21. Formation of Slip-Band Cracks in Fatigue

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ABSTRACT

The results of alternate bending, constant-deflection fatigue tests on high-purity copper single crystals are presented. The development of slip-band extrusions and cracks is described. The results demonstrate that slip displacements normal to the surface are prerequisites for the initiation of slip-band cracks. The Mott mechanism of extrusion is discussed, and a modified mechanism is proposed which relates cross slip to operative sources and is consistent with the rapid development of extrusions.

Introduction

There have now been many demonstrations of slip-band cracking in fatigue. Recent studies have also established in good detail important topographical features of the fatigue slip bands,[1–6] particularly interesting observations having been made by Wood [5] on metallographic taper sections. From the findings to date, particularly on f.c.c. metals, it seems a reasonable view that slip planes operate with a "ratchet" action under cyclic straining which generates a characteristic notch-and-peak topography and that cracks originate from notches produced in this way. Two quite specific mechanisms for such a process have appeared in the literature: The Cottrell and Hull [7] mechanism involves alternating and intersecting slip, while that of Mott [8] is based on the action of cross slip causing a screw dislocation to travel in a closed circuit to extrude a tongue of material from the free surface and leave behind a fatigue crack. A paper current also argues for the importance of cross slip in crack formation, using the results of an experimental program on fatigue in single crystals of copper.[9]

For all of the progress of recent years, however, the basic mechanism by which an irregular surface contour develops and gives rise to fatigue cracks still remains unclear. Some additional insight into this problem is the contribution intended from experiments described in the present chapter.

Experimental

Copper single crystals for these experiments were grown from the melt as cylindrical rods of 99.999% purity. The test specimen (Fig. 1) is formed by an accelerated electropolishing operation termed electroshaping; vacuum annealing at 950°C, further electroshaping, and a final conventional electropolish complete the processing and yield a strain-free crystal. The specimens are deformed in alternating four-point bending to constant deflection (giving a total strain amplitude of 0.2%) in a specially constructed machine which imposes deflection through mechanically phased sinusoidal cams. Provisions are made in the grips to reduce possible axial and torsional restraints. Frequency is 10 cps, and bending moment is recorded continuously through use of electrical strain gages. The progress of hardening is represented by the variation in moment amplitude M_0 (half value of peak-to-peak range) with the number of bending cycles N. Further detail will be provided elsewhere.[9]

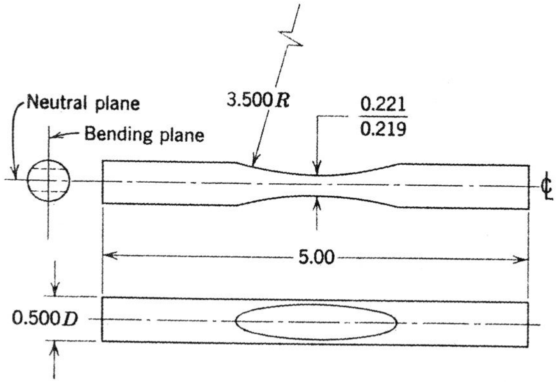

Fig. 1. Fatigue specimen (dimensions in inches).

A Geometrical Criterion for Cracking

A notch can develop and give rise to a crack in a surface that is pierced by the active slip direction. Therefore, the slip system involved in crack

formation must be oriented relative to the free surface so as to produce slip steps. Plane bending is particularly convenient for testing this corollary of the slip-band notch-to-crack proposition.

Experiments for this purpose have been made with two types of specimen, the distinction lying in rotational orientation around a common crystal axis. Figure 2 is a schematic illustration of each. Both are designated with an angle γ, which is included between the bending plane and the plane defined by specimen axis and primary slip direction. In $\gamma = 0°$ crystals, the primary system gave a step across the top (or bottom) surface, while in the $\gamma = 90°$ type the step was produced on the side; only a less highly stressed secondary system can give the step on the observation surface when $\gamma = 90°$.

