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DISLOCATION MULTIPLICATION

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I. Introduction

A complete understanding of yielding, strain hardening and the propagation of cleavage cracks will require a knowledge of how the length of moving dislocation line and the length of immobile dislocation line change with strain. For example, plastic strain rate can be expressed as:

$$\dot{\gamma} = b \ell v$$

where

$\dot{\gamma}$ is the average resolved shear strain rate

b is the Burgers vector of the moving dislocations

ℓ is the average length of moving dislocation per cubic centimeter

v is the average dislocation velocity

Gilman and Johnston[1][2] by making various assumptions about how ℓ and v vary with stress and strain have constructed theoretical stress-strain curves that show many of the features of experimental curves. For example, if the initial value of ℓ is assumed to be small, the theoretical curves show upper and lower yield points. This, of course, is an observed behavior for many materials in which nearly all the dislocations can be immobilized by impurity locking. The major difficulty with this approach and any other that considers only average values of stress and strain is to understand the true meaning of the parameters that must be used. In this case, both ℓ and v certainly depend in a very

complicated way on stress, strain, temperature and even thermal mechanical history. In the process of averaging, the really essential differences between various materials may be lost. More insight into fundamental reasons for differences in properties from one class of materials to another may possibly be gained by focusing attention on nucleation and growth of individual slip bands.

Plastic deformation is characteristically an inhomogeneous process. Growth of slip bands, within which the local shear strain may be as much as 10^5 times greater than the average strain, is the dominant mode of plastic deformation for nearly all materials and at small and large plastic strains. See, for example, the recent electron microscope observations of Mader.[3] The slope of an average stress vs average strain curve often tells nothing about the true strain hardening rate. A more fundamental parameter is the amount of damage left in the crystal by each increment of strain. The extent to which this damage shows up on the stress strain curve as an immediate increase in the stress required to continue plastic deformation must depend not only on the amount of damage but also on how it is distributed.[4] Damage in the form of immobile dislocation lines and probably small clusters of point defects is generated during plastic deformation at a much greater rate in a magnesium oxide crystal than it is in copper. Yet, the magnesium oxide stress strain curve shows essentially no hardening after two percent plastic extension. In this material, the local hardening is so great <u>within a slip band</u> that, on the average, only one dislocation is able to pass over every tenth atom layer.[5][6] No hardening is evident on the stress strain curve because plastic strain increases by continuous widening of the slip bands into previously undeformed regions of the crystal. In copper, the local strain in a slip band may be 1000 times greater yet only scattered

regions of the sheared volume within the band contain any visible damage[7][8] This is why, when deformed metals are examined by transmission electron microscopy, it is usually not possible to even identify the locations of slip bands. Fig. 1a shows a slip band in magnesium oxide and Fig. 1b shows a typical field of view for a lightly deformed copper crystal. Both are transmission electron micrographs at approximately 50000X magnification.

What is the reason for this fundamental difference in behavior? The yield stress as ordinarily measured in a tension test is not the stress at which dislocations begin to move. Etch pit observations[8][9] and highly sensitive stress strain measurements[10] show that dislocations start to move in magnesium oxide at a stress less than half the yield and in copper at 1/20th the yield. The yield is characteristically the stress at which slip bands begin to grow. A satisfactory explanation of yielding and an understanding of the very different kinds of damage left in different materials by plastic deformation must involve the details of the mechanism by which new lengths of dislocation are formed and get stuck during the growth of a slip band.

It is the purpose of this paper to discuss the mechanisms by which dislocations can multiply and to try to explain the different dislocation substructures in deformed metals compared to those in deformed ionic crystals.

II. Definition of Multiplication

In a growing slip band, it is necessary to differentiate between multiplication of shear strain, multiplication of the length of moving dislocation line, and increase in the length of immobile dislocation line.

