CYCLIC DEFORMATION AND FATIGUE MECHANISMS

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ABSTRACT

The cyclic deformation of monocrystalline and polycrystalline metal of wavy slip mode is briefly reviewed and connected to crack nucleation in long life fatigue by the idea of surface roughening. It is shown that the cyclic deformation of planar slip alloy is closely connected with that of the wavy slip base metal, even to the value of the plateau stress in the cyclic stress-strain curve. It is pointed out that persistent slip bands deform by strain avalanches and the consequences of this for both crack nucleation and propagation are considered. Speculation about ductility exhaustion at a crack tip in less ductile metals is also offered.

KEYWORDS

Cyclic deformation, wavy and planar slip mode, crack nucleation, surface roughening, strain avalanches, persistent slip bands, crack propagation.

INTRODUCTION

The problem of fatigue is one of the most complicated that modern designers must confront, and a proper design must take account of all relevant mechanical, material and environment factors. Whatever the complexities, the fatigue process can be reduced to four stages: the cyclic deformation of the material which conditions it for crack nucleation, the crack nucleation process itself and the related early growth of the crack, and the crack propagation stages through plastic and elastic regimes. In most engineering materials, plastic deformation at the crack tip will accompany all stages of propagation, irrespective of the stressing environment. Therefore, cyclic deformation plays a central role in all aspects of the fatigue process. For this reason the choice of cyclic deformation as the subject for a plenary talk is a good one. Unfortunately, space limitations are so severe that sorting out the available information is not feasible. Moreover, cyclic deformation has been reviewed repeatedly not only in preceding volumes of ICF going back to No. 3 (Munich), but also in the recent past (1-6). The approach to be adopted here, then, is to review the salient features of cyclic deformation in f.c.c. metal, and then to address those current issues which impinge most directly on the mechanisms of fatigue failure, with particular reference to long life fatigue. Much of this discussion involves new results.

CYCLIC DEFORMATION

It is well-known that an initially annealed metal will harden under the action of cyclic stresses or strains in the early stages of life. For pure metals and simple alloys, the hardening rate declines with cycling until a constant flow stress is attained. The material is then considered to be in 'saturation', and a plot of the saturated stress versus the applied strain is known as the cyclic stress-strain curve. Such a plot for monocrystalline copper oriented for single slip is shown in Fig. 1a. It happens that the form of this curve, dominated by a plateau, which was confirmed by several investigators (7-9), does not depend on orientation within the standard triangle (10).

The onset of the plateau is marked by the formation of persistent slip bands (PSB's) which develop just before saturation is attained (9). Before they form, the slip is quite homogeneous, the gage section is covered with fine slip, and the dislocation structure consists of loop patches. The behavior of loop patches at stresses lower than those of the plateau has recently been documented in detail (11). When PSB's occur, the strain is localized within them (7,8) often to very high values of localized strain (8), and the remainder of the gage section, occupied by the loop patches, is deprived of most of the strain it hitherto carried. The interpretation of cyclic deformation in pure metals has thus been analysed (see refs. 3 & 4 for review) in three parts, loop patch behavior during the initial hardening, saturation behavior involving PSB's and high strain behavior for deformation at strain amplitudes greater than that of the plateau, in which the slip again behaves homogeneously (to first order) and the dislocation structures consist of cells. For strain amplitudes lower than that of the plateau, the deformation remains homogeneous with continued cycling and the strain localization necessary for failure does not occur. No PSB's are formed and the lower end of the plateau marks the fatigue limit (12); this will be true of both pure metals and a wide range of alloys including age hardened alloy which also shows a plateau (13).

