18. Some Observations on the Early Stages of Fatigue Fracture

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ABSTRACT

Aluminum that is fatigued at low stresses for prolonged periods develops "pits" on the surface; nickel does not show this effect, and copper does so only occasionally. In polycrystalline copper, twin interfaces are a favored site for the initiation of fatigue fracture. Extrusions observed on a single crystal of copper showed certain regularities of distribution, which are discussed.

Introduction

During the past few years, a number of researches have led to the conclusion that, in some materials at least, the process that leads eventually to the complete fracture of a fatigue specimen has a detectable beginning at a very early stage of the test. Estimates of the earliest stage at which some singularity with the essential characteristics of a "crack" can be detected range from 10% to 1% of the fatigue life, depending on the material being investigated, the techniques used, and possibly, the skill of the experimenter. We may perhaps mention by way of illustration some earlier work by the author and his colleagues on specimens of annealed copper.[1] The surface, initially electropolished, showed many characteristic slip marks as the fatigue test proceeded. Most of these could be readily removed by a further electropolish, but a few persisted after this treatment. Such persistent bands could be detected quite early in the test, and it was usually from such a band that the fatal crack grew. More recent observations of the same kind by

Wadsworth and Hutchings[2] have led to the detection of such markings after only 1% of the life, and to the belief that, at that stage, they already have the nature of microcracks.

This state of affairs suggests that it would be of interest to examine more closely the earliest stages of a fatigue test to try to obtain some clue as to the nature of the process that causes the first crack to form. The present chapter reports the results of a number of investigations of this kind. The results raise more problems than they solve, but it has been thought fit to publish them as they stand, in the hope that discussion may help to clarify the situation.

The Role of Twin Boundaries

The material used in the work mentioned above was mainly polycrystalline copper with a grain size of about 0.1 mm. The material naturally contained numerous annealing twins, and it was remarked that several of the persistent markings involved portions of twin boundary in their length and might well have started in such twin boundaries. The twin interfaces in copper are, like the slip planes, (111) planes, so that it is not strictly possible to say whether a particular mark is really in a twin boundary or is a slip line lying very close to a twin boundary. However, a re-examination of the photographs taken showed that, in fact, a fairly large proportion of the persistent slip marks were associated with twin boundaries in this way.

Further observations by Bown (unpublished) have confirmed this view. The materials, apparatus, and techniques used were the same as before, but attention was particularly directed towards the behavior at low stresses, with the result that most of the specimens remained unbroken at the end of the test. The lowest stress used was 6.19 kg/mm^2, and this specimen was unbroken after 150×10^6 cycles. (Similar material gave a fatigue life of 10^7 cycles at a stress of about 8 kg/mm^2.) Very few surface marks of any kind were produced: Of the three that were observed, two lay in twin boundaries, and both of these were noticed after only 10^6 cycles of stress.

Another specimen survived 65×10^6 cycles at 7.31 kg/mm^2. After 0.22×10^6 cycles, only a few slip marks were visible on the surface, and most of these lay in twin boundaries. After 25×10^6 cycles, many more slip marks were visible, but most of them were collected in diffuse patches that were removed by a light electropolish. Of those that survived such a polish, a large proportion lay along twin boundaries. Similar results were obtained from a third specimen after 0.18×10^6 cycles at 7.38 kg/mm^2; a large fraction of the persistent slip marks observed in this specimen were associated with twin boundaries. Figure 1 shows an example, taken from this specimen after 10^7 cycles.

A few specimens of pure vacuum-melted nickel were examined, and these too showed that a twin boundary was a favored site for the initiation of a fatigue crack. This was true both at low stresses (failure in 10^7 cycles) and at higher stresses (failure in 0.2×10^6 cycles), although the crack that proved fatal frequently started not in a twin boundary but

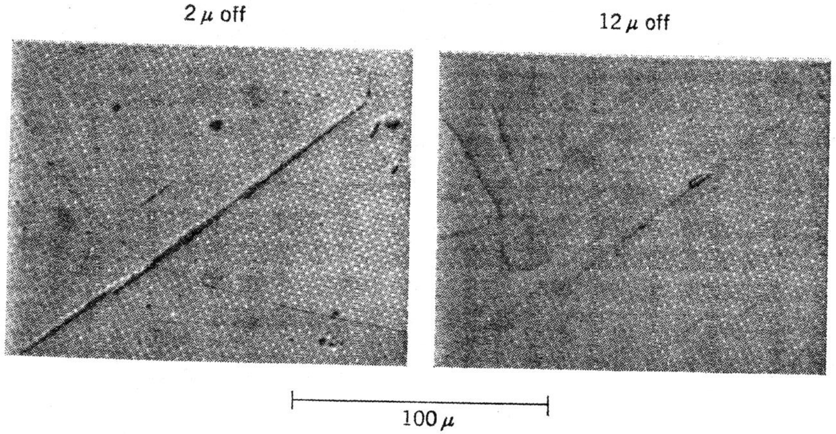

Fig. 1. Twin-boundary markings in fatigued copper.

in a slip band across a particularly large grain. It might be added that no twin-boundary markings were observed in nickel specimens of commercial purity.

Among other workers who have reported the formation of fatigue cracks in twin boundaries, we may mention Stubbington and Forsyth,[3] who worked with alloys of copper and nickel as well as the pure materials, Kemsley,[4] who experimented on copper, and Wadsworth and Hutchings,[2] who observed a similar behavior in gold.

Since it appears that the origin of a fatigue crack is connected in some way with the to-and-fro slip movements that take place in a fatigue slip band, we may first look for an explanation of the importance of twin boundaries in terms of whether slip takes place in or near them. First, we may ask if the stresses near a twin boundary are in any way exceptional. In a polycrystalline sample of any material that is elastically anisotropic, there will be local stresses near all grain boundaries, the effects of which can be seen in the nonuniform distribution of slip. The same will be true, in general, of a twin boundary, although, in view of the symmetry of the relation between the two crystals, it may happen that many of the elastic constants (referred to axes in and normal to the twin boundary)

will show no discontinuity at the boundary. As far as is known, the necessary calculations have not been carried out for face-centered cubic twins.

