of cavity nucleation and growth explain recent electron microscopic observations of cavity formation and growth during high temperature fatigue.

Though the pioneers in developing a theory for fatigue cavitation were H. D. Williams[21] and R. Broomfield[22], much credit should be given to R. P. Skelton[23], who further developed the Hull and Rimmer[24] model for diffusional growth of cavities in creep and made it account for fatigue cavitation. Realizing that the grain boundary self-diffusion coefficient

varied with the applied stress, Skelton concluded that the critical radius for void growth was given by

$$r_{crit} \geqslant \frac{4\,\gamma\,KT}{\sigma^2\Omega} \qquad (2)$$

Where σ is the stress amplitude, Ω is the atomic volume, γ is the surface energy and K and T have the conventional meanings. This is an elegant development! However, a simple calculation based on equation (2) indicated that r_{crit}, was very much larger than many observed cavities, and Skelton concluded that 'it is unlikely that a growth condition can be met solely by irreversible vacancy flow in the grain boundary'. He then considered the accelerating effect of an increased vacancy concentration in the lattice due to cyclic stressing, on cavity growth. This gave a second expression for r_{crit}

$$r_{crit} = \frac{2.5 \times 10^3\,\gamma\,\Omega\,\nu}{nKT\varepsilon f}\;\frac{Dg_b^{\rho}\;\delta}{Do\;\;a} + 2\exp\left(\frac{-Q_{sd}}{KT}\right) \qquad (3)$$

where ν is Debye frequency, n is the number of jumps made by a vacancy before its annihilation, ε is the plastic strain amplitude, f is frequency, Dg_b^{ρ} is the grain boundary diffusion coefficient, δ is the effective grain boundary width, Do is the bulk diffusion coefficient, Qsd is the activation energy for bulk self diffusion, a is the cavity spacing, and K, T, γ and Ω have the same meanings as in equation (2). From equation (3), the critical radius for magnesium was found to be 0.13 µm. It should be mentioned that some features of this model are open to question. Firstly, Skelton believed that only above a certain critical fatigue strain rate could vacancies generated during fatigue be accepted by the cavities. Although he claimed the existence of such a critical strain rate for magnesium, this conflicts with most experimental observations that show low strain rates encourage intergranular cavitation. Secondly, in using the equation developed by Lomer and Cottrell[25] to find the number of jumps made by a vacancy during its lifetime, he assumed that the activation energy for vacancy migration was equal to the activation energy for lattice self-diffusion, when in fact, vacancy migration requires only about one half of the activation energy for lattice self-diffusion.[26] Finally, he considered that the vacancy concentration generated during fatigue was equal to that during creep. However, the expansion and contraction of dislocation rings during fatigue is a prolific source of vacancies, and consequently vacancy concentration during fatigue is probably an order of magnitude higher than that during creep[27] which partly accounts for the extremely rapid cavity growth observed during high temperature fatigue[28].

Eighteen years later, Weertman[29] used a stress concentration factor introduced by Raj and Ashby[30] to resolve some of the ambiguities in the theory of cavitation initiation during high temperature fatigue. Raj and Ashby found that a stress concentration factor $2\lambda/\pi L$, where λ is the wavelength and L the amplitude of the boundary 'roughness' which is assumed to be sinusoidal, existed at the undulating cavity boundary. Weertman argues that this stress concentration reduces the critical cavity radius by a factor of $\left(\frac{\pi L}{2\lambda}\right)^2$. In his comprehensive treatment of cavity growth he concluded that under cyclic load the critical radius for a cavity to grow is given by the first equation derived by Skelton:

$$r_{crit} > \frac{4\,\gamma\,KT}{\sigma^2\,\Omega}$$

but that the large stress concentration on a sliding boundary allowed cavities with sub-critical radii to grow. He further pointed out that in most cases the contribution of volume diffusion to cavity growth is relatively unimportant and the cavity growth rate is the same as that when grain boundary diffusion acts alone. He suggested that the relative importance of the contribution to cavity growth of grain boundary diffusion as compared to that of bulk self-diffusion was dependent on the inter-cavity spacing and temperature through the ratio L/Λ, where L is the half distance between cavities, and $\Lambda = \delta Dgb/2Db$, where δ is the effective grain boundary width, and Dgb and Db are grain boundary and grain diffusivities respectively. Only when $\frac{L}{\Lambda} \gg 1$, will the transport of vacancies through the grain interior approach that through the grain boundary. Take copper as an example, $Dgb/Db \sim 10^7$ at 400°C, and $\sim 10^4$ at the melting point. Therefore, at 400°C grain self-diffusion becomes important only when the cavities are several millimeters apart!

In fatigue tests of copper at 400°C, Williams discovered that cavitation took place in a test where the applied stress was fully compressive[31]. Based on this observation, he dismissed grain boundary sliding mechanisms for cavity growth. His reasoning was simply that if GBS is the operating mechanism for cavity growth there ought to be a similarity of cavitation behaviour in compression creep and fatigue, but voids were

abundant in fatigue but absent in creep. He circumvented this difficulty by advancing a diffusion hypothesis, and found that the cavity growth rate was proportional to the plastic strain amplitude. Since both tensile and compressive fatigue involve cyclic plastic strain, cavitation is then possible for any test condition. He believed that the Hull and Rimmer treatment of creep cavitation, in which cavity growth is restricted to boundaries experiencing a component of tensile stress, could not account for cavitation in fully compressive fatigue. In fact, plastic deformation is mainly localized in the vicinity of grain boundaries during high temperature fatigue[2,31], and a local tensile residual stress can develop in the neighbourhood of related grain boundaries during unloading from compression in the same way that a compressive residual stress is developed in a notch or crack tip during unloading from tension. Therefore, the Hull and Rimmer treatment may also explain his mysterious discovery.

Using a sensitive density change technique, Gittins[28] successfully demonstrated: 1) The fractional change in density of a fatigued copper specimen, $\frac{\Delta D}{D}$, is directly proportional to time for constant plastic strain range tests; 2) for constant stress tests, when the stress is sufficient to cause appreciable hardening, $\frac{\Delta D}{D} \propto t^{2/3}$; 3) the activation energy for cavity growth is close to that for grain boundary diffusion and vacancy migration. He then proposed that after grain boundaries assumed a DGC, GBM reached a dynamic steady state, i.e., at any point in a given boundary, migration to and fro would occur as the local defect balance altered periodically, these moving boundaries would annihilate vacancies and dislocations, and the dilatation associated with these line and point defects would then contribute to cavity growth. He estimated that the order of magnitude of the total rate of change of $\frac{\Delta D}{D}$ caused by these defects was $\sim 2 \times 10^{-5}\ N_L \rho V$, where $2\ N_L$ is the boundary area per cm^3, ρ is dislocation density and V is the mean rate of GBM. Gittins claimed that satisfactory agreement between the calculated and observed cavity growth rate was obtained.

There is some experimental evidence unfavourable to this model. It has been observed that a cavity has a strong pinning effect on the migration of a grain boundary (Fig. 6). A heavily cavitated grain facet is probably immovable. As Gittins himself admitted, the reduction in the free energy for GBM on extensively cavitated facets may reach

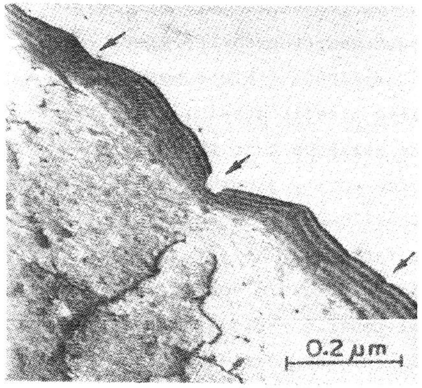

Fig.6. Cavities (arrowed) pinning grain boundary migration. Note the smallest cavity is less than 50Å in diameter. (Cu-Cr-Mg-Zr , 400 °C, 10^{-1} HZ, $\Delta\varepsilon p$ = 0.50%, interrupted at 100 cycles.)