An issue which recently has been resolved in large measure is the cyclic deformation of polycrystals. High strain cyclic deformation has been studied for decades and goes back to Bauschinger. Only recently have techniques of strain measurement improved so that the cyclic stress-strain (CSS) curve can be documented at low amplitudes. That for copper is shown in Figure 1b. It appears that there is no true plateau for tests conducted in strain control and for the small grain sizes involved, although a tendency for one can be observed. Fig. 1b does show a plateau in load control in circumstances where the strains early in life are permitted to be large. These quasi-plateau regions of the CSS curve have been correlated with the presence of PSB-like dislocation structures in the interiors of polycrystalline specimens, i.e. there are mixtures of loop patches and dipolar walls which can be considered broadly similar to those in the plateau region pertaining to copper single crystals (15,16), and observations of PSB's quite similar to those found in single crystals have also been reported (17,18). Thus the similarity of the cyclic deformation of copper single and polycrystals is now well-established. The stress for the onset of PSB's in polycrystals can be correlated with that for single crystals, for example, simply via the Schmid factor. Still more recent work (19) shows that the occurrence of a plateau in polycrystals subjected to strain testing is grain size dependent, the plateau being naturally more marked with increase of grain size, and the same behavior is observed in age-hardened alloys. Correlating the flow stresses in mono- and polycrystalline materials remains, however, a largely unsolved problem.

CYCLIC DEFORMATION IN PLANAR SLIP METAL

The cyclic deformation of many different alloys and the tendency to form plateaus in their CSS curves have been reviewed previously (3,6) and Mughrabi has tabulated the available threshold values for the formation of PSB's in f.c.c. alloys (6). Pure metals have values of 6×10^{-5} to 10^{-4}, and those of alloys are usually several times larger.

Since the cyclic stress-strain curve of a really planar slip alloy in monocrystalline form has not been reported, Mughrabi et al (6) recently speculated about the surface roughness profiles obtained by computer simulation and compared these with observations by Lukas and Klesnil on fatigued Cu-30% Zn crystals (20). The interest of Mughrabi et al (6) concerned the development of surface damage which might lead to the formation of extrusions and intrusions and thus to crack nucleation. For this work, it was necessary to have values of the plastic shear strain amplitude. Referring to the work of Lukas and Klesnil, they determined that axial stresses of 82 and 118 MPa (shear stresses ~ 41 and 59 MPa) would yield plastic shear strain amplitudes of 10^{-4} and 4×10^{-4} respectively, and they employed results on polycrystalline Cu-30% Zn to make this connection. They further made predictions about the irreversibility of slip.

The danger of this kind of speculation is indicated by an examination of the actual CSS curve of monocrystalline Cu-16 at% Al, shown in Figure 2a, recently measured in our laboratory (21) (it is unlikely that the curve for Cu-30% Zn will be much different). It is astonishing that, in spite of the high friction stress of this alloy, the plateau as measured in an ascending step test has the same flow stress as that for copper. Since the flow stresses as measured for companion specimens cycled at constant strains in the plateau read a little higher than that for the step test and more in line with alloys of lower solute concentration (6), it is probable that the plateau stress is ~ 32 MPa (see Fig. 2a). This is much lower than the stresses employed by Mughrabi et al (6) for which the strains are much higher than they believed. These results also suggest that their conclusions with regard to slip irreversibility and cracking in planar slip materials must also be regarded as suspect.

The onset of the plateau in Cu-16 at% Al lies at the highest strain so far measured for this point, and reflects the rather small difference between the types of dislocation structure found at stresses below the plateau and within the plateau (22). The relatively low value of the plateau stress is interpreted by the destruction of the solute clusters and regions of local order, caused by repeated cutting by dislocations gliding to and fro. The high value of the flow stress at low strains associated with a high friction stress, could be further depreciated at higher strains or more cycles by a reduced effect of thermal activation. This effect would decline with cycling because an increase in mobile dislocation density for constant strain rate would reduce the average dislocation velocity (22).

The behavior of polycrystalline Cu-16 at% Al alloy at low cyclic strains is shown in Fig. 2b (23). Here again the CSS curve was measured by a step test but the results for companion specimens cycled at constant strain agree better than those for monocrystalline alloy. The two steps for strains greater than 10^{-3}, ascending step test, were especially interesting because, on making the step, there was an initial stress reduction, possibly due to PSB nucleation. The strains for which this behavior was observed are roughly consonant with those associated with the plateau in the same material in monocrystalline form (Fig. 2a). Making a connection between the behavior of monocrystalline and polycrystalline material is difficult for planar slip alloy because the grain size is a most potent factor in controlling the flow stress and is poorly understood. The grain size of the material used for Fig. 2b was ~ 0.3 mm.