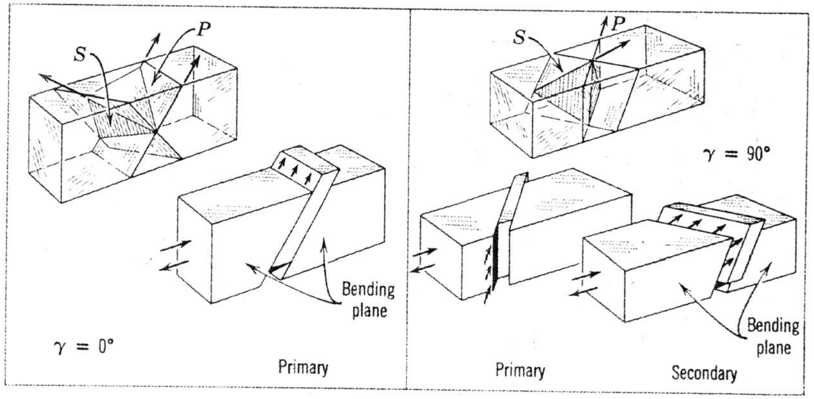

Fig. 2. (a) Primary slip step on top and bottom surfaces for $\gamma = 0°$. (b) Secondary slip step across surfaces for $\gamma = 90°$.

Examination for initial cracking was made after the removal of about 5 μ by electropolishing. In both types of crystal, crack indications were reasonably clear under the light microscope following 15 to 20 × 10^3 cycles. Total life in $\gamma = 0°$ crystals, marked by the appearance of a gross crack with an abrupt moment drop, ranged from about 75 to 100 × 10^3 cycles. However, consistent with the above views on crack formation, the slip-band cracks in crystals of $\gamma = 90°$ appeared first on nonprimary planes and developed more slowly, while total life was increased relative to $\gamma = 0°$ by a factor of 2 to 3. Some representative patterns are given in Fig. 3, where the specimen axis is in the horizontal position, as it is in all photographs in this chapter; the axial orientation of each crystal discussed is shown in the stereographic triangle in Fig. 4. The $\gamma = 0°$ crystal has been cycled through its full life and the typically fragmented primary-plane cracking revealed after a 5-μ electropolish; among the various crystals, the short fragmenting links could be identified in a number of cases with cracking on the cross-slip plane. An electropolished step is shown on the $\gamma = 90°$ crystal at 4×10^3 and at 25×10^3 cycles, with only critical-plane cracking in the latter case. After full life, primary cracks are found in $\gamma = 90°$ crystals mixed with rather heavy cracking

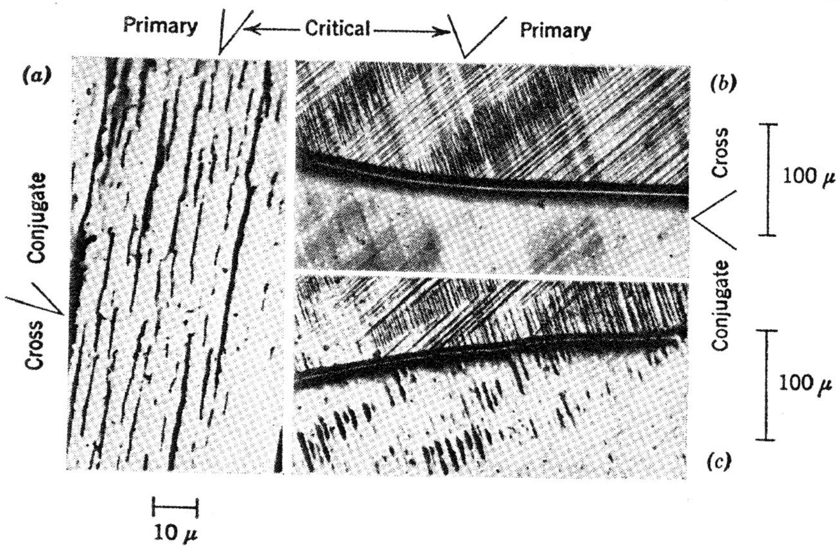

Fig. 3. Effect of γ on cracking: (*a*) Primary cracking in crystal 25($\gamma = 0°$), electropolished after total life. (*b* and *c*) Electropolished steps on crystal 31 ($\gamma = 90°$) produced by partial covering with Parlodion film; (*b*) no cracks after 4000 cycles, and (*c*) only critical-plane cracking in polished area after 25,000 cycles. Orientation shown in Fig. 4. Slip-plane traces noted on margins.

along secondary planes. These cracks are distinguished by an origin at the edge, where there are strong primary steps, and by a straight aspect without apparent fragmentation. The straight and fragmented cracks correlate with the emergence of screw and edge dislocations, respectively, from the surface of observation, a correlation previously made by Thompson [10] from the appearance of a crack encircling a copper single crystal cycled in alternating push-pull. A point of interest is that such patterns are also to be associated with operation of the cross-slip plane.