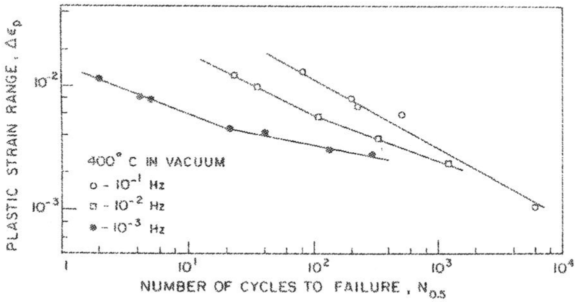

Fig.7. The effect of frequency on low cycle fatigue life of a Cu – 0.8% Cr alloy at 400°C in vacuum

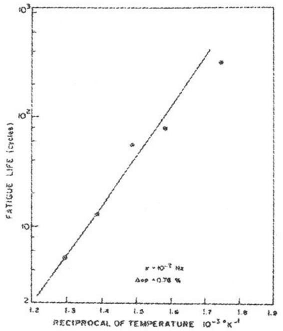

Fig.8. Fatigue life of the Cu-Cr alloy at various temperatures.

∿ 3 x 10^6 dynes cm^{-2}, and the driving force for GBM is estimated to be only 2 ∿ 3 x 10^6 dynes cm^{-2}. Grain boundary migration tends to leave cavities behind (Fig. 2) rather than to drag them on. These cavities, isolated within the grain, tend to shrink ratner than continue to grow.[8] Westwood[32] found that a perceptible density change in iron specimens occurred only after GBM was completed, which is consistent with the above observation.

Recently Beere[33] used theories developed for creep cavitation to calculate the growth rate of a grain boundary cavity during fatigue at low strain rates and high temperatures. He considers three mechanisms for cavity growth, i.e., plastic growth, grain boundary diffusion and diffusional growth controlled by the plastic accommodation strains in the surrounding matrix. Since equations developed for creep cavitation are used without modification, cavity growth from the last two mechanisms is only expected in fatigue with a tensile hold time. The number of cycles to failure N_f due to cavity growth is predicted to be inversely proportional to $\Delta\varepsilon_p'$, the elastic strain converted to plastic strain during the tensile holdtime.

$$N_f \Delta\varepsilon_p' = \frac{L^3}{24} - \frac{\alpha\,\beta}{\alpha+\beta} \quad \Delta\varepsilon_p \quad \frac{\alpha+\beta}{\beta^2}$$

$$\alpha = \frac{Dgb\ \delta\ \Omega\ \sigma_m}{KT\ q\ \dot{\varepsilon}}$$

$$\beta = \frac{L^2 d}{4}$$

Where $\Delta\varepsilon p$ is the plastic strain range, $\dot{\varepsilon}$ is the strain rate, σ_m is the maximum stress, d is the grain size, $q = \log_e \frac{L}{a} - \frac{1}{4}\left(3 - \frac{a^2}{L^2}\right)\left(1 - \frac{a^2}{L^2}\right)$, the other symbols are as defined previously. The limitation of Beere's model is rather obvious: it does not account for cavitation due to a symmetric strain rate isothermal fatigue cycle without a tensile hold time, also it predicts that at low strain rates failure will be due to crack growth rather than to cavitation. This seems to be contradictory to most experimental observations.

Wigmore and Smith[34] proposed a model in which a combination of GBS and plastic deformation at the crack-like cavity tip leads to cavity growth with ripple markings on the freshly created surfaces. No theoretical analysis of the cavity growth rate was made by these authors.