Although polycrystalline material of planar slip mode has long been known to show PSB's, their existence in monocrystalline material has been doubted because Lukas and Klesnil did not report them (24). We now know that the reason for this is that the stresses used by them lay above the plateau. Buchinger et al have recently studied microscopically the specimens for which the mechanical behavior is shown in Figure 2a (22). Within the plateau, Cu-16 at% Al does indeed show PSB's and their dislocation structure is shown in Fig. 3, compared with that given by strains below the plateau. The dislocation structure which exists between PSB's or below the plateau is the "planar analogue" of loop patches, i.e., edge dislocations of the primary system predominate either as multipoles or as planar dipolar arrays of which a good example is shown in Fig. 3a. The PSB's are contained in thin planar volumes densely packed with primary dislocations but a very high density of secondary dislocations has also been found (22). The secondary dislocations are considered to have been triggered by self-stresses associated with groups of prismatic loops of primary dislocations. These have often been considered in the past as responsible for crack nucleation and propagation, the idea being that the Schmid factor for the primary system is zero under these stresses and thus fracture results. However, it is easier to excite slip on secondary systems, as found here. More details can be found in reference (22). It is most interesting that these PSB's have a strain localization similar to that of copper. This can be seen in Fig. 4, showing a slip offset across a prominent PSB. After the specimen illustrated was saturated, the test was interrupted, the specimens was polished and it was finally strained to the tensile strain limit in order to produce the offset observed. For a PSB of the observed width, ~ 40μm, and the measured offset, ~1.4μm, the local shear strain is the order of a few percent, typical of that found in copper. The fatigue lives of such planar slip metals can therefore be expected to be similar to those of copper, provided the tests are conducted in strain control.

CRACK NUCLEATION CONSIDERATIONS

The application of many strain cycles produces a notch-peak profile in the PSBs, and the most important feature of the profile in producing cracks is the distribution of more or less deep and sharp intrusions that form gradually during the process of irreversible random slip. Different authors have approached this problem in different ways. Brown and coworkers (25,26) and Mughrabi and coworkers (6) have emphasized the microscopic details of the dislocation processes, making use of the stress fields of the dipolar PSB dislocation arrays (26) or computing surface profiles based on irreversibility of dislocation paths estimated from TEM observations (6), whereas Cheng and Laird (27) model crack initiation from statistical considerations of slip randomness and observed degrees of slip localization. Mughrabi et al (6) have correctly pointed out that these models are complementary in many respects.

Recent observations of hysteresis loop morphology shed additional light on the dislocation behavior in PSBs. Yan et al (28), by recording hysteresis loops with a rapid response device, have shown that PSB's deform by means of strain avalanches. The load drops associated with these avalanches are clearly seen in Fig. 5. This means that the stress required to propagate dislocations down the PSB channels (the bowing stress for the typical channel width is much less than the plateau stress) is less than that required to nucleate dislocation motion in the PSBs. These avalanches, which are different from the strain bursts previously observed by Neumann (29) (his bursts last over many cycles rather than fractions of a cycle) are highly frequency sensitive, becoming larger at higher strain rates. This explains why this phenomenon has not been observed in inter ferometric measurements (8) where straining was carried out at low rates.

The dislocation models for crack nucleation of Brown and coworkers and Mughrabi and coworkers have not yet confronted this new observation. These load drops are clearly not connected to the energy released by the propagation of cracks along PSB - matrix interfaces (Brown) because they occur before cracks form. Nor is it known how the disposable parameters used by Mughrabi (the number of dislocations per group or the slip irreversibility at the surface) will be affected by these observations. The less specific model of Cheng and Laird is not affected by this observation because it deals only in stochastic events, and the strain avalanches fall in this category.

It is useful to note that the characteristics of strain avalanches depend on the degree of damage (28,31) and thus can be used to measure the state of damage without reference to the history of an affected part. A patent for a method of damage assessment based on this phenomenon has been disclosed.

