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HIGH TEMPERATURE CREEP-FATIGUE FRACTURE IN Cu-Cr ALLOYS

A. L. W. Collins and D. M. R. Taplin*

INTRODUCTION

The alloy Cu/1% Cr combines moderately high strength (400 MPa tensile strength at 475K) with resistance to softening at temperatures up to 773K, allowing brazing without loss of strength. However, in service at temperatures as low as 330K it is subject to intergranular creep fracture [1]. A recently developed alloy (AMAZ-MZC)† Cu/0.5% Cr/0.1% Zr/0.03% Mg shows greater resistance to softening and creep deformation [2]. A previous investigation of the problem of hot intergranular fracture in low-alloy coppers showed a clear division of the alloys into those showing early intergranular fracture at stresses in the order of half the yield stress, and those showing no fracture below 80% of the yield stress within three months [1,3]. This makes direct comparison of the mechanisms of fracture difficult, and therefore it was considered useful to investigate the low-cycle fatigue properties of the two alloys identified above, at elevated temperatures.

EXPERIMENTAL

The alloys were received in the form of 15mm rods, having been solution heat treated, cold worked 40% and aged for two hours at 750K. The alloy CuCr had a relatively large grain size of 100μm diameter, equiaxed, and contained visible inclusions up to approximately 1μm in diameter. The alloy CuCrZrMg had very elongated grains, only 10μm in the transverse direction, but up to 1mm in the axis of the rod. The grain boundaries were rather weakly defined, and could not be differentiated by any etchant except ammoniacal ammonium persulphate. The inclusions seen in this alloy were smaller than in CuCr. Low cycle fatigue tests were conducted in strain control, at zero mean strain on an Instron machine, using specimens with 6mm gauge diameter and 7.5mm gauge length. In elevated temperature tests, the temperature of the gauge length was controlled to $\pm$3K using a resistance furnace and closed loop control. Tests were conducted at temperatures up to 673K and frequencies from 1.7 to 170mHz. The fracture surfaces were examined under the optical microscope. Tensile tests were also performed using strain rates which would give the same cyclic frequency if they were considered fatigue tests at 0.25 cycle duration, and the true strain to fracture measured.

RESULTS AND DISCUSSION

The fatigue life at each temperature was plotted logarithmically against the ratio of plastic strain range to the tensile fracture strain as shown in Figures 1 and 2. It is clear that there is a strong correlation between

*Department of Mechanical Engineering, University of Waterloo, Waterloo, Ontario, Canada, N2L 3G1.
†Trade Mark of AMAX Base Metals Inc.

ductility and the plastic strain range needed to cause fatigue failure in alloy CuCr as the ductility decreases sharply with decreasing strain rate and increasing temperature. In contrast, the ductility of alloy CuCrZrMg is virtually constant, while the fatigue life at 673K and 1.7mHz shows a decrease for the 300K results, which becomes more pronounced at longer lives. This behaviour is very similar to that reported for steels and superalloys by Coffin [11]. Numerous surface cracks were visible in the specimens with reduced life as shown in Figure 4, looking very similar to the oxidation - accelerated transgranular cracks observed by Skelton [12] in 1/2 Cr-Mo-V steel at 825K. Scanning electron microscopy shows quite clearly the difference in the striation appearance between the low temperature and high temperature fractures. The low temperature fractures, (Figure 5), show "wavy" striations with a dimpled surface and no visible branching of the cracks while the high-temperature low-frequency striations (Figure 6) are straight with flat surfaces between them and pronounced lateral cracking. All these observations may be explained by considering the reduction in fatigue life to be due to oxidation. The oxide cracking never extended across the entire specimen, probably because the presence of a crack more than 100μm deep would concentrate the plastic deformation into the plane of the crack. This sould increase the plastic strain range to a level where continuum crack propagation is faster than oxidation processes, since the processes appear to have strain dependence as approximately:

$$\frac{da}{dN} \alpha \Delta\varepsilon_p^{\ 2} a \qquad \text{(continuum)} \qquad (1)$$

$$\frac{da}{dN} \alpha \Delta\varepsilon_p \ a \qquad \text{(oxidation)} \qquad (2)$$

according to the data of Coffin [11] and Solomon [13,14]. Thus, for any temperature and frequency, a transition strain exists, above which continuum processes dominate.

The concentration of plastic strain in tests where the longitudinal plastic extension range is measured makes any conclusions based on cracks occupying more than perhaps one tenth of the specimen diameter rather dubious. For such conditions with gross-plasticity cycling, more sophisticated instrumentation, based on diametral extensometry, is necessary. This does not significantly affect the validity of endurance tests since a detectable decrease in load due to strain concentration at a crack rarely occurs before 90% of the endurance life has passed.