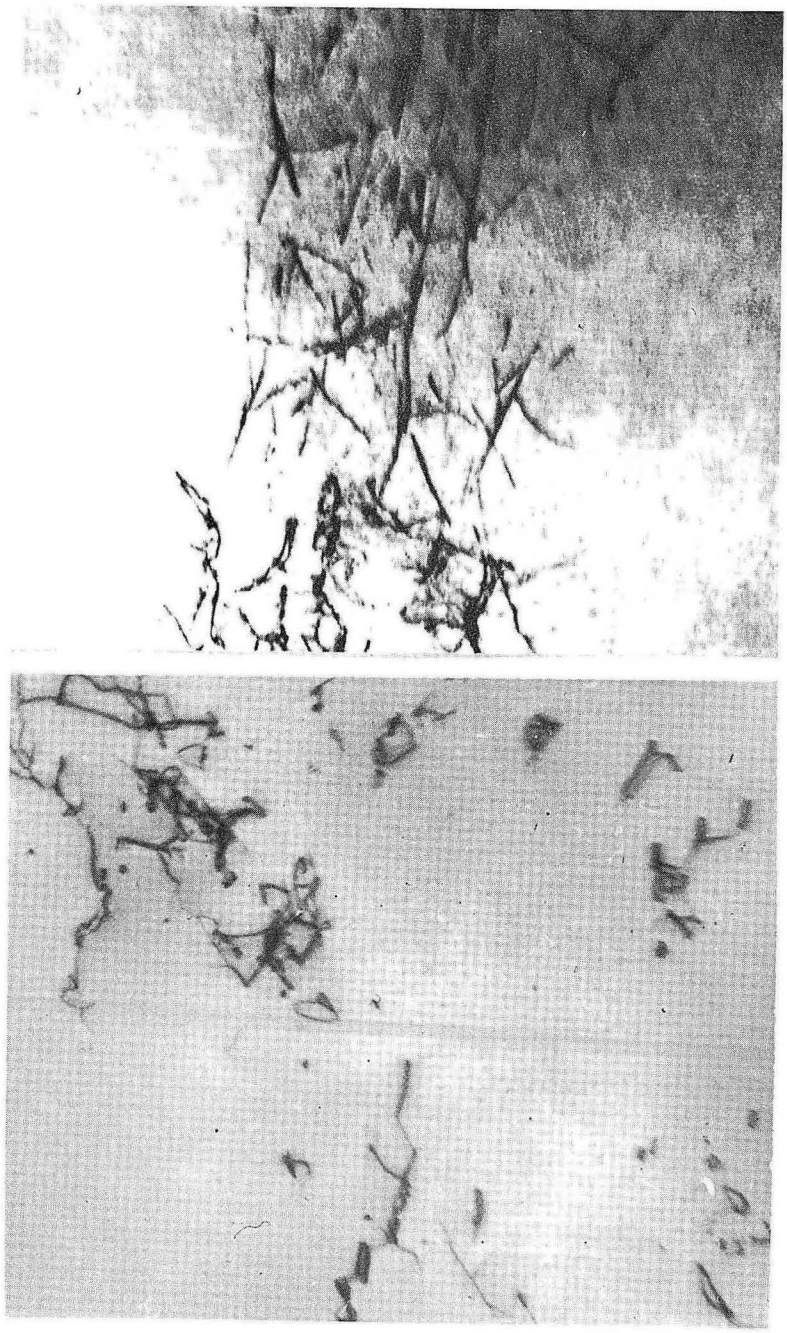

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Fig. 1. Transmission electron micrographs showing damage caused by slip -- 50 000 X.
(a) Magnesium Oxide
(b) Copper

All three of these processes usually occur together. They have frequently been confused in the literature. For example, when one counts dislocation etch pits in a slip band in a lithium fluoride or magnesium oxide crystal, the number which is obtained has nothing to do with the number of moving dislocations. In a wide band, nearly all the pits except a few along the edges of the band mark immobile dislocations in the form of elongated prismatic loops.[5][6] The density of pits also has nothing to do with the shear strain within the band. The shear strain is determined by the number of dislocations that have traversed the glide planes and left the crystal and is not always related to the amount of damage in the form of immobile dislocations that is left behind. For the purpose of this discussion, strain multiplication can be defined as the shearing by more than one Burgers vector of any line that passes through the slip band at right angles to the glide plane.