Alternatively, we may ask whether the dislocation structure of the material is exceptional near a twin boundary. Apart from the possibility of image forces, which may arise because of the changes in elastic constants already mentioned, there might be abnormal structures in the dislocations themselves. In the Appendix to this chapter, a detailed analysis is given of the dislocation geometry in a twin boundary of a face-centered cubic crystal. Perhaps the most interesting result is the conclusion that, if a dislocation lies in a $\{111\}$ plane immediately adjacent to the boundary, the two partials into which it dissociates are likely to repel one another to considerable distances. This is because, in these circumstances, the ribbon of stacking fault, which is bounded by the two partials, takes the form of a section of twin interface that is displaced by one atomic layer either above or below its normal position; thus the surface energy of the interface has the same value between the two partials as it has outside them. A rough calculation suggests that the separation of the two partials is likely to be of the same order of magnitude as the distance between the nodes of the dislocation network. This unusual configuration may well affect the value of the critical stress necessary for the operation of a Frank-Read source and seems likely to make it smaller, although again, the calculations have not been carried out. The argument given in the Appendix also brings out the known fact that if a dislocation crossing a twin boundary has a Burgers vector with a component normal to the boundary, there will be a "twinning dislocation" lying in the boundary and starting from the crossing point. In a face-centered cubic crystal, the twinning dislocations have the character of ordinary Shockley partials, and their presence seems unlikely to impede seriously the movement of other dislocations in the twin interface.

Taking all things together, it would not be unreasonable to conclude that slip in a twin boundary should be easier than slip in the body of a crystal. In unidirectional stressing, however, twin-boundary slip is so far from being a preferred mode of deformation as to be rather uncommon. Thus if the explanation of the behavior of twin boundaries in fatigue is to be found on these lines, it must depend on some more subtle factor that is not applicable to unidirectional stressing. Alternatively, the key may lie, not in the slip process itself, but in subsequent events that lead up to crack formation. Two possibilities are discussed in the next two sections of this chapter, but neither seems particularly relevant to twin boundaries. A third possibility is the aggregation of vacancies into disk-shaped cavities: In a twin boundary, the consequences of this would not be the same as they would be if it occurred in the middle of a grain, but further work is needed before any definite conclusions can be drawn.

Pits

During observations on fatigue specimens of aluminum, Harries [5] (see also Smith [6]) observed that when the stress was small, the surface slip markings were often discontinuous and that, particularly after electropolishing, they often appeared as a row of dots. Observation with an electron microscope confirmed that these were, in fact, pits in the surface of the metal. It was also demonstrated that they were not artifacts produced by the electropolishing process.

These results have also been confirmed by later observations by Bown (unpublished). The material was again aluminum, but the manner and speed of testing were different (push-pull at 1000 cps as against reverse bending at 1500 cpm), and the methods of surface preparation were not the same. In view of Harries' results, the stresses used were very low, and no specimen ever broke. Five million cycles at 1.11 or 1.17 kg/mm^2 failed to produce any surface markings of any kind, but a fair number of marks were visible after the same number of cycles at 1.39 kg/mm^2. The most extensive series of observations was accordingly made at this stress on a specimen that had a fairly large grain size. A total of 75×10^6 cycles was reached after interruption at 1, 3, and 9×10^6 cycles for observation of the surface. After 10^6 cycles, there were many slip marks, some with the characteristic appearance of fatigue slip bands and others of a more irregular and flocculent appearance. After removing about 4 μ from the surface, traces of the fatigue slip bands were still visible in many places.

After 3×10^6 cycles, the typical fatigue slip marks had reappeared in approximately the same positions as they had previously occupied, and this time the removal of 10 μ did not suffice to remove them completely, although most had disappeared when 8 μ had been polished away. However, it was now observed that one or two of them showed here and there the characteristic spotty appearance remarked by Smith and Harries. After 9×10^6 cycles, this tendency was more marked even before polishing; when 12 μ had been polished away, many markings were still visible, and many of these were very clearly broken up into spots. Finally, after 75×10^6 cycles, some grains were entirely covered with slip marks. By the time that 6 μ had been polished away, the remains showed the typical pitted appearance (although the size of the individual pits had doubtless been seriously modified by the polishing process). Some of these pits persisted down to about 30 μ, and when, at later stages of polishing, they became less numerous, it was possible to follow individual pits from one level to the next (Fig. 2).

Attempts were also made to observe similar structures in copper and nickel. In nickel, no pits were observed on any specimen; in particular, one that had survived 10^8 cycles at a low stress (11.2 kg/mm^2) showed none, although some slip marks caused by the fatigue were visible. Nor was much greater success achieved with copper. Occasionally a slip band with a slightly beaded appearance was found, and occasionally a slip mark, on electropolishing, broke up into a series of short segments. The greatest success was with one specimen that survived 10^8 cycles at 7.6 kg/mm^2. Removal of 14 μ by electropolishing revealed numerous persistent marks, many of which might well be said to consist of rather irregular rows of closely spaced pits. A further 0.36×10^6 cycles caused the reappearance of slip bands and showed that the persistent marks did lie along such bands. Further electropolishing emphasized the discontinuous nature of the markings; although many had disappeared by the time that a further 12 μ had been removed, those that still remained could be clearly followed from one photograph to the next (Fig. 3). Another specimen that survived 150×10^6 cycles at 6.19 kg/mm^2 showed only a few rather shallow markings which could not, with any degree of conviction, be called pits.

We can probably safely conclude from these observations that prolonged stressing at a low level is a necessary, but not a sufficient, condition for the occurrence of pits in a face-centered cubic material. Attempts to observe similar features in specimens stressed for fewer cycles at higher stresses were unsuccessful, although Smith [6] reports having seen them on high-stress specimens of aluminum. Pits have been described in aluminum by Forsyth,[7] who also mentions a similar structure seen in silver chloride in front of a fatigue crack. It has been suggested that the formation of pits might be an early stage in all fatigue failure, the pits coalescing to give a continuous crack. The evidence for this is very scanty, and it is certainly not established that all cracks pass through this stage. Even so, the mechanism of the formation of pits is of interest. It is noteworthy that in all recorded cases they have formed at a free surface: No new pits ever appear as more of the surface layers are polished away.* A possible connection with the intrusions mentioned in the following section should not be ruled out.

* A recent paper by Cina [8] reports the occurrence of a similar spotty appearance of fatigue slip bands, revealed by etching in the interior of an austenitic Ni-Mn-Cr steel after fatigue. If the resemblance proves to be more than superficial, further investigation would be very valuable.