The behaviour of the alloy CuCr is more surprising. Figure 1 shows that the fatigue life at both temperatures and frequencies fits the equation:

$$\frac{\Delta\varepsilon_p}{\varepsilon_f} N_f^{\ 0.46} = 0.6 \qquad (3)$$

where ε_f is the tensile ductility measurement at the same temperature and half-cycle duration. The results on alloy CuCrZrMg indicate that transgranular fracture strains are independent of temperature and strain rate so that the 300K data may be considered to be equivalent to very fast tests at 673K, where fracture would presumably be transgranular.

The expression of equation (3) is of the form originally proposed by Tavernelli and Coffin for transgranular low-cycle fatigue [15] but Manson [16] showed that a better fit could be obtained to:

$$\Delta\varepsilon_p \ \frac{N_f^{\ \beta}}{\varepsilon_f} = 1 \qquad (4)$$

which gives a very poor fit to the CuCr data. This difference must be accounted for by the difference in fracture mechanism.

Whatever the path of the fracture, it is normally considered that the Manson-Coffin law is related to a crack propagation law of the form:

$$\frac{da}{dN} = \frac{\Delta\varepsilon_p^{\ 1/\beta} f(a)}{\varepsilon_f} \qquad (5)$$

For equation (4) to be satisfied, C = 1. A very serious problem justifying this equation is that extrapolation of the measured crack growth rate to earlier than half-life shows that the distance of crack advance is much less than the average spacing of inclusions. These inclusions have been unequivocally shown to control the tensile fracture strain (6). Thus, over a distance equal to the average crack advance, the macroscopic ductility is meaningless. It may be more useful to consider the volume fraction of inclusions, since it has been shown (6) that, for weakly bonded inclusions, the ductility is controlled largely by the inclusion volume fraction V_f. The realtion is approximately

$$\varepsilon_f V_f = \text{constant} \qquad (6)$$

This equation (6) would become

$$\Delta\varepsilon_p \ (N_f V_f)^{\beta} = \text{constant} \qquad (7)$$

If intergranular fracture is considered, ductility is no longer controlled by the inclusion volume fraction. Voids nucleate on grain boundaries, and grow in area according to the equations (5-7).

$$A = A_o + B(\sigma^m t)^r \qquad (8)$$

where B is a function of temperature, m is between 1 and 3, and r is between 1/2 and 2/3. In fatigue the cavities are reduced during the compressive half cycle, and in low-cycle fatigue, the stress is only weakly dependent on strain range

$$\Delta\sigma = \sigma_o \ \Delta\varepsilon_p^{\ n'} \qquad (9)$$

with n' = 0.12 in the present case.

Thus, the area occupied by voids at the end of the tensile half cycle is controlled mainly by the cyclic frequency. This area fraction is related to the spacing of voids on the crack front in the same way as the volume fraction of voids in transgranular progagation, suggesting a fatigue law of the form

$$\Delta\varepsilon_p (N_f A_f)^{\beta'} = \text{constant} \qquad (10)$$

Tensile ductility may be related to the area fraction of voids by using Thomason's model for fibrous fracture [8], which appears to be valid for two dimensions.

$$\varepsilon_f A_f^{12} = \text{constant} \tag{11}$$

Thus, equation (10) becomes

$$\Delta\varepsilon_p \frac{N_f}{\varepsilon_f^2}^{\beta'} = \text{constant} \tag{12}$$

A_f is slightly strain dependent so that β' is slightly less than β:

$$\frac{1}{\beta'} \quad \nu \rightarrow 0 \; \frac{1}{\beta} + \text{mr n}' \tag{13}$$

Since $\beta' = 0.46$, equation (7) is very close to equation (1); equation (7) contains $\varepsilon_f^{0.92}$, instead of $\varepsilon_f^{1.0}$, a discrepancy too subtle to detect with the present data.

The reasons for the absence of intergranular fracture in alloy CuCrZrMg have not been completely determined. Similar effects have been reported for the addition of Zr, Mg, Cd and Ti to copper-chromium, and for the addition Mg to Ni-Al_2O_3, in the slowing or elimination of creep cavitation. There are two possible explanations. The first is that a segregated solute reduces the diffusion coefficient (9) of grain boundaries and the surfaces of voids, causing a marked decrease in the growth rate of voids. This could explain the reported effects of Mg and Cd, but could not explain the absence of intergranular voids in the creep of CuCrZrMg.