Slip-Band Topography

The important finding from experiments such as those described above is that displacements normal to the surface are prerequisite for the initiation of slip-band cracks. Different surface contours are observed,

however, and appear to be a function of many variables, which include the kind of material, its processing history, the amplitude of deformation, and temperature of test. The possibility of transition from one type to another has been discussed by Forsyth and Stubbington [1] and the general point made that the different forms of surface elevation may result primarily from localized deformation in soft regions. Since these details

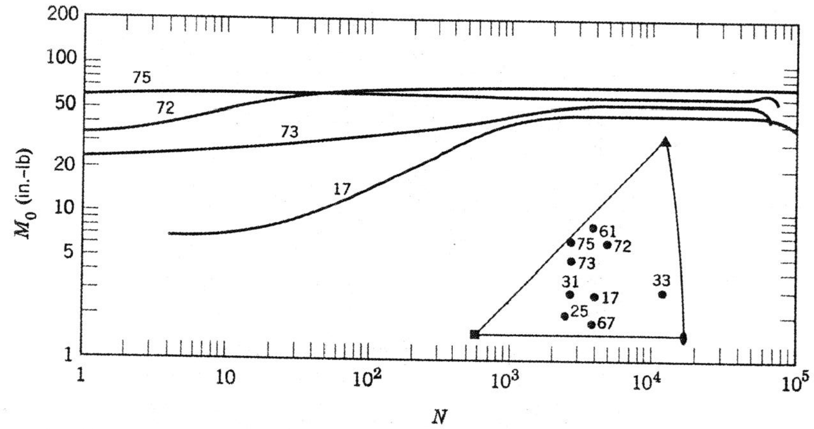

Fig. 4. Hardening curves giving variation of moment amplitude M_0 with cycles N. Crystal 17 annealed; others prestrained in tension: 73 into Stage II; 72 just into Stage III; 75 well into Stage III (approximately 5% elongation).

of topography must be closely connected with crack formation, the following experiments relating to their origin were undertaken on the $\gamma = 0°$ crystals.

Annealed Crystals

The hardening of an initially strain-free crystal (annealed and electroshaped) follows a typical path (Fig. 4, crystal 17) to saturation at about 2000 cycles with a plastic strain amplitude of approximately 0.15%. Topography changes along the hardening curve, and two methods have been used in obtaining different specimens for observation: Some crystals are prestrained before cycling, while others are cycled, electropolished, and then cycled further. All surface developments so produced lead eventually to what appears to be the same general kind of topography, in so far as resolution can be made in these observations.

In the strain-free electroshaped crystal, deformation begins with the appearance of fine, closely spaced slip lines. Clusters of slip bands become evident as hardening continues to saturation when they spread rather quickly over the surface. Representative views are shown in Fig. 5. An approximate view of surface profile across the clusters is produced by pressing a microhardness indenter into the region of interest; the effect is roughly that of taper sectioning along the band with a magnification

