20. Some Basic Studies of Fatigue in Metals

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ABSTRACT

The main theories of fracture by fatigue are reviewed in the light of recent basic experimental studies, including observations in which sections of fatigue slip bands have been examined at virtual magnifications up to 30,000 times. It is pointed out that two different mechanisms formerly have been classed as fatigue. The first, predominating when fatigue cycles impose large plastic amplitudes and corresponding to the early part of the *S-N* curve when specimens have short lives, causes failure by increasing internal stresses in much the same way as static deformation does; fatigue becomes in effect a delayed static fracture. The second, predominating when the plastic amplitudes are small and corresponding to the later part of the *S-N* curves when specimens have long lives, causes failure by (*a*) producing abnormal distortion along operative slip zones, (*b*) causing slip zones to develop surface disturbances ranging from sharp notches to sharp peaks, and (*c*) transforming the distorted zones into narrow fissures by at least one tenth of the expected specimen life; final failure then ensues by collapse of metal already long-saturated with fissures. Both mechanisms are correlated with solid-state theory.

Introduction

Any final explanation of fatigue must be given in terms of the lattice defects that, according to solid-state theory, determine processes of deformation and fracture. Fatigue deformation presumably differs from other modes of deformation in causing these defects to behave in a special way. Unfortunately, the special behavior cannot be predicted in sufficient detail by the theory in its present state. It is still necessary to proceed by a combination of theory and experiment.

Though there is already much experimental information on fatigue, most of it is for urgent purposes of mechanical design rather than for more fundamental application. For example, there are few direct observations on strain hardening during fatigue or on the total plastic strains imposed by the time fracture is reached. Indeed, theories of fatigue fracture appear to be in much the same position as theories of static fracture might be without the background information provided by static stress-strain curves.

In this chapter, an attempt is made to select observations that indicate special behavior of a metal structure during fatigue deformation and to review, in the light of these observations, the main theories of fracture. It is hoped that the treatment may help in assessing our basic knowledge of fatigue and in determining the form a final theory is likely to take.

Fatigue Deformation

It is desirable first to consider fatigue in simple form. Essentially, fatigue is concerned with the cumulative effects of small strains occurring alternately in one direction and then in the reverse. Therefore it is simplified if the cycles are symmetrical about zero mean strain, when it is free from superimposed unidirectional strains. It is further simplified if the cycles are applied to metal that is reasonably pure and initially annealed, when it is also free from superimposed internal strains. The pure fatigue so produced gives a standard of comparison. Complications due to unsymmetrical cycles, alloying elements, or internal strains can be introduced singly, and their influence on the basic mechanism studied separately.

The damage caused by pure fatigue cycles may result from continued reversal of the elastic strain or of the plastic strain, which also, as damping experiments show, is always imposed by the cycle.

Elastic strain is a reversible displacement of atoms from their normal lattice positions through distances no larger than about $\frac{1}{1000}$ of a lattice spacing. Such small reversible movements are not likely to initiate any structural flaws. They could, however, start propagation of existing flaws, for example, by processes, such as the Griffith mechanism,[1] which transform elastic energy into the surface energy of growing cracks. These processes can come fully into play, however, only in the absence of plastic deformation, which provides simpler ways of transforming the elastic energy, for instance, by slip or twinning movements. Accordingly, we may expect elastic strain to play a part in fatigue of brittle metals under all conditions and in fatigue of ductile metals under special conditions arising if the normal capacity of a ductile metal for plastic strain were somehow suppressed or exhausted.

Damage by the plastic component of a cycle, according to solid-state theory, may result not only from the intensifying of existing flaws but also from the starting of new ones. Since most metals of technical interest are capable of plastic flow, this component is likely to play the more fundamental part in practical fatigue. Its effects therefore form the main topic of the following sections.

Additive Effects of Reversed Plastic Strains

The problem of fatigue turns on how the structural effects of reversed plastic strains may cancel or add up. One major group of theories postulates that the effects add up in much the same way as do those of unidirectional strains, and that fracture finally occurs for the same reason, namely, because continued strains in each case build up internal stresses comparable to the intrinsic strength of the metal. For example, Gough and Hanson [2] proposed that fracture occurs when internal stresses, measured by strain hardening, locally reach a critical limit. Orowan,[3] postulating strain hardening that added up in particular ways, has made interesting estimates of the cycles needed to produce a critical total. The same postulate underlies some treatments in terms of dislocation theory. Static fracture on this theory, as shown by Mott [4] and Stroh,[5] may occur because dislocations (by piling up in front of lattice obstacles) can produce the necessary high concentrations of stress. On the same theory, fatigue fracture occurs as a result of comparable pilings-up, a recent typical treatment being that by Yokobori.[6]

It is interesting therefore to test the postulate first by observing through experiment how strain hardening does add up. This may be done by subjecting a test specimen to reversals of a given plastic strain e and finding the proof stresses (S_1, S_2, $S_3 \cdots$) needed to impose successive reversals (Fig. 1). These stresses measure the strain hardening and, plotted against the successive strains e added irrespective of sign, give for each amplitude the cyclic equivalent of a static stress-strain curve. Curves of this kind follow the pattern of Fig. 2, which was obtained for α-brass subjected to alternating torsion.[7] These curves show that strain hardening depends on amplitude. At larger amplitudes, strain hardening is relatively rapid and adds up progressively; at medium amplitudes, the hardening adds up more slowly; and at small enough amplitudes, hardening is small initially and then hardly changes. More extended experiments show that it is still small at fracture.[8] It appears therefore that the postulate of additive hardening needs qualification at

all amplitudes of cyclic strain and that, at the small plastic amplitudes of most interest to engineers, it may not hold at all.