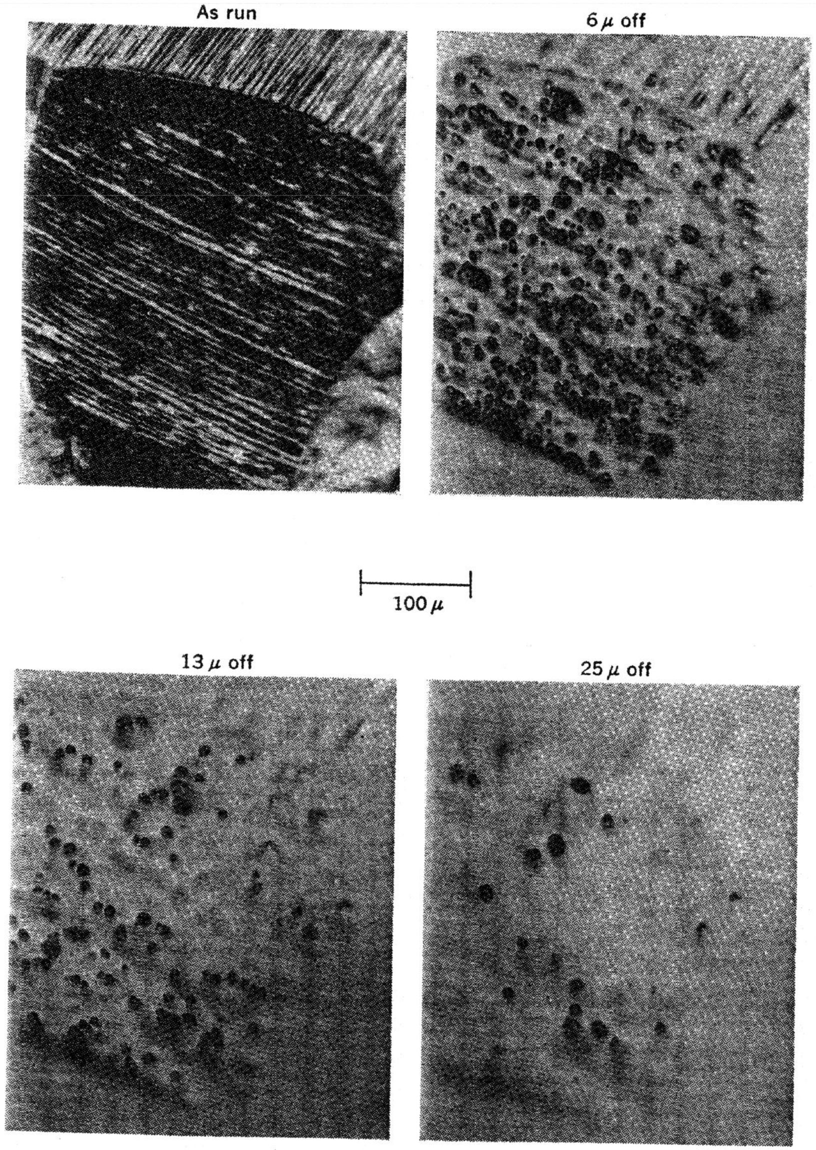

Fig. 2. Pits in aluminum after 75 × 10⁶ cycles.

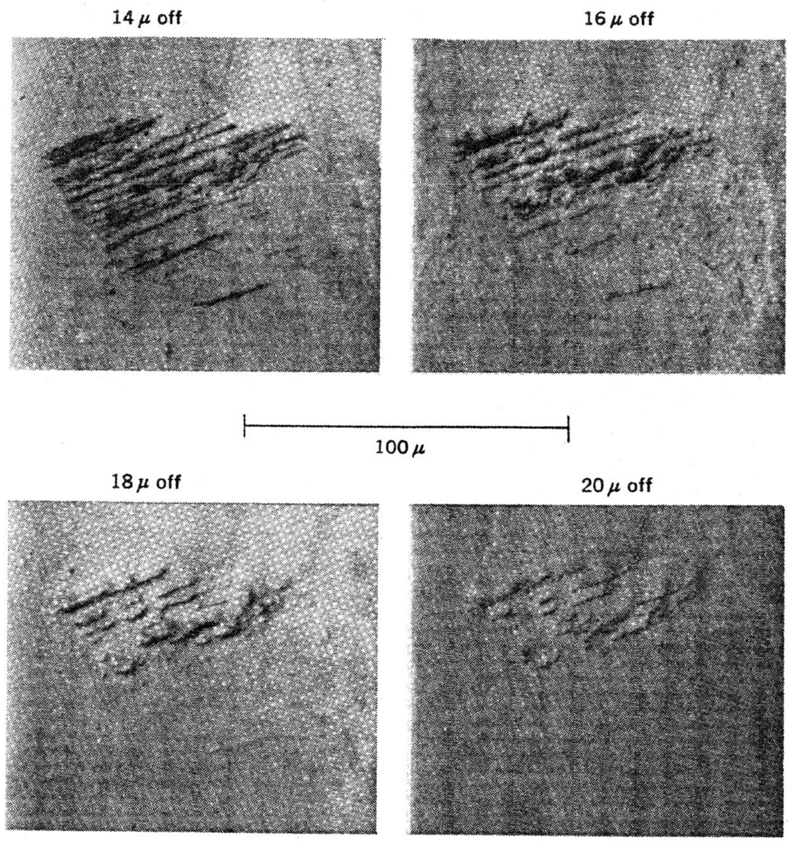

Fig. 3. Pits in copper after 103 × 10⁶ cycles.

Extrusions

The third metallographic feature to be considered has become known as "extrusion." In 1953, Forsyth [9] reported the appearance, on the surface of an aluminum alloy fatigue specimen, of a thin ribbon of metal which appeared to have been extruded from a slip band. Since that time many more similar observations have been made on pure metals and alloys, on annealed and cold-worked materials, and over a wide range of temperature. The inverse phenomenon of "intrusion" has also been described on silver chloride [7] and copper; [10] an intrusion is a narrow crevasse that appears in a fatigue slip band.

We have made some observations on extrusions from a single crystal of copper, which may throw some light on the mechanism involved.