CRACK PROPAGATION MECHANISMS

In ductile metals it is well-known that short cracks, cycled at low stress intensities, propagate by Stage I mechanism, which is closely identified with crack nucleation, and difficult to distinguish from crack nucleation. In so far as studies of cyclic deformation deal with crack nucleation, they also deal with this form of propagation.

Longer cracks, propagating by Stage II mechanism, are known to do so by the plastic blunting process (or one of its variants (30,33)) which also gives rise to striations on the fracture surface. Progress in understanding this mechanism of fatigue has been steady and has been especially useful in providing a clear model which can be quantified by finite element techniques. The observation of strain avalanches, noted above, has also been found to apply in studies of crack propagation. Figure 6 shows typical hysteresis loops measured by a rapid response device when the specimen contains a large crack. Evidence for the existence of the crack in the specimen is shown by the slight concavity of the hysteresis loop near the compressive tail. This concavity is well known to workers in fatigue but the strain avalanches are a new observation (31). Note that the avalanches become more pronounced as the crack grows longer. The interpretation here is that the deformation at the crack tip occurs as strain avalanches (31). This strain "runs away" from the strain signal used to control the electro-hydraulic machine, and the machine responds by unloading the specimen; hence the load drops. This slight decrease of load is sufficient to cause slight reverse plastic deformation at the crack tip and a consequent slight closure of the opening crack. As the load again increases, another strain avalanche is emitted from the tip and so forth during the whole of the tensile stroke. The asymmetry of the avalanches in tension and compression yield information about the closing of the crack during the compression stroke. Such avalanches should be readily detectable by an extensometer positioned to measure crack growth in a regular fracture mechanics type of specimen, and could provide a useful tool for studying fracture mechanism.

The conclusion is that the crack opens by a series of jerks alternating with slight closures due to the load drops (31). One might expect at least two consequences from such behavior: a) the striation will contain microrumples between the extreme valleys which mark the progress of the crack per cycle and b) the openings of the crack will be accomplished by discrete bands, each associated with an avalanche, and probably of the type observed by Neumann and coworkers using the SEM (32).

The plastic blunting process leads naturally to a square dependence of the crack propagation rate on stress intensity. One of the unsolved problems of fatigue is to understand why the dependence increases to the fourth power (or higher) in many complicated materials. Models involving the exhaustion of ductility at a crack tip can be manipulated to predict the correct (roughly) power dependence. The suggestion is now offered that recent studies by Rahka and Laird (34) point to a realistic physical basis for ductility exhaustion, as follows:

These workers cycled a Cr-Mo-V rotor steel at both room temperature and 550°C, and measured the diametral strain in specimens subjected to strain control tests with and without tensile holds, and where the strain was controlled by a separate axial extensometer. They were thus able to monitor the plastic Poisson's ratio continually as a function of cycles. A typical result, for a plastic strain of ~1%, is shown in Fig. 7. It will be noted that the Poisson's ratio, at the start of cycling, was close to the expected, ideal value for incompressible plastic deformation, and sometimes even exceeded this value because of strain localization. However, as cycles accumulated, the value of the Poisson's ratio was observed to drop steadily. In more recent studies (35), this effect has been found to be strongly amplitude dependent, becoming more marked with increase of strain. This result was initially baffling to interpret because little damage was found in the microstructure. We now have evidence that the reduction in ductility is caused by cavity formation which is difficult to detect because the cavities close up on unloading. Their formation, however, not only has an adverse effect on ductility but also makes the material "compressible".

Since the strains at a crack tip are extremely large, exhaustion of ductilty and periodic fracture at cavities associated with impurities can be expected to enhance the sensitivity of crack propagation rate to the stress intensity. This matter seems worthy of further investigation.

While limitations of space do not permit full justice to be rendered to the subject of cyclic deformation and fatigue fracture, it is hoped the above comments are persuasive that studies of cyclic deformation can yield valuable insight about the mechanisms of fatigue fracture.