It is interesting to test the postulate further by finding the total of plastic strain reached at fracture, again by adding the plastic reversals

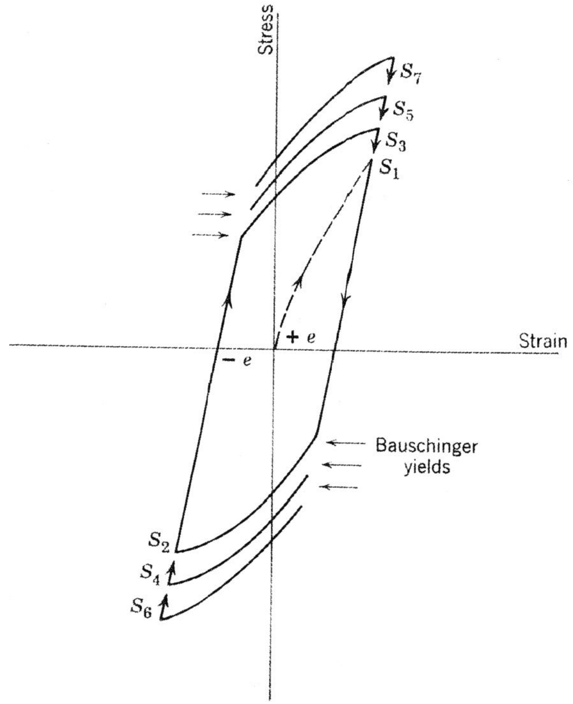


Fig. 1. Hardening by reversals of plastic strain e is given by S_1, S_2, $S_3 \cdots$. Hardening is slower than in unidirectional straining because of the Bauschinger effect. It ceases if the Bauschinger yield in one half-cycle compensates for hardening in next half-cycle.

irrespective of sign. It is found in the experiments just mentioned that this total too depends on amplitude. This is shown by curve A of Fig. 3, which gives the totals for α-brass under alternating torsion. It will be seen that the totals are relatively small until the amplitude falls to the range that ceases to strain harden efficiently. Then they increase rapidly. It is interesting also to compare this curve with curve B of Fig. 3, obtained when similar brass specimens were first coated with mercury, the mercury being used for its action in showing up high local stresses.[9] Curve B shows that, though the strain needed for fracture at each ampli-

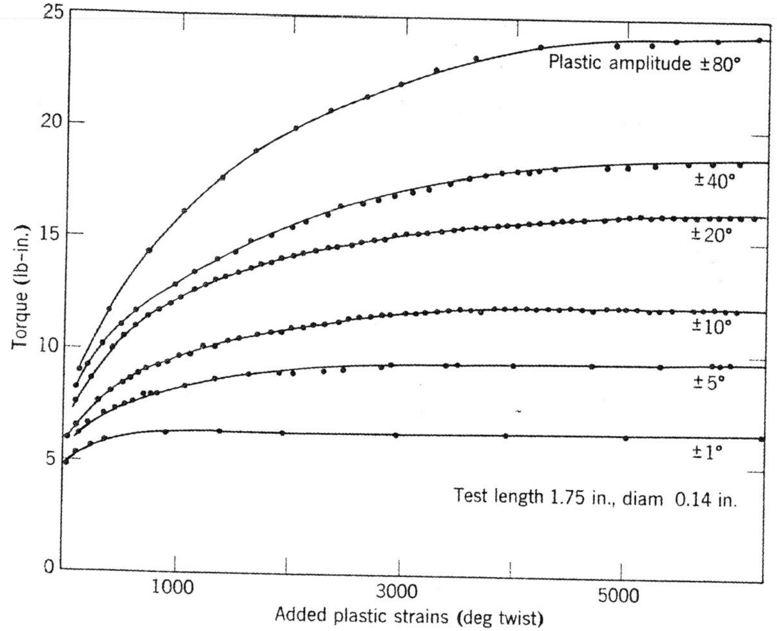


Fig. 2. Cyclic stress-strain curves for annealed α-brass subjected to various amplitudes of alternating torsion.

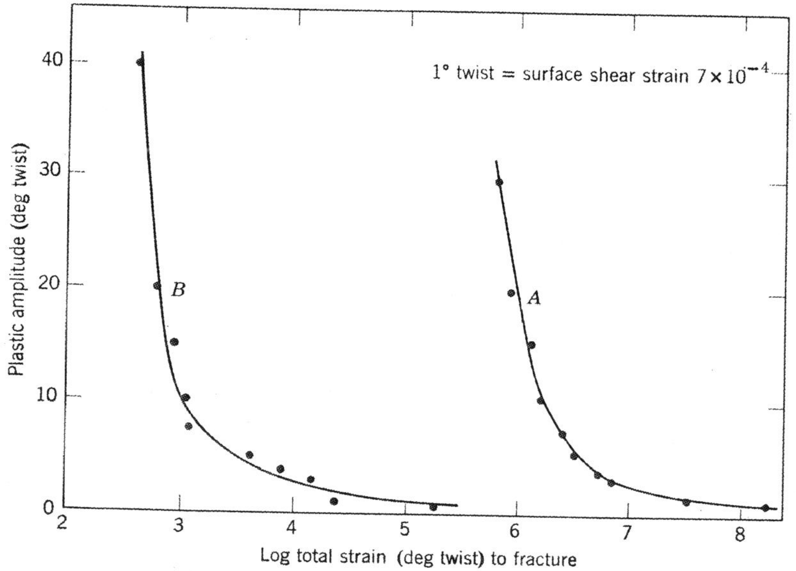


Fig. 3. Total plastic strain to fracture of brass under various plastic amplitudes of alternating torsion (*A* uncoated, *B* coated with mercury).

tude is reduced by the mercury, it still increases disproportionately when the plastic amplitude falls to the range of inefficient hardening.

Such experiments show that the mechanical effects of reversed plastic strains do not add up in the same way as those of unidirectional strains. The immediate reason is clear from the stress-strain loops depicted by Fig. 1. These show that some of the hardening by one half-cycle is lost in the reverse half-cycle (Bauschinger effect). At large amplitudes, more is added in the next half-cycle than has just been lost, so that hardening increases during successive cycles, though more slowly than during unidirectional increments of strain. At small enough amplitudes, only as much is added as is lost, so that strain hardening as a whole ceases. Accordingly, we have two peculiarities of cyclic strain: a reduced rate of strain hardening and, because of this ineffectual hardening, the possibility of abnormally high totals of plastic strain to fracture.

Local Additive Effects

It is still necessary to consider the possibility that mechanical effects may be additive only in highly localized regions so that they escape detection by such measurements as strain hardening, which determine an average condition. This is always an awkward point to test, but it is not supported by such observations as can be made.

For instance, we may expect that hardening in a metal deforming by slip is localized, if at all, along the slip bands. Therefore, if hardening is to remain undetected, the slip bands must be few and far between. It is found, however, that slip bands accompanying the nonhardening deformation after enough cycles may become as densely spaced as those accompanying large-amplitude or static deformation, which produces easily observed hardening.[10] Moreover, any hardening that does occur at the small plastic amplitudes occurs most during early stages of straining, when slip bands are few and faint, and least during later stages, when the bands multiply and intensify. Hence, observations on slip afford no support for a hardening that is too local to detect. On the contrary, they confirm the possibility of nonhardening slip.