Recent work at Waterloo on cavity growth during fatigue strongly suggests that different mechanisms are involved in different temperature-plastic strain range-strain rate regions[11]. A commercial Cu-0.8% Cr alloy was found to contain a large number of small voids ($\sim$ 0.1 μm diameter) at grain boundaries in its as-received condition. These were thought to be due to gas bubble (H_2O) precipitation during processing[42]. Low cycle fatigue tests were carried out on this alloy at 400^{o}C in vacuum. The results are shown in Fig. 7. It can be seen from this diagram that when the plastic strain range is large the fatigue lives of this alloy are inversely proportional to frequency, but the time to fracture is almost constant. On the other hand, when the plastic strain range is small, the frequency dependence of the fatigue life is much weaker. The fatigue fracture behaviour was such that this alloy always failed intergranularly, apparently from the growth and interlinkage of these pre-existing bubbles. The fatigue lives for the Cu-Cr alloy at $\Delta\varepsilon p$ = 0.78% for temperatures between 300 and 500^{o}C are plotted in an Arrhenius plot in Fig. 8. The data approximately falls onto a straight line whose slope corresponds to an apparent activation energy of 83 KJ/mole. This value is close to the activation energy for surface diffusion and grain boundary diffusion in pure copper[35]. Based on this experimental evidence a model shown in Fig. 9 is advanced. At higher plastic strain ranges grain boundary sliding produces sufficient newly separated surface, and a combination of surface diffusion (to modify the cavity shape) and grain boundary diffusion results in an increase in cavity size. Diffusional mass transport is rate controlling. If the failure criterion is such that the average radius of cavities reaches a critical value, then at a constant plastic strain range, the time to failure is constant, but the fatigue life (number of cycles to failure) would be inversely proportional to frequency. On the other hand at lower plastic strain ranges the grain boundary sliding rate was limited by the existence of grain boundary precipitates, consequently the cavity growth rate would become grain boundary sliding rate controlled. It would be expected that the GBS distance was mainly cycle-dependent, therefore in the low plastic strain range region the frequency effect is reduced.

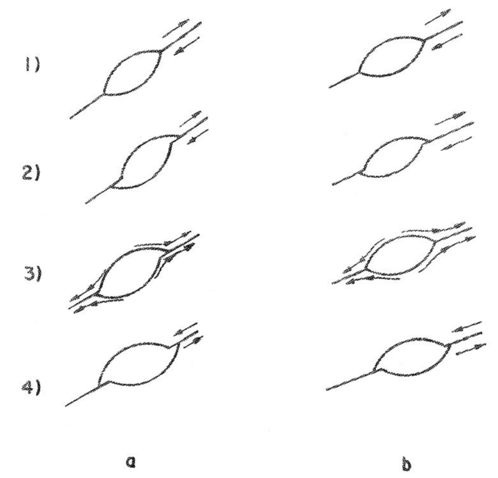

Fig.9. Fatigue cavity growth model. The driving force for cavity growth here is the high surface tension of a cavity due to the large curvature of the newly created surface.

a: At high strain range where diffusion is the rate-limiting process, rewelding of the newly created surfaces takes place upon reversal of the sliding.

b: At low strain range where grain boundary sliding is the rate-limiting process.

From this review it can be seen that a few fundamental questions remain unresolved.

(1) Does matrix plastic flow contribute to cavity growth?

(2) Does surface diffusion play a role in the cavitation process?

(3) What is the effect of the compressive half cycle on cavity growth?

(4) How does an aggressive environment affect fatigue cavitation?