The second effect is that a zirconium solute segregates to grain boundaries and precipitates during aging as Cu_3Zr or Cu_5Zr, producing a very large area fraction of precipitates (10). This decreases the stress concentration produced by grain boundary sliding on any single precipitate and therefore prevents the initiation of voids. This structure also decreases the sliding rate, but this effect, like that of void growth could only explain a change in cavitation rate, not an elimination of cavitation.

The tests described in this paper show effects rather different from those described in other papers presented to this Conference. The data of Scarlin [17] on nickel-base alloys shows acceleration of crack growth at low frequency in vacuum, but suggest that crack branching, similar to that seen in the present CuCr data, reduces this acceleration. It is not clear whether a similar effect exists in wrought copper alloys, since the branching appears to be dependent on the structure of the cast nickel-base alloy.

The creep-fatigue interaction effects in iron-base alloys [18-20] are explained largely by changes in the deformation behaviour of the materials. Intergranular fracture only occurs at low stress level in steels, so that it would not be expected that grain-boundary behaviour would affect fatigue crack propagation at the high strain rates used, in sequential tests. Coffin's paper [11] makes it clear that interaction of the fracture processes, as such, does occur in steels when the creep and fatigue damage is incurred almost simultaneously by cycling with tensile hold-periods.

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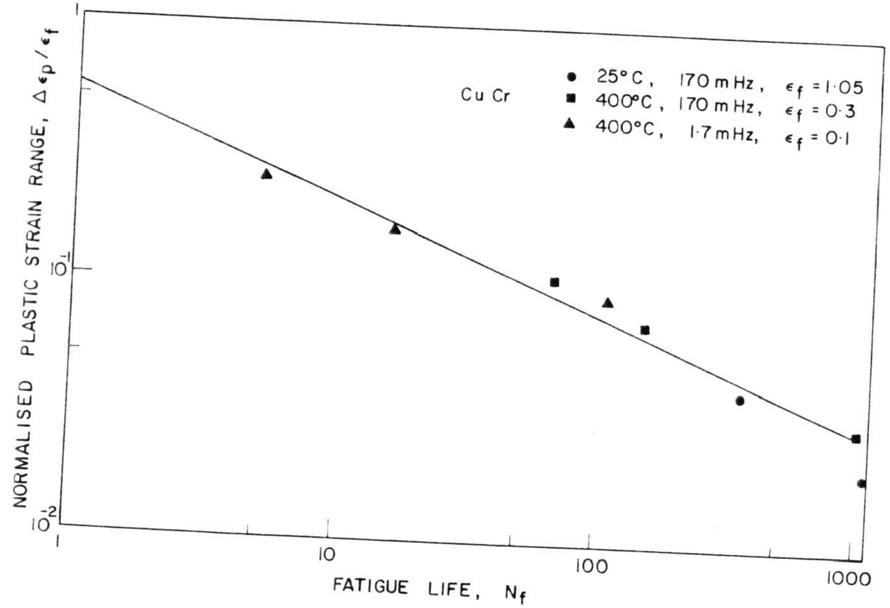

Figure 1 Variation of Fatigue Life with Plastic Strain Normalized by Tensile Fracture Strain. Alloy CuCr

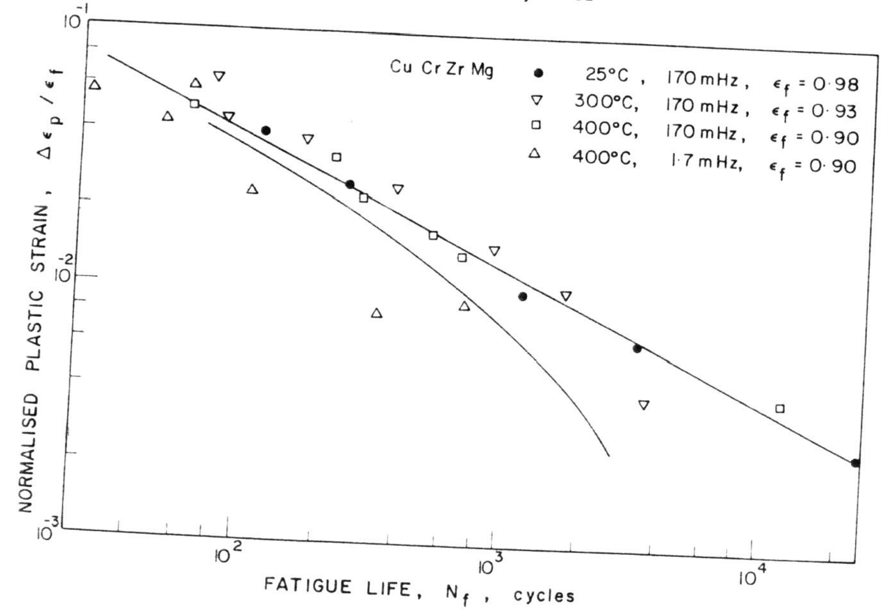

Figure 2 Variation of Fatigue Life with Plastic Strain Normalized by Tensile Fracture Strain. Alloy CuCrZrMg