Another test for localized hardening is to observe how fatigue specimens recrystallize on heating. Regions of hardening should form centers of easy recrystallization. It is found, however, that such centers form only in specimens deformed statically or by large plastic amplitudes. Specimens deformed by the nonhardening small amplitudes recrystallize only after severe heating, which would disturb an already annealed specimen.[11,12] Therefore, tests of this kind provide no evidence of local hardening. A further test, though of limited application, is to use mercury as an indicator of local stresses in susceptible alloys. This test, applied to α-brass, also is negative; it shows, as already indicated, that the mercury becomes rapidly less effective at the small amplitudes that permit nonhardening strain. Finally, there is the evidence from observations on fatigue in cold-worked metals.[13,14] These show that a small plastic amplitude may produce softening, not hardening, and that softening occurs by "recovery," that is, by dispersing regions of distortion, not accentuating them. Accordingly, it appears that small plastic amplitudes must impose a distinctive mode of deformation.

Interpretation of *S-N* Curves

This last conclusion accounts for the shape of the *S-N* curves obtained in standard fatigue testing. The well-known feature of these curves is a rapid increase in specimen life when the test amplitude falls to a range of the same order as the static elastic range of the metal. This feature is of course clearest when a metal exhibits the practically safe range

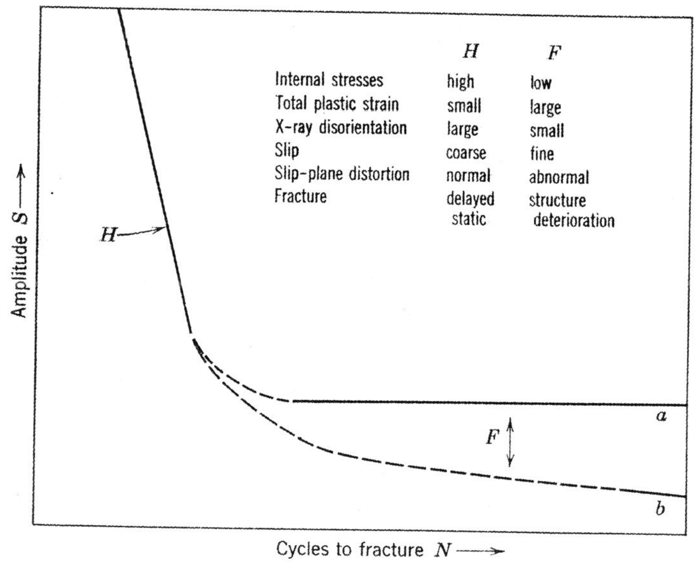


Fig. 4. *S–N* curve divided into *H* and *F* parts representing two distinct mechanisms.

typified by curve *a* (Fig. 4), but it holds to some extent when a metal shows the less-certain safe range of curve *b*, which may be regarded as an extension of the transitional bend indicated by the dotted line. The curves in effect are made up of an *H* part, corresponding to large amplitudes and short lives, and an *F* part, corresponding to small amplitudes and long lives.

When this feature is correlated with the preceding observations on strain hardening and strain to fracture, it is found that the *H* part corresponds to the larger plastic amplitudes that cause pronounced and progressive hardening and that the *F* part corresponds to the small amplitudes that permit nonhardening plastic strain. Apparently, the *S-N* curve takes the shape it does because it represents one mechanism at large amplitudes and another at small amplitudes, the first producing additive internal stresses and the second not. Observations described later confirm this point.

Additive Structural Changes

The additive effects so far considered are mechanical. The differences noticed must arise from different structural changes. It is interesting next to consider observations that correlate the two.

Early work, notably by Gough and his colleagues,[15] shows that the structural changes produced by cyclic straining are broadly the same as those produced by static straining; thus if a metal deforms predominantly by slip in one case, it does so in the other and according to the same laws. (Since most metals of practical interest deform chiefly by slip, we shall concentrate on this process.) This early work, however, partly for historical reasons, emphasizes similarities between the structural changes in cyclic and static deformation, for, at the time it was written, some unconventional processes of fatigue failure had been proposed, such as the crystallizing of possibly amorphous metal. Evidently it is necessary now to find differences between the effects of static slip and cyclic slip at various amplitudes.

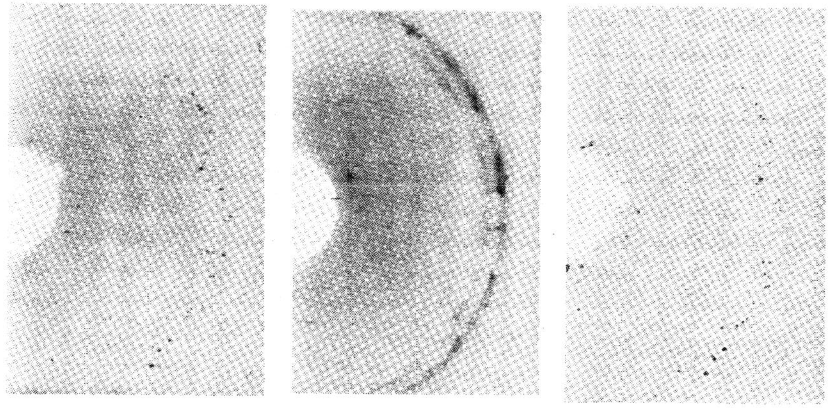
Several interesting peculiarities of cyclic slip have already been reported, such as the extrusion effects found by Forsyth [16] and the persistent slip by Thompson.[17] The following two seem particularly significant.

The first concerns the way slip bands develop during progressive deformation. Static deformation produces bands that are immediately strong — a band is either there or not there — and continued deformation multiplies such bands. Cyclic deformation at large plastic amplitudes, those corresponding to the *H* part of the *S-N* curve, produces similar bands; but cyclic deformation at small plastic amplitudes, corresponding to the *F* part of the *S-N* curve, produces bands that first appear faintly, after a thousand or so cycles, and then grow in strength gradually.[10] Each kind of slip is additive during progressive straining, but in a different sense. The large amplitudes primarily multiply coarse slip movements at different parts of a grain. The small amplitudes intensify existing bands by adding fine slip movements. This difference is significant in suggesting that the large amplitudes excite *coarse slip*, that is, avalanches of slip movements through some hundreds of lattice spacings at one time, thus accounting for the immediately visible bands; whereas the small amplitudes excite *fine slip*, or movements through only a few lattice spacings at a time, thus building up surface disturbances gradually where they happen to concentrate in bands.

