FATIGUE CRACK NUCLEATION AND EARLY GROWTH IN METAL SINGLE AND POLYCRYSTALS

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Single crystals and polycrystals of copper were cyclically strained with constant plastic strain amplitudes and mechanical behaviour, structure and surface evolution were studied. Cyclic slip localization can be detected using loop shape parameter changes. The surface relief evolution and crack nucleation, crack growth and formation of a dominant crack in single and polycrystals were discussed. The early stages of relief evolution were found to be similar. The crack growth rate of short cracks was plotted vs the stress intensity amplitude. The approximate analytic relation was integrated to obtain the fatigue life prediction.

INTRODUCTION

The nucleation and growth of fatigue cracks plays a prominent role in the study of the fatigue failure of materials. Recently, simple metallic materials subjected to well defined loading conditions have been studied in order to understand the mechanism of fatigue damage (Mughrabi (1), Wittmer et al. (2) and Polák et al. (3)). The surface observations and transmission electron microscopy study of a copper single crystal (Polák et al. (4), Laird et al. (5), Polák and Obrtlík (6)) revealed the relation of the mechanical behaviour, structure and surface evolution in cyclic straining. Several new mechanisms of fatigue crack nucleation were proposed (Essmann et al. (7), Polák (8) and Jackson (9)). These proposals were compared with the observations on single and polycrystals ((8), Polák and Liškutín (10)).

Recently the strain localization in polycrystals has been also detected using mechanical measurements. It was found to be less prominent than in single crystals (2,3) but of a similar character. It also results in stress amplitude saturation and

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crack nucleation in strain cycling. In this contribution the behaviour of single and polycrystalline copper in cyclic straining is compared and fatigue crack nucleation and early growth is discussed.

CYCLIC SLIP LOCALIZATION

The period of early hardening in cyclically strained materials is followed by cyclic slip localization. The slip localization can be detected by the measurement of the hysteresis loop shape. Figure 1 shows the plot of the shear stress amplitude τ_a and loop shape parameter V_H in constant plastic strain amplitude cycling of copper single crystal. The onset of the strain localization can be identified with the minimum of the loop shape parameter V_H. When V_H increases growing number of distinct slip markings appear on the surface of the crystal, ladder-like structure of individual PSBs and cell and ladder structure of the macro-PSBs characterizes the internal structure and the stress amplitude drops from its peak value and reaches saturation.

Figure 2 shows analogous dependence for polycrystalline copper cycled with the plastic strain amplitude which results in approximately the same life as the fatigue life of the single crystal. The comparison of Figs. 1 and 2 shows that not only fatigue lives but also characteristic cycle numbers where V_H reaches minimum, N_n, and maximum, N_s, are close in both specimens. At the same time the ratio of the γ_{ap} in single crystal to ε_{ap} in polycrystal is equal to 7, which is considerably higher than the Sachs' factor for f.c.c. lattice, which is 2.24. It shows that the plastic strain is not distributed homogeneously over all grains of a polycrystal. The distribution of plastic strains is difficult to estimate, however, the amplification factor is about 3.

CRACK NUCLEATION

The intensive plastic shear strain which is localized into PSBs both in single and polycrystals leads to the progressive formation of the characteristic surface relief. Figure 3 shows the early surface relief adjacent to an embryonic PSB in a single crystal and Fig. 4 in a polycrystal. This comparison shows that in both cases a semiregular relief consisting of alternating extrusions and intrusions is observed. Therefore, if the plastic strain amplitude in a polycrystal is low enough, mostly single slip conditions govern the deformation in a particular grain and both the early surface relief and the internal dislocation structure are similar.

Further cyclic straining results both in a single crystal and within a grain of a polycrystal in the growth of extrusions and deepening of intrusions. The mechanism of the surface relief formation has been discussed earlier (8) and considers point defect migration, mass transport and dislocation assisted intrusion and extrusion formation. The primary shallow surface crack nucleation results either from the interconnection of the closely spaced intrusions or from the irreversible environment assisted slip at the tip of a sharp intrusion (10).