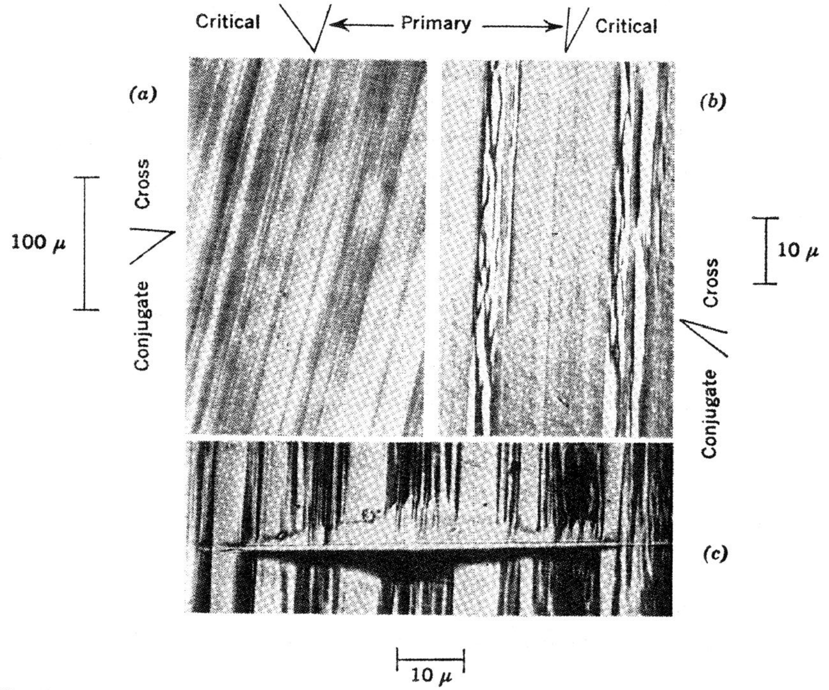

Fig. 5. Slip markings on annealed crystals: (*a*) Crystal 33 ($\gamma = 0°$) after 17 cycles. (*b* and *c*) Crystal 17 ($\gamma = 0°$) at 2000 cycles; (*b*) beginning of clustering, and (*c*) topography as revealed by microhardness indentation. Orientation in Fig. 4. Slip-plane traces noted on margins.

of about $2\frac{1}{2}\times$. Fine detail would naturally be obscured, but the over-all impression of notches and peaks should be reasonably representative (Fig. 5*c*). The development of extrusions at saturation to give such a profile can be demonstrated by smoothing the surface with an electropolish and recycling; with no more than about 5 additional cycles, thin tongues appear and spread over the surface to re-establish the clustered look. Although intrusions cannot be seen by the direct examination, they are found in taper sections described below. The rapid growth of extrusions has also been observed by Forsyth and Stubbington [1] in aluminum-alloy polycrystals.

Prestrained Crystals

The prestraining has been accomplished in two ways: by a tensile deformation prior to electroshaping and by surface abrasion after the shaping. Following an "intermediate" prestrain, broader and widely spaced slip bands appear in the electropolished surface at the end of but one cycle. Figure 6 shows a hardness indentation indicating an early type

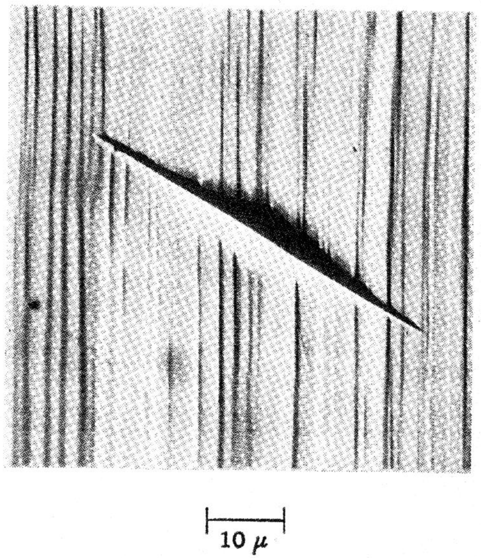

Fig. 6. Slip in crystal 67 (abraded and electropolished before cycling) after 5 cycles, showing block movement. View also representative of crystal 72.

of contour resembling strongly the "block movement" described by Wood.[5] This particular crystal was abraded with 0000 emery paper followed by coarse (<1-μ) and fine ($<\frac{1}{4}$-μ) diamond paste and finally electropolished. The hardening record for crystal 72, prestrained in tension by being pulled just into Stage III and forming similar topography, is shown in Fig. 4; little additional hardening occurs, and saturation is reached now in approximately one tenth as many cycles. The band spacing shrinks with cycling, and again at saturation the surface is covered by the narrow undulations.

Thin extrusions can be made to appear at an early stage with heavier prestrain. The abrasion treatment was repeated several times on one surface of a crystal after electroshaping. The surface was then divided longitudinally by covering one half with a Parlodion film (removed before cycling), and the exposed area was electropolished. A mixture of block movement and extrusion develops over the electropolished part as shown in Fig. 7, but the former detail does not change while the extruded regions