Figure 3 Optical Micrograph of Fracture Surface of Alloy CuCr, Fatigued at 673 K and 1.7 mHz x 65

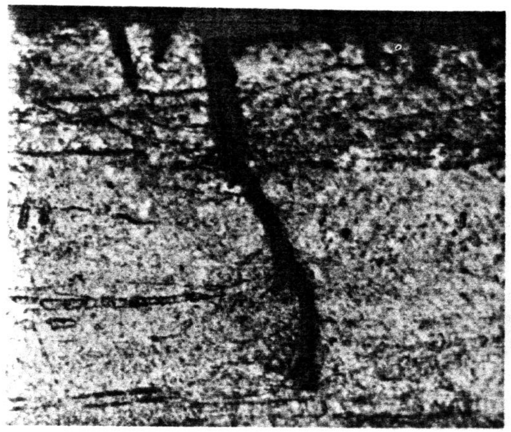

Figure 4 Optical Micrograph of Free Surface of Alloy CuCrZrMg, Fatigued at 673 K and 1.7 mHz Showing Massive Surface Cracking

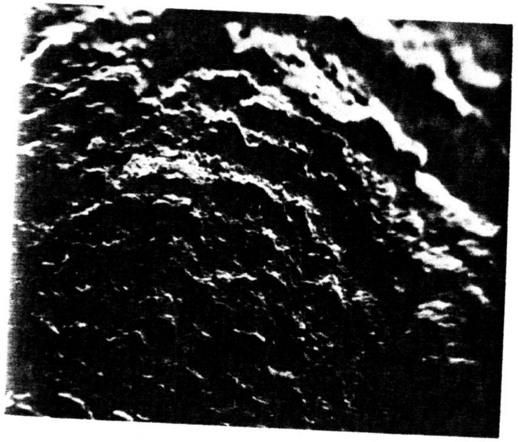

Figure 5 Scanning Electron Micrograph Showing Ductile Striations in Centre of CuCrZrMg Specimen, Fatigued at 673 K and 1.7 mHz. x 50.

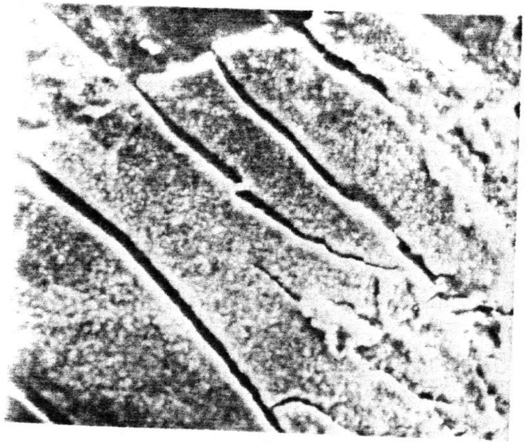

Figure 6 Scanning Electron Micrograph Showing Brittle Corrosion-Fatigue Striations near Start of Crack in the Specimen of Figure 5. X 2000

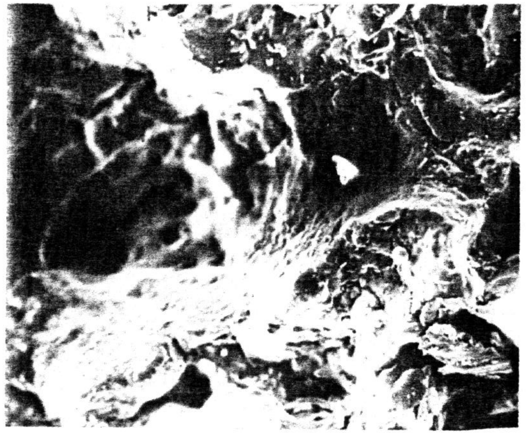

Figure 7 Scanning Electron Micrograph Showing Intergranular Cracking and Large Void in Alloy CuCr Fatigued at 673 K and 1.7 mHz. X 200