The second peculiarity is brought out by accompanying X-ray diffraction observations.[8,10] These confirm the two different extremes of slip by showing that they produce quite different effects on the metal grains. Beginning with annealed metal, we know that the grains, since they contain relatively few imperfections, give sharp X-ray reflection spots (Fig. 5*a*). We then find that cycles of large plastic amplitude cause the grains to reflect over wide arcs, showing that each grain is reduced to disoriented elements, a distortion that in practice is greater the larger the amplitude (Fig. 5*b*). However, cycles of small plastic amplitude, even continued to the stage of fracture, leave the grains reflecting almost as sharply as they did at first (Fig. 5*c*). Thus the X rays show in a striking manner when slip movements become too coarse to be accommodated without disorientation of the constrained grains.

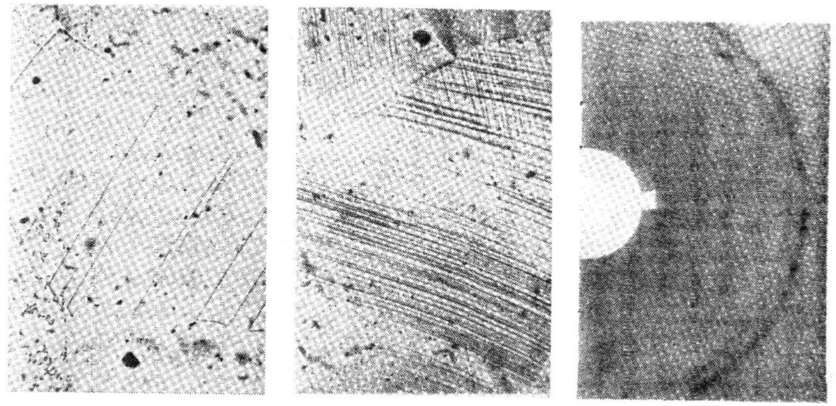
These distinctions between coarse and fine slip would be of geometrical interest only, were it not for the different physical consequence demonstrated by the following kind of experiment.[18] This is an attempt to determine by the above-mentioned tests the type of slip that will appear when small plastic strains, each small enough to cause only fine slip movements when applied in alternate directions, are applied in unidirectional increments. It is found that these unidirectional applications, unlike their cyclic equivalents, give rise virtually to the same structural changes as they would if imposed as a single large strain: They produce similar hardening, similar coarse slip movements, and the same heavy disorientation of the grains (Fig. 6). Now the small strains can cause coarse slip during the unidirectional applications only if some potential fine slip movements in each increment are held up before lattice obstacles and, after sufficiently accumulating, are released as coarse avalanches. Apparently it is impossible, at ordinary temperatures, to produce unidirectional deformation without in this way piling up some dislocations and, at intervals, releasing them in bursts; but it would seem possible to do this in cyclic deformation. We are thus led by experiment to this fundamental and distinguishing feature of cyclic deformation at small plastic amplitudes.

It is reasonable to suppose that the condition for avoiding this piling-up



(*a*) (*b*) (*c*)

Fig. 5. (*a*) Sharp X-ray reflections from annealed α-brass. (*b*) Heavy disorientation after only a few cycles at large amplitude $\pm 40°$ (alternating torsion, 1° equals surface shear 7×10^{-4}). (*c*) From same specimen as (*a*) after 1500 reversals of plastic strain 0.5°, and showing same reflections as (*a*), virtually as sharp.



(*a*) 10 μ (*b*) 10 μ (*c*)

Fig. 6. (*a*) Brass (as used for Fig. 5) after 1500 reversals of small plastic strain 0.5°; the slip movements are too fine to see. (*b*) Slip after only 150 unidirectional applications of the same small strain. (*c*) From the same specimen as (*b*) and showing accompanying heavy disorientation. Thus the unidirectional applications, unlike the cyclic applications, generate coarse slip avalanches.

is that the small cyclic strains, and the dislocations carrying them, should be reversed before their mean free path exceeds the average spacing of lattice obstacles. Therefore the kind of experiment previously described, especially with the unambiguous X-ray test for coarse slip, should lead to interesting information on the spacing of relevant lattice imperfections. Here, however, it is sufficient to notice how this avoidance of piled-up dislocations, on a scale comparable to the piling-up in unidirectional deformation, and the consequent avoidance of high internal stresses account for the special mechanical effects of small amplitudes: in particular, the low rate of strain hardening, the resulting high totals of plastic strain to fracture, and the introduction of a distinctive mechanism of crack formation.

Crack Formation

Plastic deformation may give rise to more than one structural process capable of fracture; examples have been discussed by Orowan.[19] Broadly, they may be divided into processes resulting in high local stresses and processes resulting in local deterioration of structure. In practice, the process precipitating fracture in any particular case will be the one producing a critical condition most rapidly. It appears from the observations so far described that the process developing most rapidly at large plastic amplitudes is of the first kind, leading to fracture by increasing internal stresses. At the small amplitudes, however, where rapid hardening is ruled out, the operative process must be of the second kind. This is discussed next.

Deterioration of various kinds has formed the basis of several theories. In crude form, it underlies the attrition theory of Ewing and Humfrey,[20] based on their observation that fatigue slip bands sometimes extrude debris; it also underlies the theory of Rosenhain [21] that slip-band movements may produce amorphous layers. In more general form, it underlies the treatment by Freudenthal and Dolan,[22] who postulated a weakening of atomic bonds in disordered zones; and also the further suggestion by Freudenthal and Weiner [23] that deterioration by slip movements may be enhanced by accompanying heat bursts. In a form based on modern solid-state theory, it underlies suggestions that deterioration along slip zones may result from accumulations of point defects, such as vacant lattice sites, suggestions that are especially attractive in view of the large totals of plastic strain that may be imposed by small amplitudes and the high concentrations of point defects that therefore may be expected.

Unfortunately, it is not easy to test these suggestions because the structural changes preceding crack formation are on so small a scale.