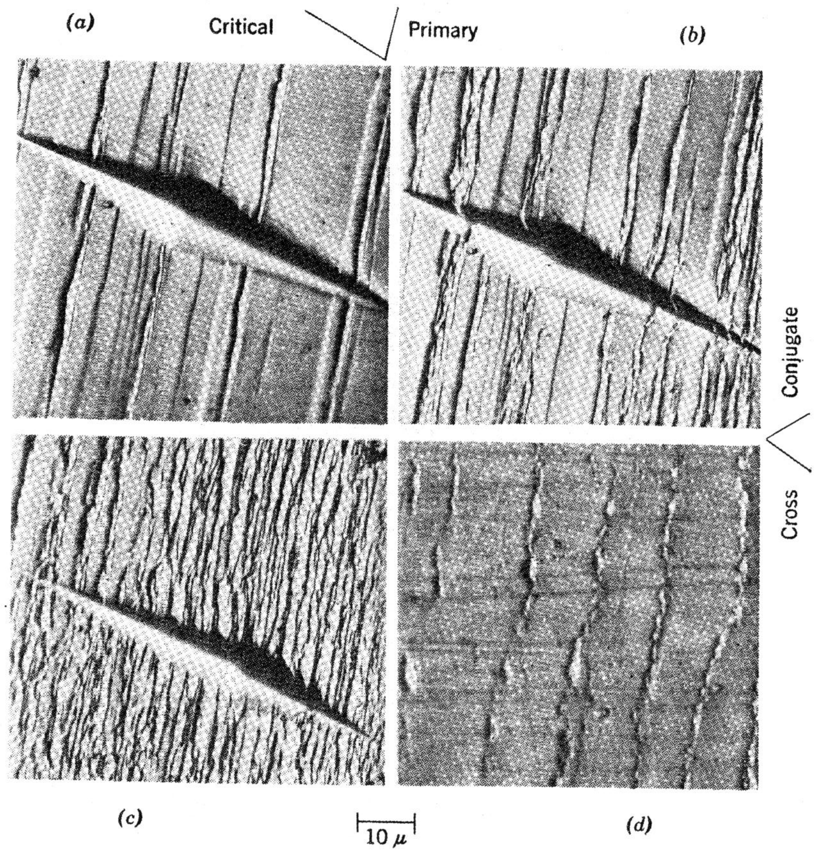

Fig. 7. Slip on abraded side of crystal 61: (*a*) Extrusion and block movement in abraded and electropolished area after 500 cycles with indentation at 500 cycles; (*b*) same view after 2000 cycles; (*c*) same area with new indentation after 5000 cycles; (*d*) extrusion over abraded area after 500 cycles.

spread into the clear areas and give the clustered effect. With abrasion alone, a narrower extrusion forms initially, to be followed in time by the clustering (Fig. 7*d*). Such effects are generally similar to those found recently by Stubbington and Forsyth [11] in polycrystalline copper after cold-rolling, annealing, and after a mechanical surface abrasion between annealing and final polishing. Crystal 75, prestrained in tension further into Stage III, developed topography on cycling similar to that in Figs.

7*a* to *c* and gave the record shown in Fig. 4. (The dependence of saturation moment on crystal orientation has been investigated, and, from that previous work, differences in the values given in Fig. 4 are considered to reflect only differences in axial orientation, not the effects of pretreatment.[9]) Now there is no hardening period; rather a slight softening leads to saturation. More pronounced work softening has been produced by interrupting cycling after saturation, twisting forward and back, and recycling. The resulting moment increase is transient, rapidly falling away until the original saturation level is reached; by electropolishing before recycling, active extrusion is observed with the softening.

True metallographic taper sections were prepared from the abraded crystal (Fig. 7) after the final gross fracture (Fig. 8). Surfaces were nickel-plated and polished to produce a taper magnification of about

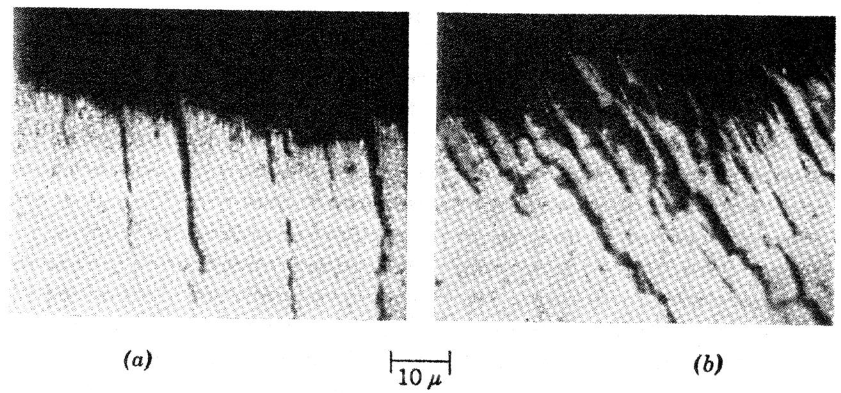

Fig. 8. Taper sections from crystal 61: (*a*) Block movement; (*b*) fragmented primary-plane cracking. Taper magnification, 10X.

10X. The block movement, strongly resembling Wood's views,[5] is found away from the minimum section where strain amplitude is less. A heavily fragmented pattern lies at about the minimum section, and here the sharp crevice or intrusion is fairly evident.