III. Frank Read Source

Since 1950 it has been generally thought that the problem of strain multiplication was satisfactorily solved by the Frank Read source. This classical mechanism of multiplication is pictured in Fig. 2. The essential idea is that under certain conditions a dislocation that is pinned at one or more points, AB, on its glide plane can rotate about its pinning points and, therefore, traverse each part of the glide plane an unlimited number of times. In a broad sense, all possible mechanisms of strain multiplication, except direct nucleation of a new dislocation loop by the combined action of stress and a thermal fluctuation, are only modifications of the Frank Read mechanism. However, it is worth discussing, in detail, some of the general configurations of dislocations that can result in strain multiplication

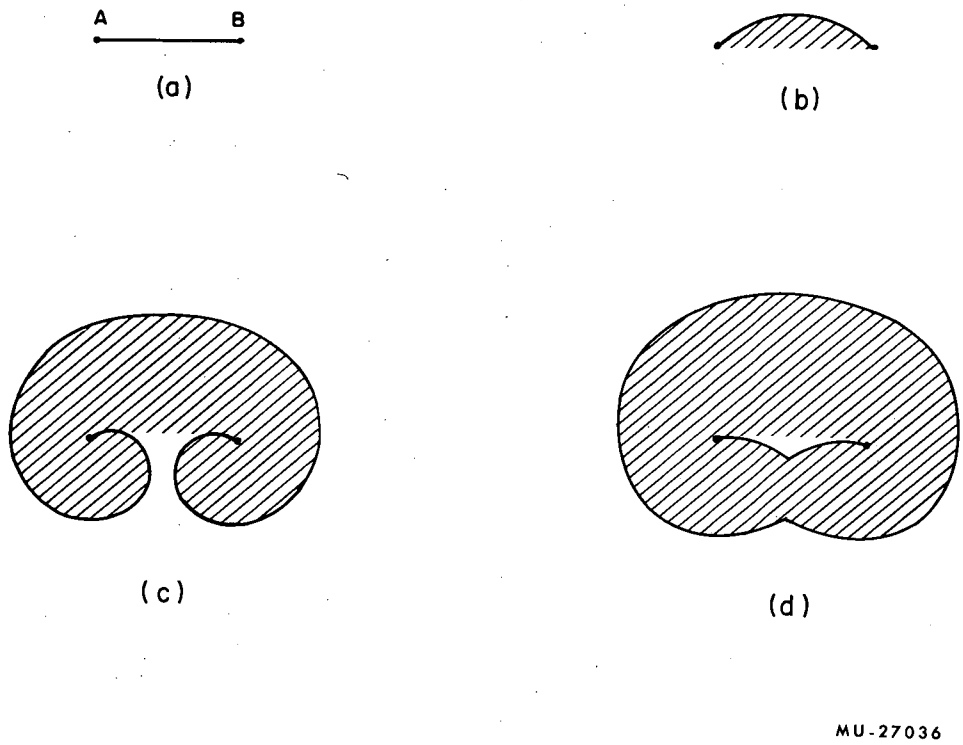

Fig. 2. Frank Read source.

because it has recently been suggested by many experiments that sources like that shown in Fig. 2 do not explain the growth of slip bands of the type seen in ionic crystals. Entire wide bands of slip in lithium fluoride[12] and magnesium oxide can grow from small isolated half loops of dislocation line that have been introduced at the surface of the crystal. Centers of strain multiplication appear to develop continuously wherever there are moving dislocations; in this case, at the interfaces between the deformed material that contains a high density of immobile dislocations within the band and the undeformed crystal to either side.[5] The shear displacement in any one glide plane in these slip bands is probably never more than a few Burgers vectors.