The crystal was of oxygen-free, high-conductivity copper, grown from the melt and tested in push-pull at a frequency of about 1000 cps. The cross section was circular, and the form of the gage length is shown in Fig. 4*a*; the central portion was given a metallographic polish, concluding with an electropolish. The fatigue test ended when the main crack had extended about one third of the way around the circumference. Microscopic examination of the surface of the cracked specimen showed that

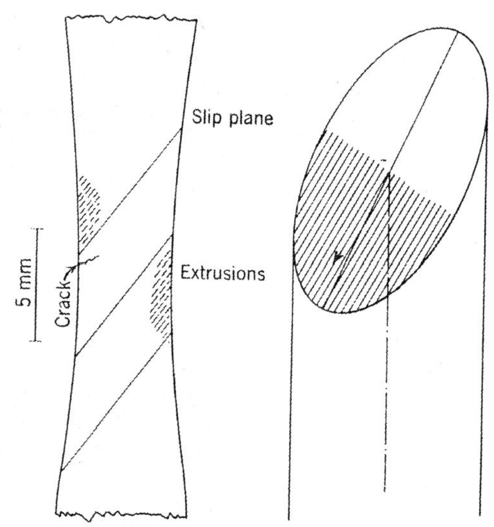

Fig. 4. Specimen for extrusion studies. (*a*) Form of gage length. (*b*) Segmented slip on most favored slip system.

the whole was covered with slip markings and that in two large areas numerous extrusions were visible. Traces of all four slip systems were found, although two were much more in evidence than the others. Measurement of the slip lines served to determine the orientation of the crystal with an accuracy of about $\pm 1°$.

Each of the two patches of extrusions extended for about 90° around the circumference and was situated as shown in Fig. 4*a*. A series of photomicrographs was taken at azimuthal intervals of about 7°, covering one of the regions in which the extrusions were most marked. Figure 5 shows examples of these photographs; the slip plane involved was that one for which the resolved shear stress factor was the greatest (= 0.457). It will be noticed that not only do the extrusions lie in rows along the slip bands, but there is some regularity in their spacing. To obtain

quantitative evidence on this point, the average spacing was measured on ten photographs by measuring the total length of a number of the rows. The results are shown in Fig. 6. The "spacing" is the average distance, measured along a slip band, between the center of one extrusion and the

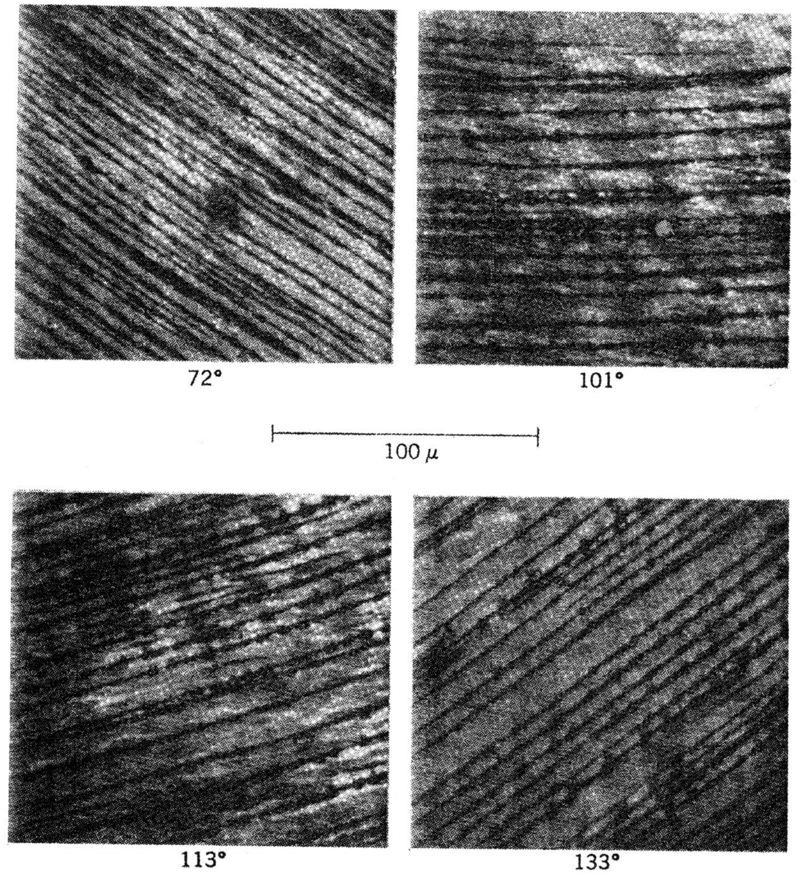

Fig. 5. Extrusions on copper crystal.

next in the same group. The consistency of the results in Fig. 6 is sufficient to justify a belief that the spacing varies systematically with azimuth around the crystal.

An attempt was also made to measure the density of extrusions by placing at random a window of known area on those parts of the photographs that were in good focus and counting the number of extrusions visible. This was done two or three times on each print and an average taken, but even so, the statistics are not very reliable. The results are

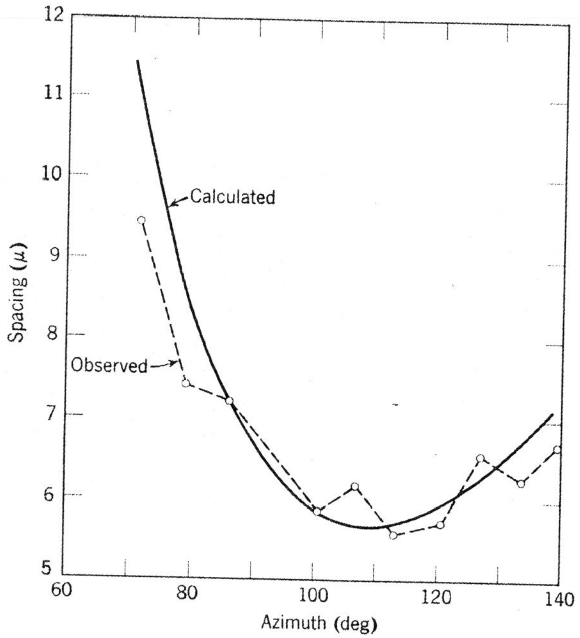

Fig. 6. Spacing of extrusions on copper crystal.

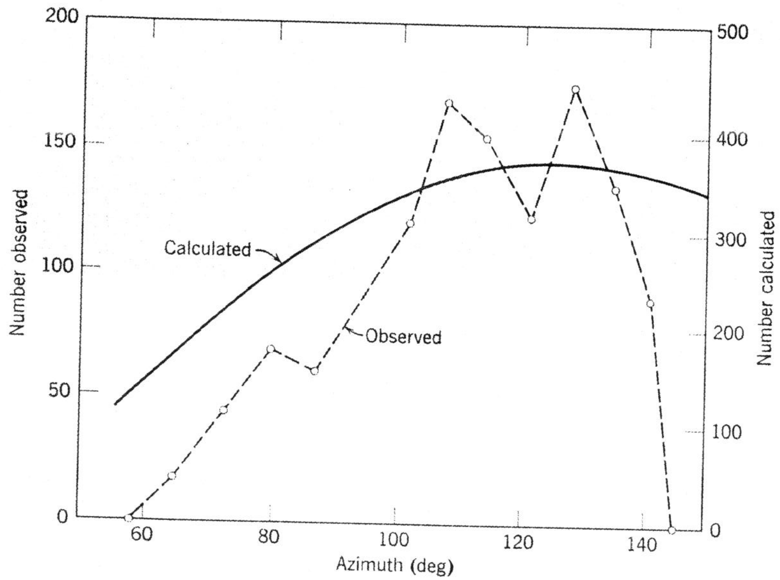

Fig. 7. Density of extrusions on copper crystal.

shown in Fig. 7. Finally, it was observed that the *size* of the extrusions (as distinct from their spacing) was greatest towards the center of the range in which they were visible, and that they were to be seen most clearly near the azimuth at which the preferred slip direction met the specimen surface.