Although more detailed observations are still in order, a trend is evident in the change in topography with cycling. At saturation, broad patches of tightly clustered extrusions (and intrusions) develop, and it is within these that the fragmented fatigue cracks are found. Previous stages in the transition from initial fine slip to this end point prove to be more variable with experimental conditions. A correlation of metallography and hardening characteristics, however, does show the block movement to cease at saturation or during work softening.

Discussion

Various measurements have made clear the operation of a work-softening process or a process giving nonhardening plastic strain during fatigue. There are further indications that the final hardness level in a material under completely reversed loading is directly related to plastic strain amplitude; the above tests on copper crystals may be cited as an example, as may be the work of Coffin and Tavernelli [12] on softening in cold-worked polycrystals and recent work by Kemsley [13] on cold-worked copper. These suggest a unique relationship at saturation between plastic strain amplitude and the associated dislocation array: At saturation, for a given strain amplitude, a balance is maintained between annihilation and generation; upsetting this balance by extra straining then leads to work softening as annihilation predominates to re-establish the appropriate array. Viewed in this way, the same process would seem responsible for both work softening and the absence of work hardening.

From the observations that have been made, a basis can be found for suggesting a common mechanism underlying both work softening, or the nonhardening strain, and slip-band cracking. The connecting link is considered to be slip-band extrusion (and intrusion). Extrusion is observed with work softening or nonhardening strain. It has long been associated with soft spots and may be regarded as a means of relieving an excess of hardening. Furthermore, deformation under conditions causing extrusion leads to cracking. Therefore, the common mechanism seems a reasonable conclusion. From current theory of work softening in f.c.c. metals, an important feature of the mechanism is cross slip.[14] For a "dynamic" dislocation balance, however, more than just cross slip would appear essential. The screw parts of a dislocation line could be eliminated in this way, yet the possibility of large, accumulated plastic strain without hardening or X-ray line broadening requires that edge parts also be kept in check. Edge annihilation with alternating slip on nearby planes can be imagined in general terms, and one specific possibility is the process of internal void formation suggested by Fujita [15] and used by Mott [8] in a theory on the origin of fatigue cracks.

In Mott's theory, this problem of edge annihilation does not come up in quite such an important way, as it is assumed that multiplication ceases with hardening and that a few mobile dislocations travel up and down the slip bands as cycling continues. A schematic view of the result is given in Fig. 9*a*. If a length of screw dislocation is visualized as coming out of the page, a complete circuit as indicated in one full cycle would produce extrusion (or perhaps intrusion) through one slip step. The mechanism could well be an important one, but there are two reasons

why it may have less than general application: (1) Strain hardening may be absent at rather large amplitudes of plastic strain when the view of no multiplication would seem less reasonable. (2) An extrusion develops rapidly in a very small number of cycles, as if it were a burst of unstable flow, which must demand the near simultaneous movement of a great many dislocations. From rough measurements on the photographs, extrusion length in the copper crystals is around 2 to 5 μ. A slip displacement of this size would require 10^4 or more dislocations. Since

(*a*) (*b*)

Fig. 9. (*a*) Dislocation movement for extrusion according to Mott.[8] (*b*) Modification showing cross slip at pile-ups.

extrusions were observed to form within 10 cycles, an average movement of the order of 10^3 dislocations of one kind must occur in each cycle.

A possible modification to provide for these two points is illustrated in Fig. 9*b*. Imagine two surface sources, S and S', of opposite sense, forming pile-ups as shown. On the next half-cycle, cross slip occurs and reduces the size of the pile-ups; one source is then free to operate and throw out many dislocations. In the remaining half-cycle, the other source, having lost some of its piled-up dislocations by the cross slip, is similarly free to operate. The number of cycles during which the pair will be active depends on how the production-annihilation balance is maintained. It may be reasonable to expect only a few cycles of operation when another nearby pair operates in similar fashion. In this way, an explanation can be found for the observation that extrusions develop rapidly and spread in clusters without a steady growth of individual extrusions.

The possible influence of plastic strain amplitude on the mechanism has been clearly indicated in past work.[5] If the proposals in Fig. 9 are realistic, then plastic strain amplitude is likely to be an important variable in determining which of these applies. More systematic tests of its effect on topography and hardening characteristics should be of much use in establishing conditions under which any given type of mechanism is applicable.

The Cottrell and Hull mechanism, like that of Mott, implies a more-or-less steady growth of extrusion as well as intrusion and may therefore be subject to the same restriction. Furthermore, in this mechanism, the extrusions and intrusions are to be expected on intersecting planes, yet views such as those in Fig. 8 show them to be parallel.