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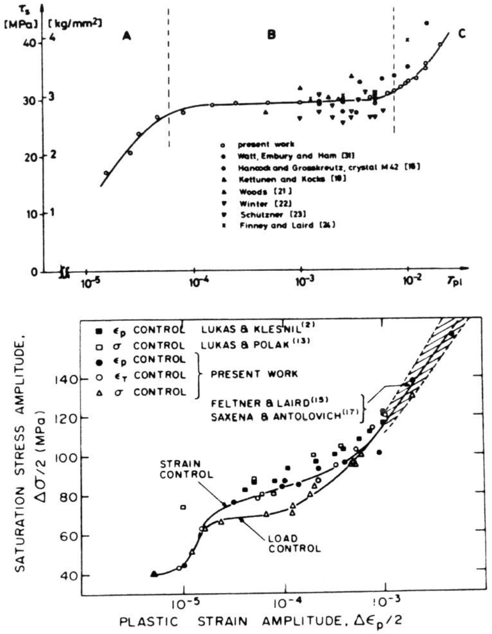

Fig. 1

Cyclic stress-strain curves for copper: a) single crystals shear stress vs plastic resolved shear strain amplitude (Courtesy of Mughrabi and Elsevier (9)); b) polycrystals: obtained for tests under various forms of control compared to results by Lukas and Klesnil (14) at low strains and other workers at high strains. Courtesy of Figueroa et al and Pergamon (15).

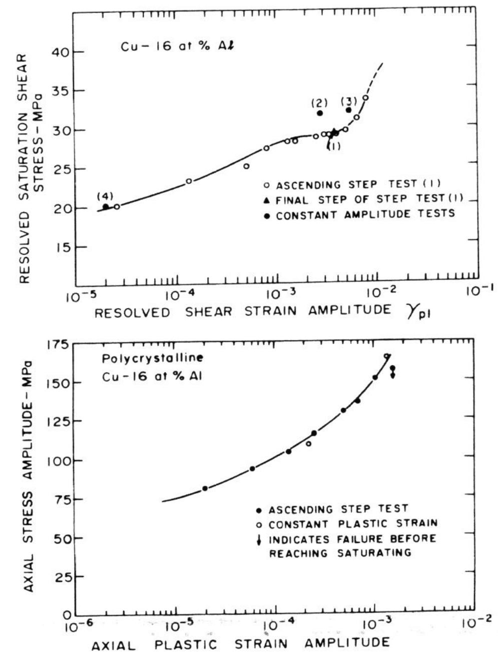

Fig. 2

Cyclic stress-strain curves for Cu-16 at% Al alloy: a) single c in single slip orientation, showing plateau behavior, for both an ascending step test and companion specimens cycled at constant strain; b) polycrystalline alloy of small grain size for similar kinds of tests. The specimen used for the step test failed during the last step. Unpublished results (21,23).

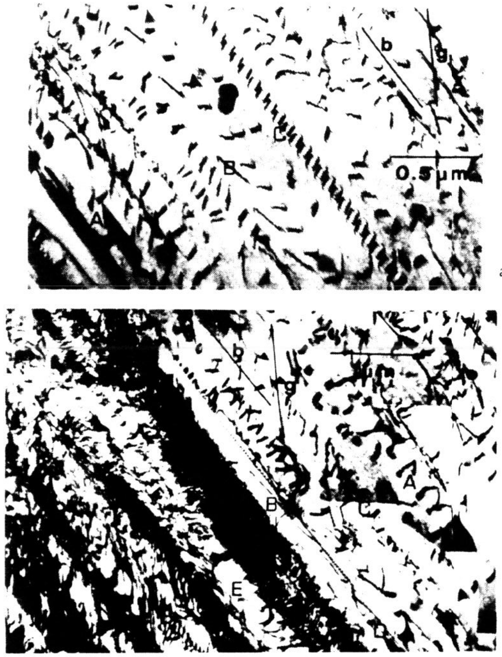
a
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Fig. 3
a) The dislocation structure which forms between PSBor below the plateau in a planar slip alloy - Cu-16 at% Al. The dipolar array referred to in the text is labelled C. Foil sliced on $(1\bar{2}1)$, g = $(\overline{111})$. b) PSB structure in Cu-16 at% Al viewed normal to the primary plane (foil slice $(1\bar{2}1)$, g = $(\overline{111})$. The PSB occupies the left hand half of the micrograph; matrix structure is to the right. Unpublished results (22).