This suggested that it was the operation of the most-favored slip system which was determining the spacing. Suppose one assumes that a slip front, moving towards the surface along the slip plane, is broken up into a series of segments which are equally spaced along a direction lying in the slip plane and normal to the slip direction (Fig. 4*b*). If these segments determine the appearance of extrusions on the surface, then it is a matter of straightforward geometry to calculate the resulting spacing of the extrusions along the slip bands in terms of the angles involved and the spacing of the segments. The full line in Fig. 6 shows the result of such a calculation for the case in point, where the spacing of the segments (5.7 μ) has been chosen to fit with the observations. In view of the scatter of the experimental points, all that can be said is that they are not inconsistent with a description along these lines.

We can also calculate the density of extrusions if we know the spacing of the active slip planes along the axis of the specimen. An average value for this quantity was obtained by counting the number of slip bands crossing the center line of each photograph. Combined with a segment spacing of 5.7 μ as already found, this gives the calculated densities as shown by the full curve in Fig. 7. The calculated values are, of course, much larger than those observed, since they represent the result that would be obtained if every slip plane produced extrusions along its whole length. The agreement between the general form of the two curves in Fig. 7 is not very good, but the divergencies are worth comment. In the center of the range, the fluctuations of observed density are to be attributed mainly to corresponding variations in the number of active slip planes that happened to cross those parts of the photograph that were in good focus. At the two extremes, and particularly near 140°, this is not true; there was a marked tendency for the slip-plane spacing to be small in these regions. In spite of, or because of, this, the number of visible extrusions falls rapidly. It has already been remarked that their size is also less at the two ends, and it may be that when the slip bands were closely spaced, the extrusions were too small to be clearly seen.

Suggested explanations of the extrusion effect have approached the problem from two viewpoints. On the one hand, we may think in terms of a soft region in the crystal formed either by the overaging of an alloy, or by recrystallization of a pure metal, or by some other mechanism facilitated by the to-and-fro movements in the fatigue slip bands. Alternatively, we may discuss the process in terms of dislocation movements. The above observations suggest that, in this instance at least, the formation of extrusions is intimately connected with the operation of the most favored slip system, and we may seek an explanation in terms of dislocations on this basis.

Two general mechanisms have already been suggested. Mott,[11] extending an idea of Fujita,[12] considers two rows of edge dislocations of opposite sign moving in opposite directions on nearby glide planes (Fig. 8*a*), the two rows having started from two sources S_1 and S_2. The combined tensile stress between the leading members can be sufficient to give rise to a cavity in the crystal, which is enlarged by the arrival of further dislocations (Fig. 8*b*). The screw sections of the dislocations can become detached from the cavity, although this requires a nonconservative motion of the jogs at J, J' (Fig. 8*c*). If some screw dislocations remain, such as AA' (Fig. 8*d*), connecting the cavity with a free surface of the crystal, then, under the action of an alternating stress, they can move around the path $BCDE$, cross slipping from the upper to the lower slip plane at BC and DE. This will lead to the continuous extrusion from the surface of a section such as $B'C'D'E'$, if the direction of motion is correct.

Cottrell and Hull [13] consider the action of two sources of dislocations belonging to different slip systems. The most favored source (S_1 in Fig. 9*a*) produces a slip step on the surface at P during a tensile half-cycle. At a slightly greater stress during the same half-cycle, the second source S_2 produces a second step at Q (Fig. 9*b*). During the following compression half-cycle, the source S_1 produces a surface step of opposite sign at P' (Fig. 9*c*); owing to the intervening action of S_2, this is not in the same plane as the first, and thus an intrusion results. The subsequent operation of S_2 produces an extrusion QQ' (Fig. 9*d*) in a similar manner. The screw sections of the dislocations cancel one another by cross slip (apart from a little difficulty where the screws on the two glide planes intersect), while the edge sections in the interior of the crystal cancel one another in a straightforward way. This mechanism has the attraction that all the dislocation motions are conservative, and no residual stresses are left behind in the crystal. The atoms needed to build the extrusion are made available from the cavity representing the intrusion. It would predict intrusions and extrusions occurring with comparable frequency but not, in general, in the same slip band. There is insufficient experimental evidence upon which to judge the first prediction, but there is some indication [2] that both intrusions and extrusions can sometimes be found in the same band.

The Cottrell and Hull mechanism requires the occurrence of slip on

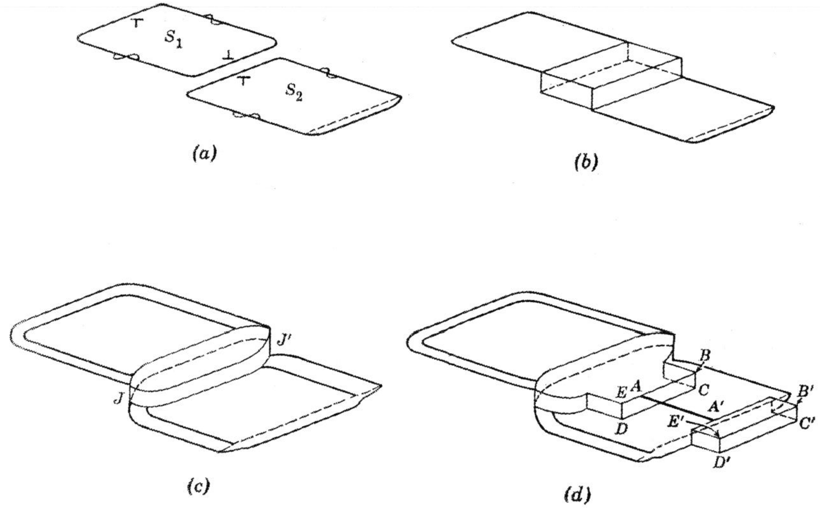

Fig. 8. The Mott extrusion mechanism.