Other data might be referenced in arguing for the importance of cross slip in slip-band cracking.[9] A particularly striking result that can be interpreted as support for the argument has recently been published by Fegredo and Greenough:[16] Zinc single crystals with basal plane at about 45° to the stress axis were cycled in alternating push-pull at −196°C without failure, even at stress amplitudes (built up in cycling) greater than the usual "static" tensile fracture stress; nor were slip markings apparent, even under low magnification and after long periods of cycling. Cross slip would hardly be expected in these experiments. The perverse fatigue characteristics of high-strength aluminum alloys,[17] on the other hand, may correlate with the propensity to cross slip in aluminum.

One corollary to the cross-slip argument is that fatigue can be expected at stress levels where slip bands form, as discussed by Feltham.[18] However, if extrusion is the requirement for fatigue, cross slip alone may not be sufficient, as shown by the results with crystal 72. Extrusions in this case were developed only after hardening to saturation (Fig. 4), even though cross slip is reasonably expected from the prestrain into Stage III as cycling begins, and it is indicated by slip-band formation in Fig. 6.

Regardless of how the evidence is presented, however, the cross-slip idea at present seems to find reasonable experimental support and is useful for the rather specific predictions it can give about the influence of temperature, stress level, processing history, and even material on fatigue resistance. Conditions suppressing cross slip would in general be taken to increase resistance to fatigue failure. Many experiments can be imagined for testing its role in fatigue; the results that they give should be followed by a further advance in the understanding of basic mechanism.

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REFERENCES

1. P. J. E. Forsyth and C. A. Stubbington, *J. Inst. Metals*, **83,** 395 (1954–1955).
2. G. C. Smith, *Proc. Roy. Soc.* (*London*), *A*, **242,** 109 (1957).
3. D. Hull, *J. Inst. Metals*, **86,** 425 (1957–1958).
4. M. Hempel, *Proceedings of the International Conference on Fatigue of Metals* (*London, 1956*), Institution of Mechanical Engineers, London, p. 543 (1957); American Society of Mechanical Engineers, New York (1957).

5. W. A. Wood, *Phil. Mag.*, **3,** 692 (1958).
6. N. J. Wadsworth and J. Hutchings, *Phil. Mag.*, **3,** 1154 (1958).
7. A. H. Cottrell and D. Hull, *Proc. Roy. Soc.* (*London*), *A*, **242,** 211 (1957).
8. N. F. Mott, *Acta Met.*, **6,** 195 (1958).
9. M. L. Ebner and W. A. Backofen, *Trans. AIME*, **215,** 510 (1959).
10. N. Thompson, *Proceedings of the International Conference on Fatigue of Metals* (*London, 1956*), Institution of Mechanical Engineers, London, p. 527 (1957); American Society of Mechanical Engineers, New York (1957).
11. C. A. Stubbington and P. J. E. Forsyth, *J. Inst. Metals*, **25,** 90 (1957–1958).
12. L. F. Coffin, Jr., and J. Tavernelli, General Electric Research Laboratory Report No. 58–RL–2100, Schenectady, N.Y. (1958).
13. D. S. Kemsley, *J. Inst. Metals*, **87,** 10 (1958–1959).
14. A. Seeger, *Dislocations and Mechanical Properties of Crystals*, J. C. Fisher, W. G. Johnston, R. Thomson, and T. Vreeland, Jr., Eds., John Wiley & Sons, New York, p. 243 (1957).
15. F. E. Fujita, *Sci. Repts. Research Inst. Tôhoku Univ.*, *A*, **6,** 564 (1954).
16. D. M. Fegredo and G. B. Greenough, *J. Inst. Metals*, **87,** 1 (1958–1959).
17. R. F. Hanstock, *Fatigue in Aircraft Structures*, Academic Press, New York, p. 62 (1956).
18. P. Feltham, *Phil. Mag.*, **3,** 806 (1958).