Thus while it is not difficult by normal metallographic methods to resolve slip bands *laterally* in the metal surface and to show that continued cycles build up in them some disturbance which often becomes a crack, especially since fatigue slip bands are usually broad, it has been difficult to define this disturbance, because the microscope does not show what is happening *in depth*. Recently, however, this difficulty has been partly overcome by observing fatigue slip bands in section at magnifications effectively up

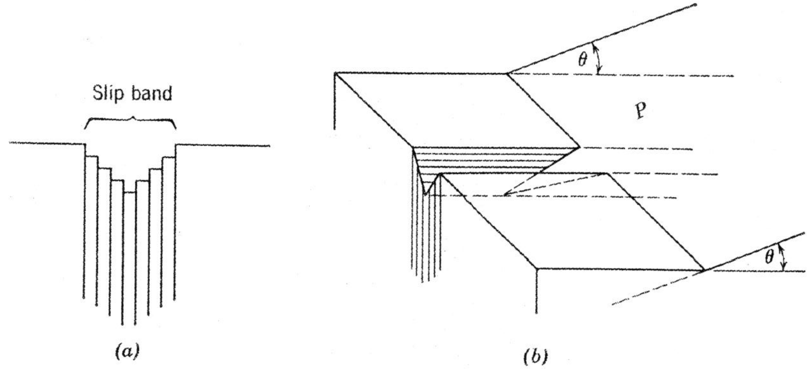


Fig. 7. To-and-fro fine slip in slip band, producing surface contour changes according to distribution across band, for example, a notch as in (*a*). Contour change magnified by taper sectioning plane *P*, as in (*b*).

to 30,000 times, and it will be convenient here to summarize the principal observations bearing on crack initiation.

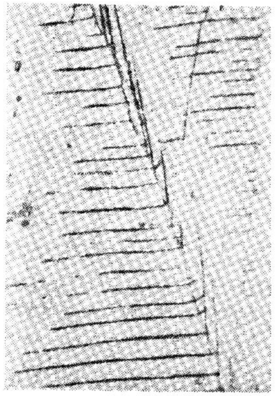
These magnifications are obtained by first taper sectioning a fatigue specimen at a specially small angle to the surface after protecting the slip bands by an electrodeposited layer.[24] The objective will be clear from Fig. 7. A surface disturbance in the slip band is magnified in depth along the taper-sectioning plane P by the cosecant of the section angle θ. By making this angle as small as 2° or 3°, we get a taper magnification t of 20 to 30 times, and then by examining the polished section by microscope at 500 to 1000 times magnification m, we get the high effective total $t \times m$. The examples reproduced are from annealed alpha-brass and copper specimens after various degrees of alternating torsion, which so far have been studied most. The electrodeposited coating is silver, and the etchant used on polished sections is ammonium persulphate, which leaves the silver untouched so that the microstructure of the sections and the contours of the slip bands stand out against a white background.

Disturbances thus observed are described below, broadly in the order that they appear during progressive cycling. For ease of comparison they are illustrated, unless otherwise mentioned, by specimens exhibiting an expected life of 5×10^6 cycles and are viewed at magnifications of about

20,000 times. The torsional amplitudes are given in terms of the angle of twist applied to standardized specimens of test length 1.75 in. and diameter 0.2 in.

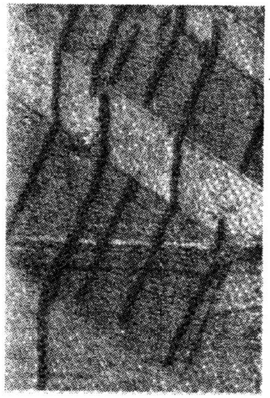
Stage 1

Specimens after 1000 or so cycles develop slip zones of abnormal distortion. These show as traces, readily brought up by the etching, where active slip planes have crossed the section. Figure 8 shows typical traces, and Fig. 9 is of interest in showing how such traces persist after deep etching.



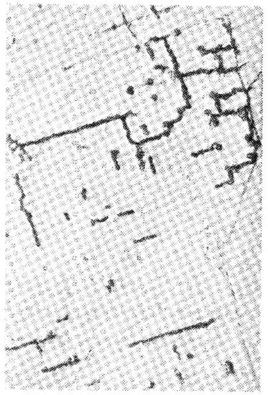
100 μ

Fig. 8. Typical etched-up slip traces after cyclic straining at small amplitudes.



10 μ

Fig. 9. Traces persisting after deep etching.

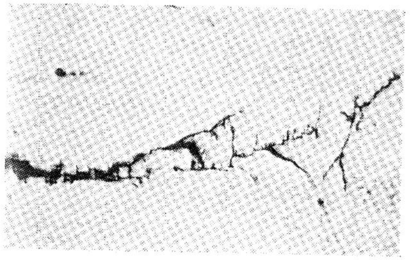


10 μ

Fig. 10. Shorter traces after large-amplitude straining.

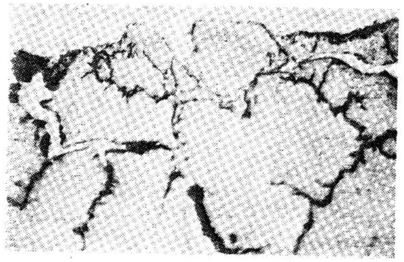
That the distortion is abnormal is shown by the fact that these traces etch up readily only in specimens subjected to small amplitudes corresponding to the F part of the S-N curve. At larger amplitudes the traces become shorter, and at large enough amplitudes they no longer appear; thus Fig. 10, taken from brass after an amplitude $\pm 20°$, large enough to cause fracture after only 50,000 cycles, shows how the traces have contracted to short remnants. This effect, furthermore, provides independent proof that the small amplitudes evoke a special mechanism of fracture; for, as the traces disappear with increasing amplitude, the mode of fracture alters. At small amplitudes, as illustrated later, fracture begins by the active slip zones turning into fissures. At large amplitudes, where there are no active zones to turn into fissures, fracture is irregular; cracks branch at once into noncrystallographic paths. A typical example of

such a crack, and of the now-absent slip traces, is shown in Fig. 11, taken from brass after an amplitude of $\pm 30°$, which is large enough to cause failure after only 10^4 cycles. The irregularity is even clearer when this kind of crack is viewed after taper sectioning at the surface where it begins; this may be seen from Fig. 12, obtained at about 20,000 times magnification from the same specimen.