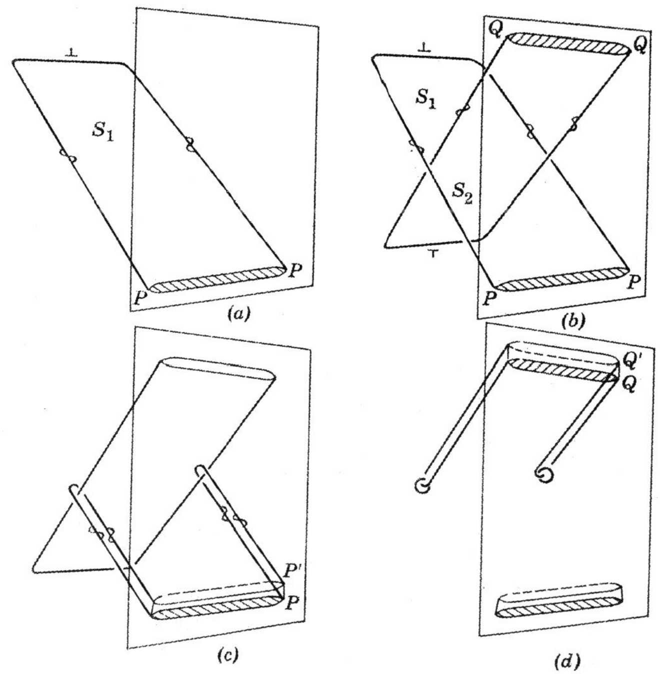

Fig. 9. The Cottrell-Hull extrusion mechanism.

two intersecting slip systems with comparable facility (a condition that is satisfied in the experiment described above), but neither it nor the Mott process readily accounts for any regularity in spacing of extrusions in a slip band. As an alternative, therefore, we consider another possible mechanism in which the most favored glide plane is traversed by a "forest" of screw dislocations (F in Fig. 10a). If these are arranged in groups in which dislocations of one sign predominate, the resulting jogs on the moving edge dislocations will cause the slip step to be itself stepped

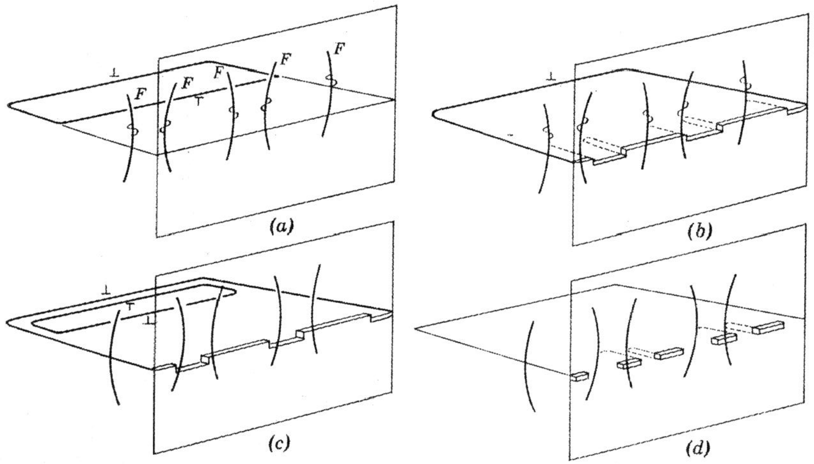

Fig. 10. Suggested extrusion mechanism.

(Fig. 10b). If the screw dislocations F were stationary, events in the second half-cycle would annul the surface step formed in the first. However, if the applied stress has a component in the appropriate direction, the screw dislocations will swing to and fro with the periodicity of this stress (compare Figs. 10a and 10c). The effect of this will be that the negative slip steps will not coincide everywhere with the positive steps, and this in turn will give rise to extrusions (Fig. 10d). These will occur in rows along a slip band, their spacing being determined by the distribution of the screw dislocations that cross the slip plane. This distribution will be independent of the proximity and orientation of the surface, being determined by the previous history of the crystal. The agreement between calculated and observed lines in Figs. 6 and 7 is thus consistent with the proposed model. The length of the individual extrusions, on the other hand, will be determined by the amplitude of the motion of the intersecting screws; it is suggested that this may account for the small size of the extrusions at the edges of the observed region, where the slip-band spacing was itself smaller than elsewhere. The thickness of the

extruded portions will be related to the net number of screw dislocations involved in their production.

There are, however, two objections to the model. First, to an even greater extent than Mott's suggestion, it requires a nonconservative dislocation movement. The intersecting screw dislocations develop jogs, and these jogs can only move with the production of a sufficient number of vacant lattice sites to compensate for the volume of the extruded material. Second, it does not seem possible to modify the mechanism to give rise to intrusions. It may well be that, in fact, different mechanisms of extrusion operate under different conditions. Which of them, if any, is concerned with the initiation of a fatigue crack is another matter. In the experiments described above, both the main crack and the solitary subsidiary crack formed in regions remote from those in which the extrusions were observed. On the other hand, Forsyth [7] states that if the extruded material is carefully polished away it is sometimes possible to reveal an incipient fatigue crack below it. The recent elegant work by Wood,[14] in which the surface topography of a fatigue specimen was revealed by taper sectioning, shows structures which may well be the "extrusions" and "intrusions" discussed above, and shows further some of the ways in which such incipient cracks can develop. Further work on these lines should do much to improve our understanding of the origins of fatigue fracture.

APPENDIX

We have previously discussed [15] the possible configurations of extended dislocations in face-centered cubic lattices; the purpose of this note is to consider the special conditions that prevail at a twin interface in such a crystal.

For convenience of description, we consider the interface, which is a (111) plane, to be horizontal. The crystal above the interface will be called the "matrix" and that below, the "twin": They are, in fact, equivalent. The sequence of close-packed layers in the matrix, going upward from the surface, will be denoted in the usual way by *abcabc* · · · and called Δ stacking; the sequence in the twin, going upward *toward* the interface, will then be · · · *cbacba* and will be called ∇ stacking. The possible Burgers vectors of the matrix can be represented by the edges of a regular tetrahedron, placed with its base parallel to the twin interface and its vertex above its base. The tetrahedron should be so oriented that its vertices correspond to the centers of four contiguous atoms, three lying in (say) an a layer and the fourth in a b layer. The vertices will be denoted by D, A, B, and C, where ABC lies parallel to the twin interface.