DISCUSSION

R. L. FLEISCHER, *Massachusetts Institute of Technology*. The recent experimental study of copper single crystals under fatigue loading reported by Backofen has indicated the need for further study of the nature of the mechanism for fatigue crack initiation. He has shown that surface ridges or depressions, of the type that ultimately develops into cracks, appear along traces of primary slip planes during as few as 10 cycles of stressing. The fact that these markings measure as much as 1 μ in slip displacement implies that for such a surface marking at least 300 dislocations of each sign must escape the surface during 1 cycle. The rapid appearance of so many dislocations seems to indicate that new ones are created, and hence, that sources are acting. To form and enlarge a ridge or depression, such sources would need to act alternately in pairs; each source would act during only half of the cycle, but it would not reverse itself when the stress was reversed, and hence, it would not allow the step formed at the surface to be removed. If for the moment it is accepted that the action of dislocation sources is important, it becomes necessary to indicate why a source should give predominantly one-way slip. One possibility will now be suggested.

Consider a Frank-Read source, such as is indicated in Fig. D.1*a*, with the pinning points at a distance $2l$ apart and at a distance d from a free surface: The effect of the surface is approximately equivalent to the presence of an image dislocation of opposite sign placed a distance $2d$

from the source.[1] This means that the portion AB in Fig. D.1*a* will be attracted to the surface, and hence, it will bend more easily when it is acting under an applied stress toward the surface (Fig. D.1*b*) than away from it (Fig. D.1*c*). In general, there may be two relative maxima of energy as a loop is generated. Figures D.1*b* and D.1*c* show one of these

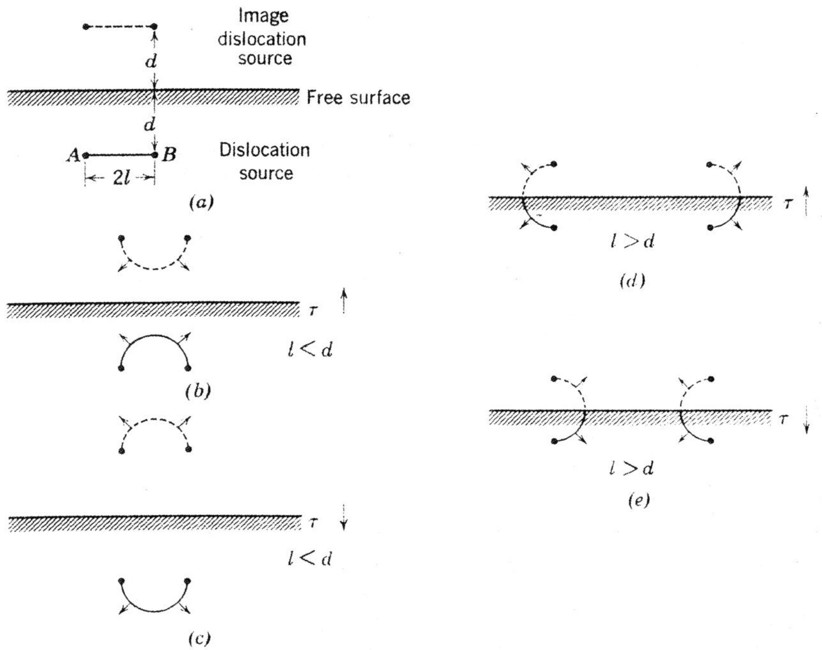

Fig. D.1. Action of a dislocation source near a free surface. τ indicates the relative sense of the applied stress.

maxima for positive and negative applied stress, and Figs. D.1*d* and D.1*e* show the other. It can now be seen that, in general, the source should operate at different values of stress, depending on whether it is positive or negative: First, assume that l is appreciably less than d. Then Figs. D.1*b* and D.1*c* indicate the approximate configuration at the stress required to operate the source. If l is appreciably more than d, Figs. D.1*d* and D.1*e* are appropriate. In both cases, because of the dislocation of opposite sign (whether image or real), the magnitude of the stress required is greater for one sense of rotation about a pinning point than for the opposite rotation.

The same effect would exist if a slip plane contained a pair of sources arranged in the same relative positions as the source and image pairs just described but located well within the crystal.

Although the description just given provides a reason for sources being "one way," it does not answer the question of why the dislocation loops generated in 1 half-cycle do not then return to the source from within the crystal and hence cancel out the surface step. Two possible reasons may be suggested: One is that they move far into the crystal and become stuck; another is that they are annihilated by other dislocations of opposite sign. Backofen has suggested that this latter process may be important and that cross slip would be an important means of annihilating dislocations on neighboring planes. This suggestion is consistent with the observations that single crystals of zinc, which do not show cross slip, have extremely long fatigue lives.[2]

References

1. J. S. Koehler, *Phys. Rev.*, **60**, 397 (1941).
2. D. M. Fegredo and G. B. Greenough, *J. Inst. Metals*, **87**, 1 (1958–1959).