IV. Cross-over Mechanism of Strain Multiplication

A situation that must often occur in real crystals is illustrated in Fig. 3. Consider a moving dislocation that develops an irregular shape because it encounters obstacles to its motion that are not uniformly distributed. Parts of the same line may often meet after following different paths through the crystal, Fig. 3c. Three different things can happen when these segments of opposite sign come together: (1) They can annihilate. (2) An immobile close pair can be formed. (3) The two segments can pass each other. These possibilities are illustrated in Figs. 3d, e, and f. Annihilation will occur if the two segments lie on the same glide plane or are near screw orientation; if not, a pair will be formed when the separation between the two glide planes is less than a critical value, h^*, and cross-over will occur if it is greater than h^*. The critical separation depends on the local shear stress

$$h^* = \frac{Gb}{8\pi(1-\nu)\tau}$$

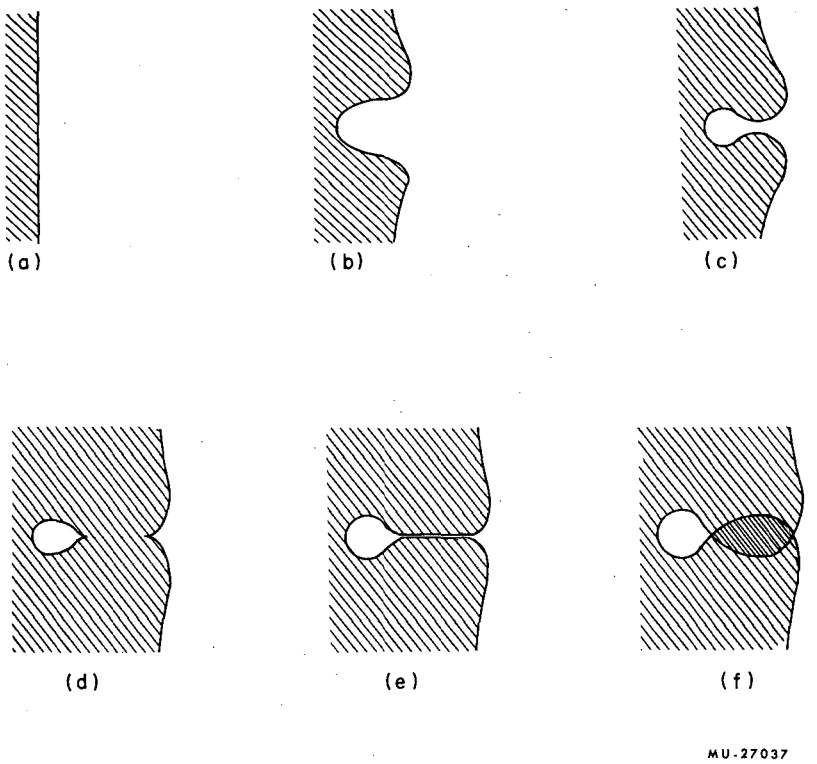

Fig. 3. Motion of a dislocation that meets obstacles that are uniformly distributed.
(a, b, c) Possible successive configurations.
(d) Annihilation
(e) Pair formation
(f) Cross-over

h* is the separation between the two glide planes

b is the Burgers vector

is Poissons ratio

G is the shear modulus

τ is the local resolved shear stress

Pair formation appears to be an important mechanism by which the length of immobile dislocation line in a crystal increases during plastic deformation.[6] Cross-over may be an important mechanism of strain multiplication.[4]

There are two reasons why segments of the same dislocation line may be on separate glide planes on arriving at the same point by two different paths. The first depends on the presence of a network of small angle boundaries. Virtually all the metal and ionic crystals that have been studied up to the present time have contained some subgrain boundaries. Consider the motion of a dislocation through such a crystal.