Possible Burgers vectors of the twin will then be represented by a similar tetrahedron $D'ABC$, standing on the same base ABC, but on the under side of it, that is, D' is below ABC. The Burgers vectors of the familiar type of partial dislocation in either matrix or twin are represented by the lines joining the mid-points of the faces of the tetrahedron to adjoining vertices. These mid-points will be denoted by α, β, γ, and δ in the matrix and α', β', and γ' in the twin, α being opposite the vertex A, and so forth (Fig. 11).

We can introduce a partial dislocation into the matrix (say), so that the associated stacking fault lies parallel to ABC by making a cut part way across the crystal along a plane parallel to ABC and giving the two cut faces a relative displacement $C\delta$ (or $B\delta$ or $A\delta$) before sticking them together again. The stacking sequence across the faulted region is now $\cdots \Delta\Delta\nabla\Delta\Delta \cdots$.*

The Burgers vector of the resulting partial dislocation will be either $C\delta$ (or $A\delta$ or $B\delta$) or else δC (or δA or δB), depending on the manner of the displacement. The rules for determining the sign of the Burgers vector are taken to be those set out by Frank.[16] It will then be found that a partial dislocation of the type $C\delta$ always has the faulted region on the right-hand side of the line, while one of the type δC has the fault on the left. The same result can be expressed more generally and without reference to an observer as follows: If the positive direction of the dislocation line is denoted by $\overline{Oy}$, then the vector $\overline{Oy} \times \overline{\delta D}$ lies in that half of the glide plane containing the fault in the stacking sequence if the Burgers vector is of type $C\delta$ and in the unfaulted half if the vector is δC.

We can similarly introduce a partial dislocation and a stacking fault into the twin crystal. The faulted sequence will here be $\cdots \nabla\nabla\Delta\nabla\nabla \cdots$. The rule given at the end of the previous paragraph will be valid for twin as for matrix if, for the former, we make use of the reference tetrahedron $ABCD'$ instead of $ABCD$. However, if we express the rule in the less general phraseology which depends on the viewpoint of an observer and if we consider only stacking faults parallel to the twin interface, the signs are reversed as between twin and matrix. Thus partial dislocations of the type $C\delta$ in the twin have the faulted region on the left, and vice versa.

* This is the same stacking sequence as is obtained by the removal of a half-plane of atoms which lay parallel to ABC and has been called an intrinsic fault.[16] If, instead, we *insert* an additional half-plane, we obtain an extrinsic fault, with the stacking sequence $\cdots \Delta\Delta\nabla\nabla\Delta\Delta \cdots$. The energy of the two resulting structures is, to a first order, the same, since they have the same number of next-nearest-neighbor faulted layers. The observations of Hirsch [17] on the configurations taken up by networks of extended dislocations in stainless steel are consistent with the view that only one of the two types of stacking fault exists. In order to produce an extrinsic fault by slip processes from a perfect lattice, it is necessary to carry out *two* cut-and-displace operations on neighboring planes, and the glide of such a partial dislocation involves a correspondingly more complicated motion of the atoms. For these reasons, which admittedly do not amount to a rigorous justification, the following discussion is limited to a consideration of intrinsic faults.

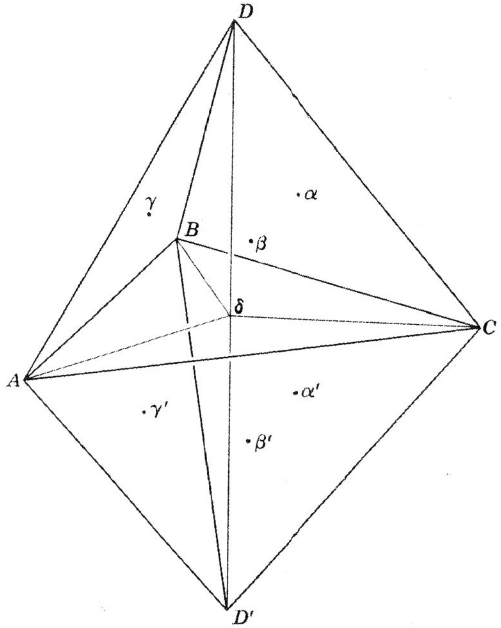

Fig. 11. Reference tetrahedra.

In short, this means that, in the matrix, an ordinary, extended dislocation must have the form

$$\Delta\, X\delta\, \nabla\, \delta Y\, \Delta$$

while in the twin it is

$$\nabla\, \delta X\, \Delta\, Y\delta\, \nabla$$

where X and Y stand for any two of A, B, or C, and the symbols Δ and ∇ denote the stacking sequence either between, or outside, the two partials.

We are now in a position to consider dislocations near twin boundaries. We consider, in turn, (1) those that lie parallel to the boundary and (2) those that intersect it.

1. We may note first that the plane of symmetry of the twin interface

passes *through* the centers of atoms, while the plane on which a dislocation line lies or on which a stacking fault exists passes *between* atoms. Thus a dislocation cannot strictly lie in the boundary; it must lie in one crystal or the other. If it lies anywhere except in the plane immediately adjacent to the boundary, it is, to a first order, not different from a dislocation remote from the boundary. We see from the previous paragraph, however, that an ordinary, extended dislocation, lying in the plane adjacent to and above the twin boundary, has the effect of raising the boundary by one layer in the region between the two partials as a consequence of the change of stacking sequence. Similarly, a dislocation lying in the twin and adjacent to the boundary causes the boundary to step down and then up again as it passes the two partials. Thus the interfacial energy is the same between the two partials as it is outside them, both being equal to the twin-boundary energy. The tendency of the partials to repel one another is thus not opposed by any increase in the stacking-fault energy as in the more familiar case, and the usual "ribbon" of stacking fault will broaden out into a considerable area.

2. If the dislocation line crosses the twin interface we can usefully distinguish three different cases:

(*a*) The Burgers vector in both matrix and twin is AB (or BC or CA). In both matrix and twin, the dislocation will split into two partials separated by a ribbon of stacking fault. The partials cannot be the same on both twin and matrix, so they must either join to form a fourfold node of partials in the interface, or else some kind of "stair-rod" dislocation [15] must exist in the interface.