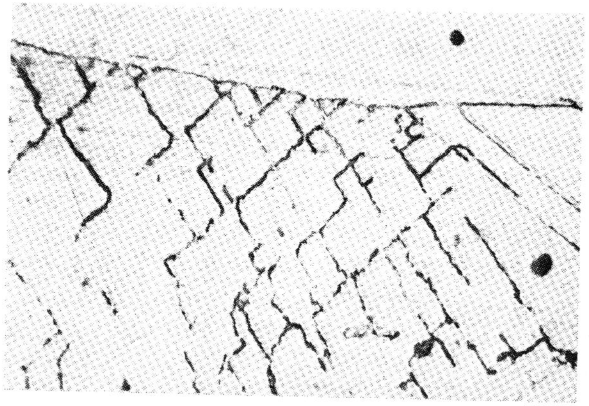
10 μ

Fig. 11. Irregular cracking (and now-absent slip traces) at large-amplitude straining.



10 μ $t = 20$

Fig. 12. Similar crack taper sectioned at specimen surface.



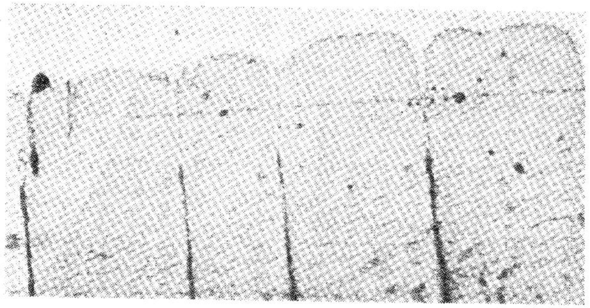
100 μ $t = 20$

Fig. 13. Copper after $\frac{1}{10}$ of life. The active slip zones are limited in length.

A final feature of these traces is that they are limited in length, usually shorter than a grain, thereby implying that the abnormal distortion resulting from the to-and-fro slip movements is usually concentrated on limited lengths of a slip plane. These limited "active zones" are shown up clearly in grains undergoing cross slip, for then the active zone may jump from one slip system to the other in steps that are obviously limited, as seen in the example shown in Fig. 13. Attention is drawn to this feature because it will appear later that the active zones turn into fissures of correspondingly limited length.

Stage 2

An active slip zone may develop a surface notch, peak, or intermediate irregular contour where it meets the specimen surface. The disturbances are about 10^{-4} to 10^{-5} cm in depth and appear as early as $\frac{1}{10}$ of the expected specimen life. Figure 14 shows typical notches, Fig. 15 a typical peak, and Fig. 16 intermediate jagged contours.



$t = 20$

10 μ

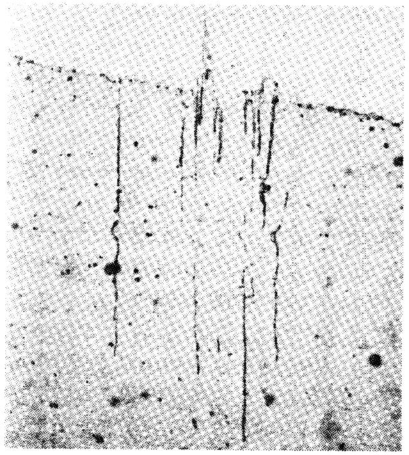
Fig. 14. Copper after $\frac{1}{10}$ of life. Slip bands developing surface notches.

The explanation appears to follow readily from the feature already discussed, that the slip movements at small plastic amplitudes are made up of fine slip translations, for these will build up in a slip-band contours of the kind observed according to the way the backward and forward translations become distributed across the band during progressive cycles. This was suggested by Fig. 7.

These surface disturbances do not seem to grow much with continued cycling. Instead the next stage occurs.

Stage 3

The surface disturbances appear to penetrate the active slip zones and turn them into sharp fissures. Figure 17 shows an example. This stage

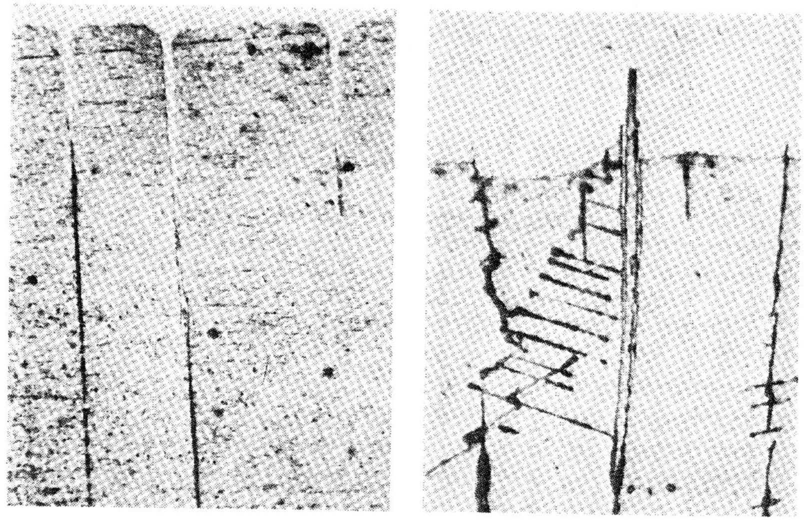


10 μ $t = 20$

Fig. 15. Slip bands developing typical peak.

10 μ $t = 20$

Fig. 16. Intermediate contours.



(a) 10 μ $t = 20$ (b)

Fig. 17. Copper at $\frac{1}{10}$ of life: (a) notches penetrating slip trace to form fissures, (b) fissures formed down each side of slip-band peak.

also begins early in the specimen life, almost concurrently with the preceding one.

It is of interest that, as already indicated, the fissures apparently run only to the end of the limited active zones in which they start. There they seem to stop, since neither zones nor fissures on the average become longer after continued cycles; the further cycles merely open more slip zones into fissures. This limit to propagation is shown clearly in grains where the active slip zones form a stepped pattern of the type that was illustrated by Fig. 13, for the fissures follow a corresponding pattern, sharply changing direction from one step to the next instead of propagating in one direction. The effect is of interest therefore in showing that the active slip zones provide ready-made, but limited, paths of weakness. It shows also that a specimen may become suffused with apparently nonpropagating fissures for most of its life.