An important difference affecting the formation of a dominant crack in a single crystal and a polycrystal is the presence of grain boundaries. Contrary to the single crystal in which only nucleation along the PSBs was observed, additional crack nucleation sites are present in polycrystals, i.e. grain and twin boundaries. Numerous cracks in PSBs parallel to the primary slip plane arise in a single crystal before a macroscopic crack is formed by interconnection of several parallel cracks due to nucleation of the cracks along the secondary systems in the bridges between the primary cracks. This situation is illustrated in Fig. 5. In polycrystalline copper cycled in elasto-plastic region numerous cracks arise both in PSBs and along grain or twin boundaries which also become persistent (10). The density of the nucleated cracks increases considerably with the applied plastic strain amplitude.

CRACK GROWTH

The most plausible mechanism of the growth of the shallow primary fatigue cracks is the slip-unslip mechanism. The crack grows originally in the primary slip plane and at a later stage deviates in the plane normal to the specimen axis. In a polycrystal several cracks grow independently and create an anisotropic plastic zone ahead its tip. Due to linking with smaller cracks the longest crack need not be the same up to fracture. In order to characterize the crack growth rate in a material during the whole fatigue life the concept of an 'equivalent crack' has been introduced (10). The equivalent crack is the crack whose length is close to the temporary longest crack and its rate characterizes the crack growth rate of a material at the particular stress or plastic strain amplitude.

Figure 6 shows the crack growth rates of equivalent short cracks at several plastic strain amplitudes versus the stress intensity factor amplitude simultaneously with the curve corresponding to long cracks. The initial crack growth rate at each amplitude is approximately constant, then increases and approaches the long crack growth rate. Assuming the validity of the Paris law for long cracks a reasonable approximation of the crack growth rate as the sum of two terms can be given

$$da/dN = (da/dN)_i \, [\, 1 + (K_a/K_{ai})^{\beta} \,] . \tag{1}$$

The initial crack growth rate $(da/dN)_i$ is independent of the crack length and depends on the plastic strain amplitude. The constants in the Paris law β and $(da/dN)_i/K_{ai}$ are parameters characterizing the material.

The fatigue life of a smooth specimen can be calculated by integration of eqn. (1). Provided the cracks grow as semicircular cracks of a diameter a up to the fatal crack of a diameter a_F we can use the approximate expression for stress intensity amplitude

$$K_a = 0.65\ \sigma_a \sqrt{a}\ . \qquad (2)$$

Inserting eqn. (2) into eqn. (1) we get

$$da/dN = (da/dN)_i\ [\ 1 + (a/a_i)^{\beta/2}\ ] \qquad (3)$$

where

$$a_i = (K_{ai}/0.65\ \sigma_a)^2 \qquad (4)$$

is the characteristic crack length which depends slightly on the stress amplitude.

Equation (3) can be integrated and for $\beta = 2$, 4 and 6 analytic expressions for the number of cycles spent in propagating the crack can be obtained. For $\beta = 4$

$$N = a_i\ \mathrm{arctg}(a/a_i)/(da/dN)_i \qquad (5)$$

and the number of cycles to fracture N_F is

$$N_F = a_i\ \mathrm{arctg}(a_F/a_i)/(da/dN)_i\ . \qquad (6)$$

Since $a_F \gg a_i$, $\mathrm{arctg}(a_F/a_i) \doteq \pi/2$ and fatigue life is thus inversely proportional to the initial crack growth rate of a small equivalent crack. This approach can be used for fatigue life prediction of smooth specimens cycled with constant plastic strain amplitude and yields the fatigue life curve in the form of the Manson-Coffin law (10).