(*b*) The Burgers vector is of the type DA in the matrix and BD' (or CD') in the twin. If, starting from a perfect twinned crystal, we introduce the dislocation DA into the matrix by the usual cut-and-shift process, the resulting dislocation in the twin has a Burgers vector that is not a lattice vector. A second operation is necessary to remedy this, and, if the situation in the matrix is not to be disturbed, this necessitates the introduction of a third dislocation associated with the twin interface whose Burgers vector is δC, as seen by an observer at the resulting node in the interface. From this viewpoint, the dislocation in the twin has a vector $D'B$, and we have, for the lines meeting at the node,

$$\begin{aligned} & DA + D'B + \delta C \\ & \quad = (D\delta + \delta A) + (D'\delta + \delta B) + \delta C \\ & \quad \equiv 0 \end{aligned}$$

since $D'\delta \equiv \delta D$ and $\delta A + \delta B + \delta C \equiv 0$

It is clear that, so far as its effect on the stacking sequence is concerned, the *twinning* dislocation δC is similar to an ordinary partial; we have, in fact, the arrangement

$$\nabla\ \delta C\ \Delta$$

This too causes a step in the twin boundary; it is the presence of this step that preserves the twin interface as a single sheet when the atomic planes parallel to the original twin interface are transformed into a helical surface by the introduction of the screw dislocation $BD' - DA$. If the sign of the screw dislocation is reversed, the corresponding twinning dislocation is

$$\Delta\ C\delta\ \nabla$$

(*c*) The Burgers vector is DA in the matrix and AD' in the twin. As in case (*b*), this requires the existence of a twinning dislocation, and the same kind of argument shows that its Burgers vector is $2A\delta$. The vector sum at the node is thus $DA + D'A + 2A\delta \equiv 0$. It appears to be possible for such a twinning dislocation to split up into three partials of the common type in one of two ways:

$$\nabla\ \delta B\ \Delta\ A\delta\ \nabla\ \delta C\ \Delta$$

or

$$\nabla\ \delta C\ \Delta\ A\delta\ \nabla\ \delta B\ \Delta$$

The net change in the stacking sequence, which must be associated with any twinning dislocation, is preserved, and the total energy per unit length of the dislocation line is reduced, if one takes the simple view that it is proportional to the square of the Burgers vector. As in section (1), the three partials will repel one another, unrestrained by considerations of changes in stacking-fault energy.

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DISCUSSION

J. J. Gilman, *General Electric Research Laboratory.* Some recent results of Coffin on constant strain-amplitude fatigue tests seem to bear on the question of the mechanism of fatigue. Coffin and Tavernelli[1] find that a simple relationship exists between the cyclic strain amplitude that is imposed on a material and the number of cycles that are required to fracture it. Their most extended results have been obtained for 2S aluminum and Type 347 stainless steel (Figs. D.1 and D.2), but similar results have been obtained for a wide variety of metals. It may be seen from Fig. D.1 that the data follow a particularly simple relationship: The number of cycles needed to cause fracture (N_f) is inversely proportional to the square of plastic strain range, $\Delta\epsilon_p$.

$$N_f \sim \frac{1}{\Delta\epsilon_p{}^2}$$

The relationship holds for as many as 5×10^5 cycles of strain.

It is of particular significance to the mechanism of the process that the $\frac{1}{4}$-cycle point is simply the fracture strain for a uniaxial tension test. It may be seen that this value falls neatly on the same curve as the push-pull fatigue data. This fact seems to rule out immediately the central role of intrusions in fatigue processes because they would not exist for a simple tension test. Even for tests of several cycles, it is difficult to imagine how sharp intrusions would develop; yet the results of Fig. D.1 indicate that one mechanism operates over a wide range of numbers of cycles. Thus it would seem that intrusions are a manifestation, but not a cause, of fatigue.

The above relationship suggests an interpretation in terms of defects produced by dislocation intersections. These will either be strings of vacancies having various lengths or dislocation dipoles of various lengths. One postulates that fatigue cracks form and grow at places where the defect concentration reaches a critical value. The number of intersections that a given dislocation will make as it moves across a glide plane will depend on the dislocation density ρ and on how much area A the dislocation sweeps out. Therefore, for all the dislocations, the number of intersections will be $\sim(\rho A)^2$. Now, in terms of dislocation motions, the plastic strain is $\epsilon_p = b\rho A$, so the number of intersections per cycle

will be $I \sim (\Delta\epsilon_p)^2$. When the number of cycles reaches N_f, a critical total number of intersections will have occurred, and fracture ensues.

Direct evidence of the production of defects during cyclic straining

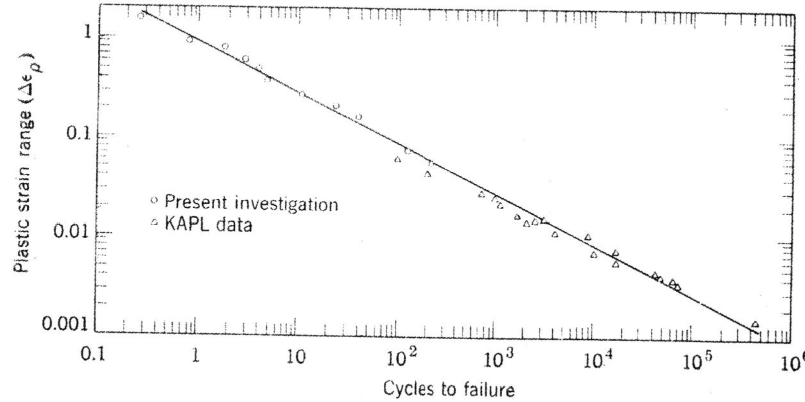

Fig. D.1. Plastic strain range versus cycles to failure (2S aluminum, annealed).

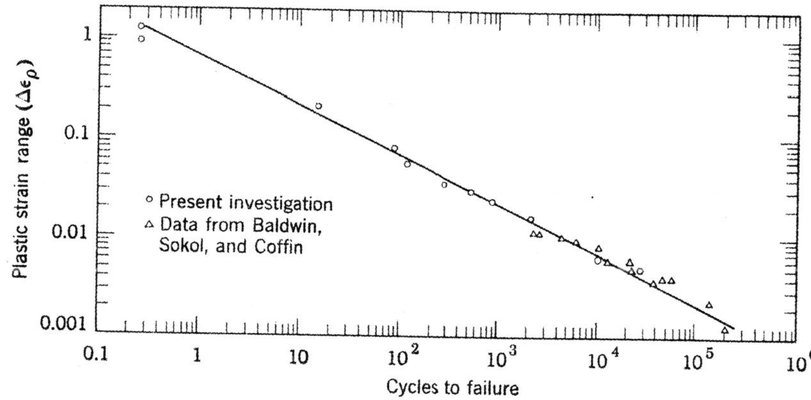

Fig. D.2. Plastic strain range versus cycles to failure (347 stainless steel, solution treated).

has been found by Keith and Gilman.[2] Therefore, the defects do play some role in fatigue, and it seems quite reasonable that their role should be the central one.

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