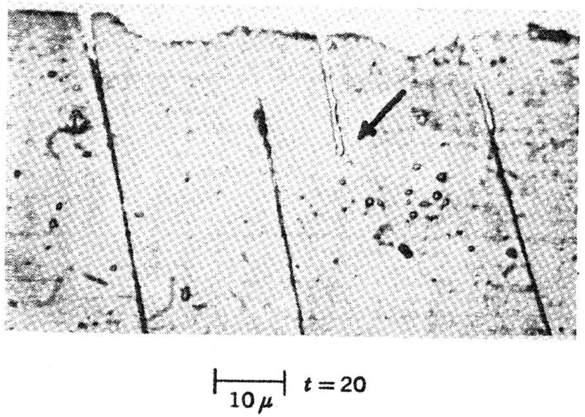


Fig. 18. Effect of heating fatigued specimen before taper sectioning. Copper taken to $\frac{1}{10}$ of life. Notch from which slip trace has been dispersed after 1 hr at 600°C.

Some indication of when an active slip zone becomes a fissure can be obtained by heating a fatigued specimen before sectioning and etching. It is found first, that some surface disturbances appear without an attached slip zone (Fig. 18); such zones, being thus dispersible, are presumably at the stage of lattice distortion only; second, other zones still etch up, but now they do so as discontinuous links of what appear to be cavities (Fig. 19); these must be zones that have turned into fissures but which are still narrow enough to sinter at intervals along their length; and finally, some zones persist as obvious fissures, evidently having become wide enough to resist sintering (Fig. 20: The fissure shown has

survived heat-treatment severe enough to start recrystallization at its root, though such recrystallization is rare). It is possible to show by experiments of this kind that with continued cycling, the zones of dispersible distortion become fewer and the permanent fissures more numerous.

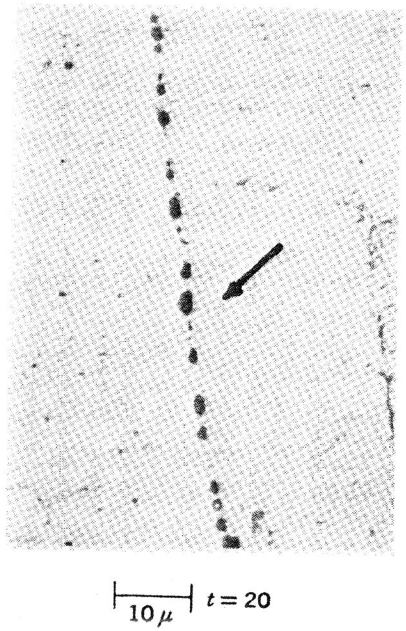


Fig. 19. Effect of heating fatigued specimen before taper sectioning. Copper taken to $\frac{1}{10}$ of life. Slip trace balled-up into cavities because of partial sintering of fissure.

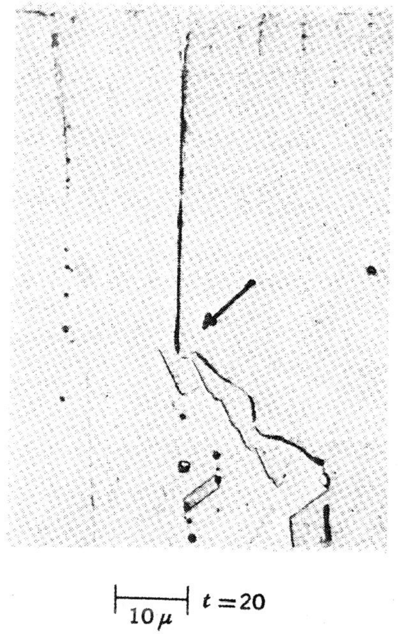


Fig. 20. Effect of heating fatigued specimen before taper sectioning. Copper taken to $\frac{1}{10}$ of life. Fissure surviving heating sufficient to cause rare recrystallization at its root after 1 hr at 950°C.

Stage 4

The fissures, loosening cohesion between neighboring blocks of metal, permit whole blocks to move relative to one another. These movements occur actively throughout the specimen life.

Figure 21, for example, shows the beginning of block movement; Fig. 22 shows the movements crowding together and raising a surface peak, which readily disintegrates; Fig. 23 shows how fissures, running nearly parallel to a surface, start local surface disintegration; and Fig. 24 shows how fissures in grains undergoing cross slip may intersect and

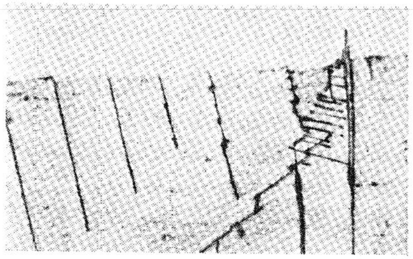


Fig. 21. Copper at $\frac{1}{10}$ of life, showing beginnings of block movement.

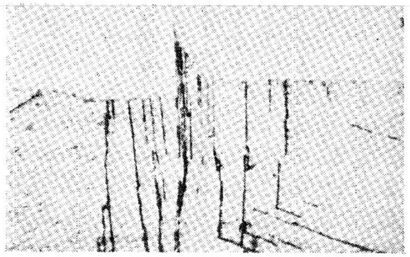


Fig. 22. Brass at half of life, showing block movements crowding together to raise peak.

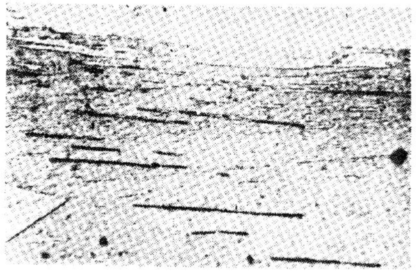


Fig. 23. Copper at half of life, showing surface disintegration by fissures parallel to surface.

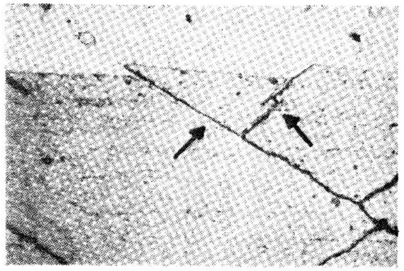


Fig. 24. Brass at half of life, showing intersecting fissures undermining surface block.

undermine surface blocks. Thus these examples show that the fissures start a secondary but effective phase of structural deterioration.

Final Stage

A crack forms at one fissure and traverses a grain by jumping from one fissure to the next in the manner illustrated by Fig. 25. (The term crack is used here for convenience to distinguish this propagating process from the essentially nonpropagating fissure.)