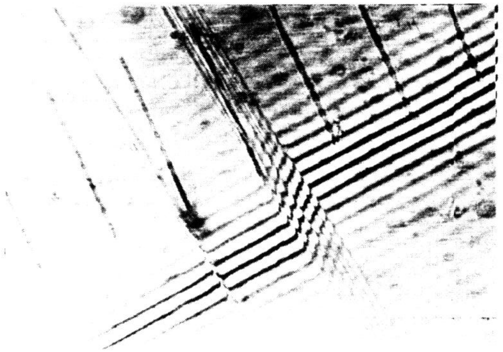

Fig. 4
Interferogram of the fatal PSB, i.e. the one that ultimately caused the fatal crack, in Cu-16 at% Al. The specimen was cycled at a shear strain amplitude of 2.8×10^{-3} and was well saturated. The short equally-spaced marks are fiducial lines indicating the magnification. The separation of the marks represents 100μm. Unpublished results (21).

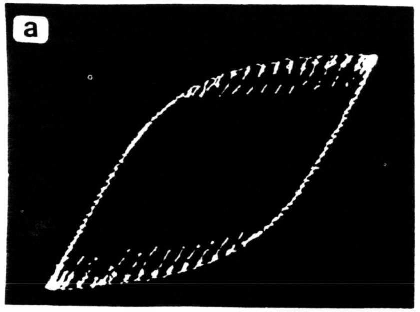

Fig. 5
Strain avalanches shown as load drops in the course of a hysteresis in monocrystalline copper cycled at a strain amplitude of 2×10^{-3} and undergoing deformation by PSBs. Unpublished research (28). The ordinate indicates the load and the abscissa axial strain.

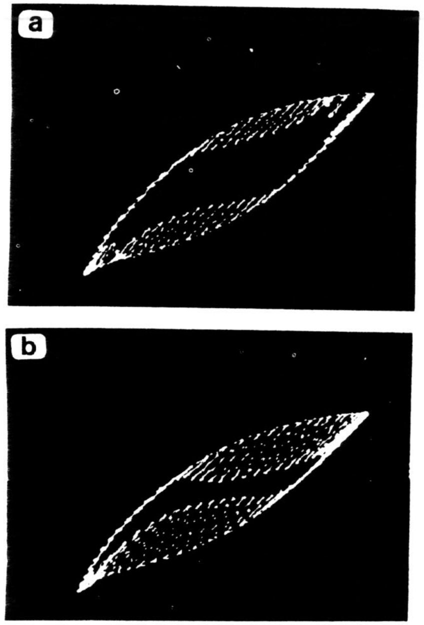

Fig. 6 Strain avalanches systematically occurring in a copper crystal containing a large crack. The hysteresis loops were measured in the same manner as shown in Fig. 5. a) 224,203 cycles b) 230,530 cycles. The specimen failed in 234,048 cycles. Note the asymmetry of the strain avalanche behavior in tension (up) and compression (down). Courtesy of Yan (31).

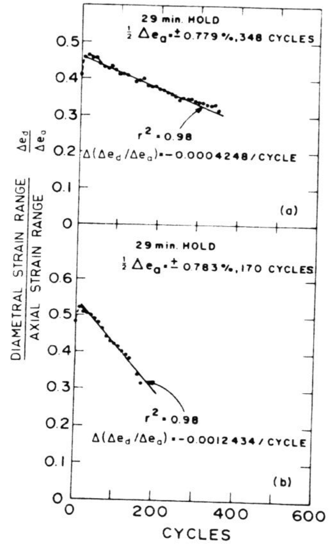

Fig. 7 Variation of diametral strain range to axial strain range in cycling 1 Cr-1 Mo-1/4 V steel with hold in tension. r^2 indicates the value of a regression correlation. Taken from ref. (34).