4. ENVIRONMENTAL EFFECT

At high temperatures an oxidizing atmosphere may exert a marked influence on fatigue behaviour. There is no shortage of data indicating that oxidation may direct the crack propagation path to grain boundaries, and therefore considerably shorten the fatigue life of engineering materials. McMahon and Coffin[36] suggested that fatigue crack initiation was preceded by oxidation along grain boundaries. Fleetwood et al[37] also noted that local oxidation at grain boundaries significantly affects crack nucleation in the thermal fatigue of cast nickel-base alloys. Taplin et al[38] recently found that interfacial cavities existed in a Cu-Cr based alloy fatigue tested in air but not in vacuum. It is worthwhile mentioning that oxygen attack not only enhances cavitation, but also introduces some new cavity nucleation mechanism during creep of nickel and nickel-base alloys.[39] Recently, Bricknell and Woodford found that the existence of residual carbon (30 ~400 ppm) in nickel enhanced intergranular cavitation.[40,41] They further discovered that nickel's creep ductility could be improved very dramatically by reducing its carbon level. Apparently, carbon reacts with oxygen to form gas bubbles of CO_2 at grain boundaries and provides cavity nuclei. Does this mechanism of cavity nucleation also operate in fatigue? This is an interesting question to follow-up.

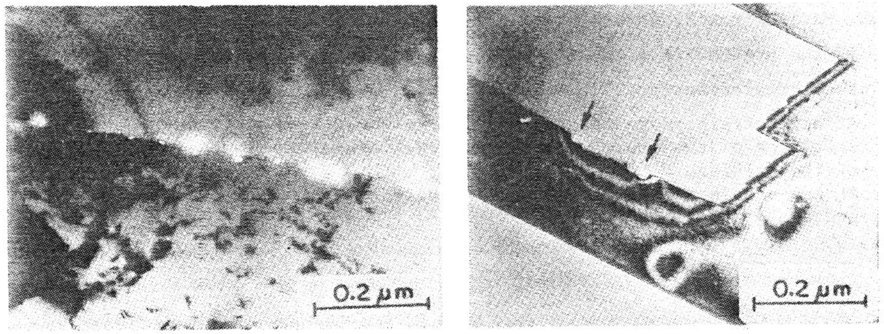

Fig.10.a.Cavities at a grain boundary.
b.Cavities(arrowed)at a coherent twin boundary (Cu-Cr-Mg-Zr,same as Fig.6)

A serious limitation of previous observations on fatigue cavitation is that most of them are confined to optical microscopy. The smallest cavity detectable by this method is about 0.5 µm, but cavities as small as 50 Å were observed under transmission electron microscopy (Fig.10a.) This probably indicated that residual gas diffused into cavities and stablized these cavities by reducing the tendency for sintering. It was also observed that cavities could nucleate at coherent twin boundaries at an early stage of fatigue (Fig.10b.)

Another problem is that work in the past focused mainly on model metals, (well annealed pure or single phase materials). Engineering materials designed for high temperature applications are rather complicated.

It has been found that small additions of reactive elements (Hf, Zr are particularly effective) can strongly affect the grain boundary structure and chemistry and retard cavitation[42,43]; a retained cold worked structure can totally suppress cavitation,[38] and grain boundary precipitates can hinder GBS and GBM and therefore play an important role in fatigue cavitation.

5. CONCLUSIONS

A few conclusions can be drawn from this review. These are:

1. It is generally accepted that GBS is a critical step for fatigue cavitation, especially the cavity nucleation stage.

2. The critical radius of a cavity above which it can grow is given by

$$r_{crit} = \frac{4\ \gamma\ KT}{\sigma^2\ \Omega}$$

3. Though the whole picture of cavity growth is still not clear, it seems that different mechanisms are involved in different temperature-plastic strain range-strain rate regions. This point deserves further investigation.

4. There is a dearth of information on detailed microstructural changes during high temperature fatigue, especially neglected is the electron-microscopic observation of fatigue cavitation.

5. At high temperatures, environment plays an important role in the fatigue cavitation process. Its effect should be clarified.

6. The role in GBM and GBS of grain boundary dislocations (both intrinsic and extrinsic) is not known for fatigue. Work in this line would be of interest to clarify the fundamental mechanisms involved.

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