Fig. 4 shows a loop of dislocation approaching a system of small angle boundaries A, B, C, D and E. Assume that the moving dislocation is able to break through boundaries B, C, D and E but is held up by boundary A. It may then assume the successive shapes: 1 2 3. The separation between the glide planes of the two parts of the dislocation that meet on the other side of boundary A will be equal to the length of segment A times the twist component of its boundary angle. In a magnesium oxide crystal a slip band grows at a shear stress of 10^4 g/mm^2. If the segment A is 10^{-3} cm long, the boundary would have to have a twist component of only .15° to allow the dislocation loops to cross-over. There is a second reason why the glide planes of the two segments may differ when they meet. Consider the expansion of a dislocation loop in

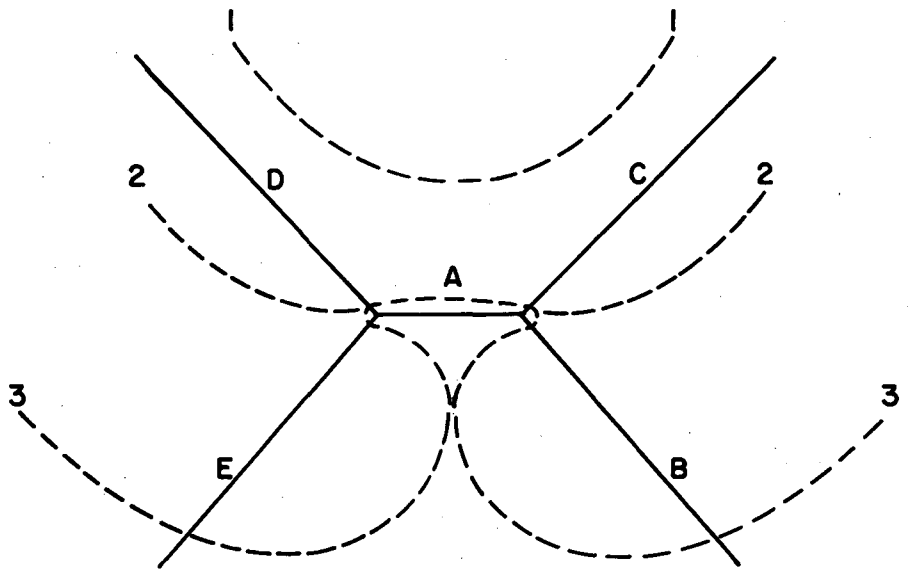

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Fig. 4. Glide of a dislocation through a crystal containing subgrain boundaries -- 1, 2 and 3 represent successive positions of a line that breaks through boundaries B, C, D and E but not through A.

a crystal that contains vacant lattice sites in a form that can be swept up by moving dislocations. For example, a quenched aluminum crystal contains vacancy clusters in the form of small prismatic dislocation loops. It has been shown that these loops are swept away by moving dislocations.[13] When vacancies are absorbed at a dislocation that is not pure screw, the line climbs off its original glide plane. Assume that the loop of dislocation in Fig. 5 initially lies on its glide plane and that its Burgers vector is horizontal. Then it is in edge orientation at the two points marked E and E' and in screw orientation at S. If absorption at E' causes it to climb up off the original glide plane, then absorption at E' causes it to climb down. The parts of the loop that are not pure edge may also move off the original glide plane due to absorption of vacancies by a combination of climb and cross-slip. It is not known under exactly what conditions vacancies or vacancy clusters will actually be absorbed by moving dislocations.