This stage, compared with the preceding ones, is noncrystallographic. The path of the crack appears to be crystallographic only so far as it can follow a fissure. Once it leaves a fissure, or extends beyond the end of one, it diverges irregularly, though it may include part of a grain boundary if one is near. This stage, too, ends the processes leading up to crack formation, with which this chapter is primarily concerned. The rest,

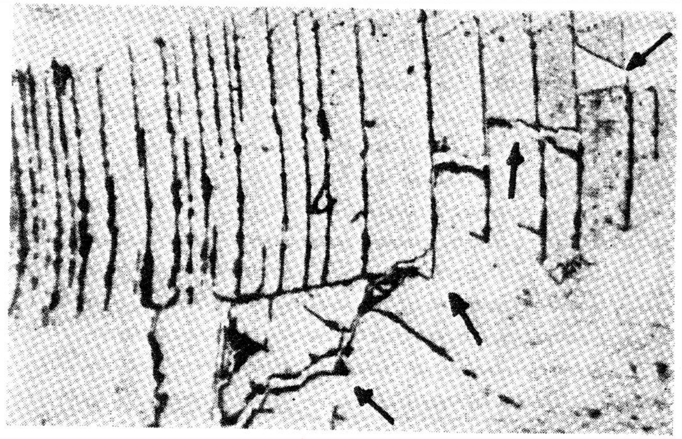


Fig. 25. Brass approaching final fracture, showing crack propagating by jumping from fissure in structure also showing marked block movements.

dealing with the spreading of the cracks thus initiated and with final collapse of a specimen, is already familiar to engineers.

Present Position

The discussions have been confined to what was defined as pure fatigue, since this must be elucidated first, and to concepts for which there is tangible evidence. Still more fundamental experiments are needed, but some firm conclusions can be drawn from work already done.

First it is desirable to recognize, as the evidence shows, that two diverse mechanisms have formerly been classed together under the same heading of fatigue. One predominates at large plastic amplitudes when the S-N curve descends rapidly towards the N-axis, and the other at the small amplitudes where the curve tends to turn parallel to the N-axis. These were termed, respectively, the H and F mechanisms.

The H or hardening mechanism is essentially that of unidirectional or static deformation. The evidence indicates that it produces fracture by progressively building up internal stresses, though it does so less efficiently than in static deformation. It accords with those treatments by solid-state theory that account for progressive internal strains by the continued holding-up of dislocations at lattice obstacles. The mechanism may therefore be described as delayed static fracture.

It is curious that though this H mechanism has been the one largely advocated by engineers, it applies most to the early part of the S-N curve, which is of least engineering interest. Of more practical interest is the later part, when the fracture is disproportionately delayed. Then the F mechanism takes over. Here the evidence shows that the abnormally high totals of plastic strain which then result produce, along limited lengths of the slip planes, zones of abnormally high lattice distortion; and that these active zones, at an early stage of specimen life (not more than $\frac{1}{10}$), begin to turn into observable fissures of correspondingly limited length. The lattice distortion must still be identified in detail, but there is the indirect evidence that, since it is not accompanied by significant hardening, it must be due to concentrations of point defects, such as vacant lattice sites or interstitial ions. The transformation to fissures also needs further study, but a direct explanation is suggested by observations that the to-and-fro fine slip in the active zones gives rise to striking changes of surface contour ranging from sharp peaks to sharp notches and by indications that these notches open up the active zones by a process of simple penetration. The fissures are not cracks by our definition because they are strictly crystallographic in the sense of being confined to slip planes and because they do not appear to propagate readily beyond the active zone in which they form. The crack is a final irregular phase, propagating initially by jumps from fissure to fissure through a structure already weakened by their presence and soon becoming a macroscopic phenomenon.

The observations thus provide a guide for interpretations of fatigue in terms of solid-state theory. The essential feature of the F mechanism, the avoidance of influential piling-up of dislocations, follows from the reversal of dislocation paths before they exceed the spacing of relevant lattice obstacles. The concentration of point defects along the active slip zones is to be expected from continual passage of dislocations through an imperfect lattice, as shown by Seitz.[25] How these concentrations may turn into fissures is a moot point for the theory; suggestions that fissures form simply by coalescence of vacancies, for instance, need confirmation, and it is possible that the explanation lies in the penetration by surface disturbances, already suggested on the basis of observation. The surface disturbances themselves are more easily explained: They are either a consequence of the to-and-fro fine translations in the fatigue slip band, as proposed by the writer,[26] or a consequence of the more intricate dislocation movements proposed more recently by Mott[27] and by Hull and Cottrell.[28] There is, however, no evidence in the experimental work for the refined effects also implied in these later suggestions or for their further implication that the surface disturbances alone suffice to start cracks. It appears, rather, that equally essential to the fatigue process is the abnormal distortion arising in the active slip zones, which is an independent phenomenon. However, though many details call for further study, it now seems possible by combination of theory and experiment to account for the main processes initiating fracture by pure fatigue.

The next step of immediate practical interest will be to find how these basic processes are modified in more complex metals and by more complex systems of stressing.

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DISCUSSION

N. A. Tiner, *Douglas Aircraft Company, Inc.* It would be helpful if Wood clarified a specific structural effect which he briefly mentioned in this chapter but fully reported elsewhere.[1]

Wood's Fig. 3 shows that the total plastic strains are reduced during fatigue testing of brass by coating specimens with mercury. This effect is associated with the penetration of mercury into the grain boundaries, a phenomenon well publicized since the comprehensive work of Moore and his associates on season cracking.[2] The latter authors attributed this intercrystalline attack to the special etching properties of the grain boundaries. The observations made in our laboratory indicate that mercury, under progressively increased tension loading, causes brittle (cleavage) fracture in single crystals of pure zinc, where no grain boundaries and their special properties exist.

It would be of interest to know (1) whether a microexamination of the mercury-coated specimens of brass has been made in order to clarify how mercury affects the surface disturbances during fatigue testing and (2) whether it is necessary to bring up the effect of grain-boundary penetration to explain the fatigue behavior of mercury-coated samples, particularly at small amplitudes of alternating torsion where the so-called *F* mechanism is operative.

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2. H. Moore, S. Beckinsale, and C. E. Mallinson, *J. Inst. Metals*, **25,** 35 (1921).

W. A. Wood (Author's Reply). Mercury embrittlement is not necessarily related to the grain boundaries but produces roughness and notches within the grains.