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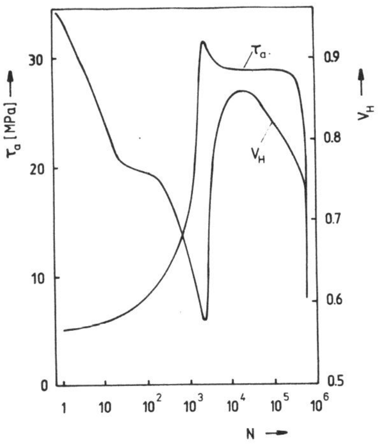

Figure 1 τ_a and V_H vs N in single crystal; $\gamma_{ap} = 1.2 \times 10^{-3}$

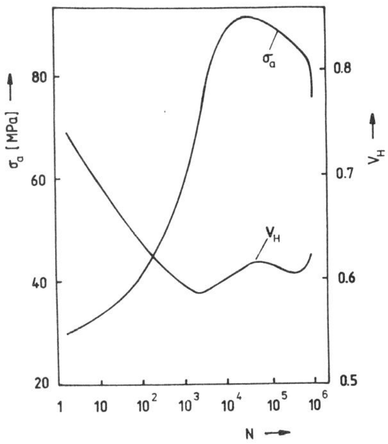

Figure 2 σ_a and V_H vs N in polycrystal; $\varepsilon_{ap} \cong 1.7 \times 10^{-4}$

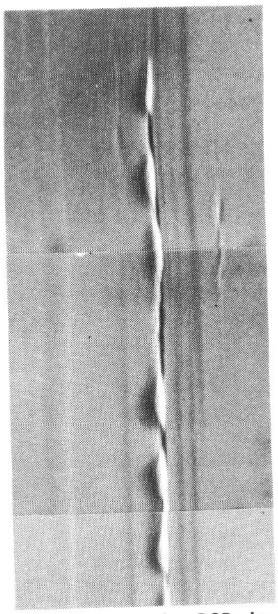

Figure 3 Embryonic PSB in single crystal

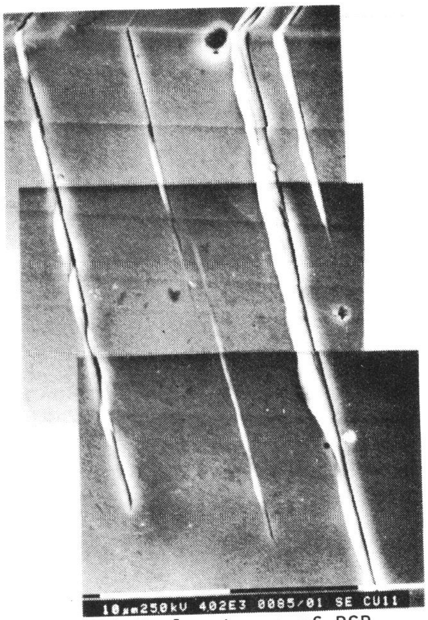

Figure 4 Early stages of PSB in polycrystal

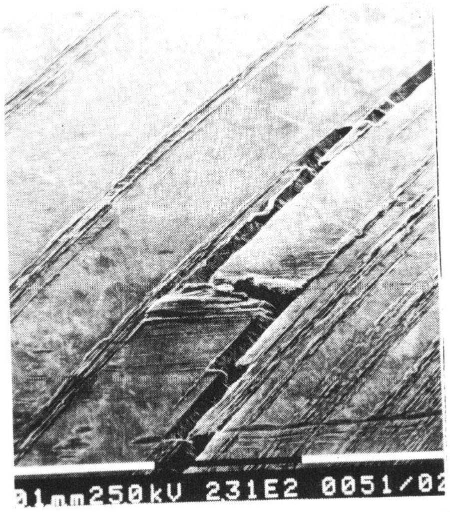

Figure 5 Dominant crack in single crystal

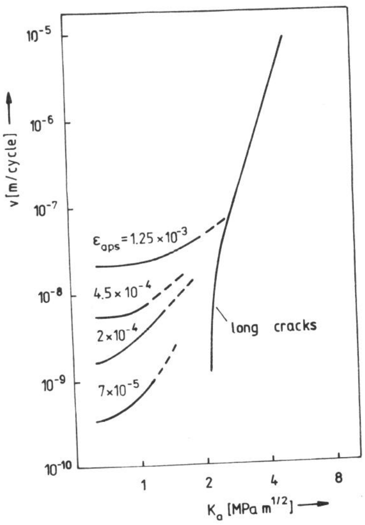

Figure 6 Crack growth rates in polycrystal