If absorption is possible on dislocations of mixed character, as well as those in pure edge orientation, climb displacement is given by

$$d = \frac{nb^2}{\sin\alpha}$$

d = climb displacement

n = number of vacancies absorbed per unit length of line

α = angle between the line and its Burgers vector

Edge views of the loop before and after absorption of a given number of vacancies per unit length are shown in Fig. 5b. The important point is that any two diametrically opposite segments on the loop always climb in opposite directions. Another way of stating the same thing is that dislocation segments that are moving in opposite directions under the

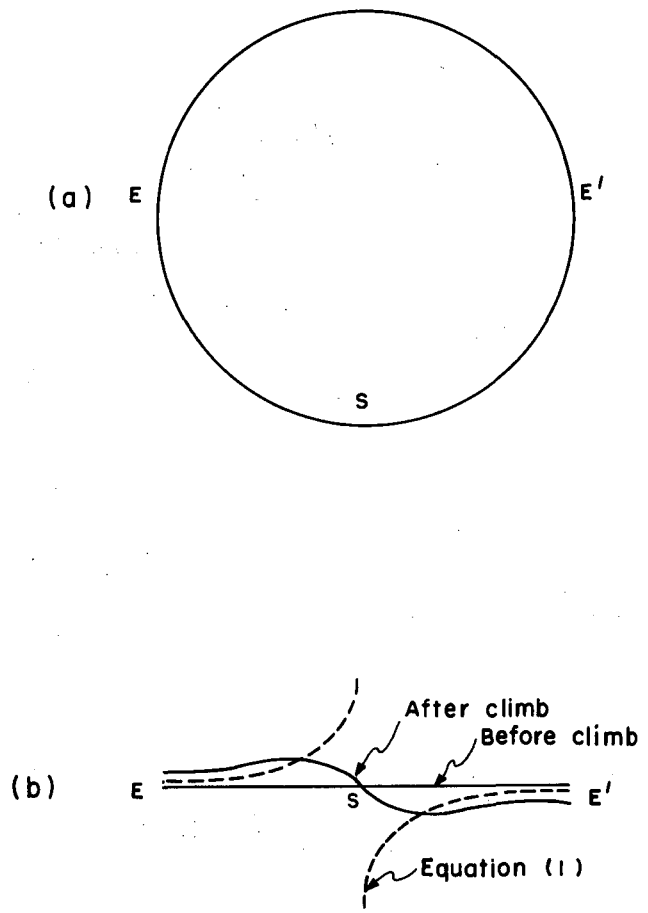

Fig. 5. Climb of a dislocation loop off its glide plane.
(a) The dislocation is in edge orientation at E and E' and screw orientation at S
(b) Side view of the loop before and after absorbing vacancies.

action of the same shear stress climb in opposite directions. Now consider a dislocation that meets an obstacle to its motion and starts to pass around it as in Fig. 3. The two parts of the line that approach each other on the far side of the obstacle are always moving in opposite directions under the action of the same stress. If they are absorbing vacancies, then they are climbing in opposite directions off the original glide plane. For reasonable concentrations of individual vacancies it does not seem likely that this mechanism can often produce wide enough separations for cross-over. For example, if the vacancy concentration is 10^{-5} and all vacancies within one interatomic distance of the glide plane are absorbed during motion, an edge dislocation would climb a distance of about b on moving a distance of 10μ, or if $\alpha = 0.1$, after moving about 1μ. It does seem likely that climb may often start the growth of a long dislocation pair, Fig. 3e. As further motion of the dislocation increases the length of the pair, its spacing may also increase. This is because jogs acquired by the segments of line near the pair tend to glide conservatively toward it.[14] These jogs may be produced by further vacancy absorption or by cutting of intersecting dislocations. Such a pair would have to grow to a length of about 100μ in a crystal containing a concentration of vacancies of 10^{-5} before the spacing due to climb alone would be great enough for cross-over.

V. Annihilation Mechanism of Multiplication

The grown in dislocation substructure may contribute in a different way than those already described to the multiplication process. Consider a moving screw dislocation, P, that encounters elements of the grown in dislocation network A, B, C, D, E, Fig. 6a. If the glide plane of the moving dislocation is within a critical distance of the immobile dislocation

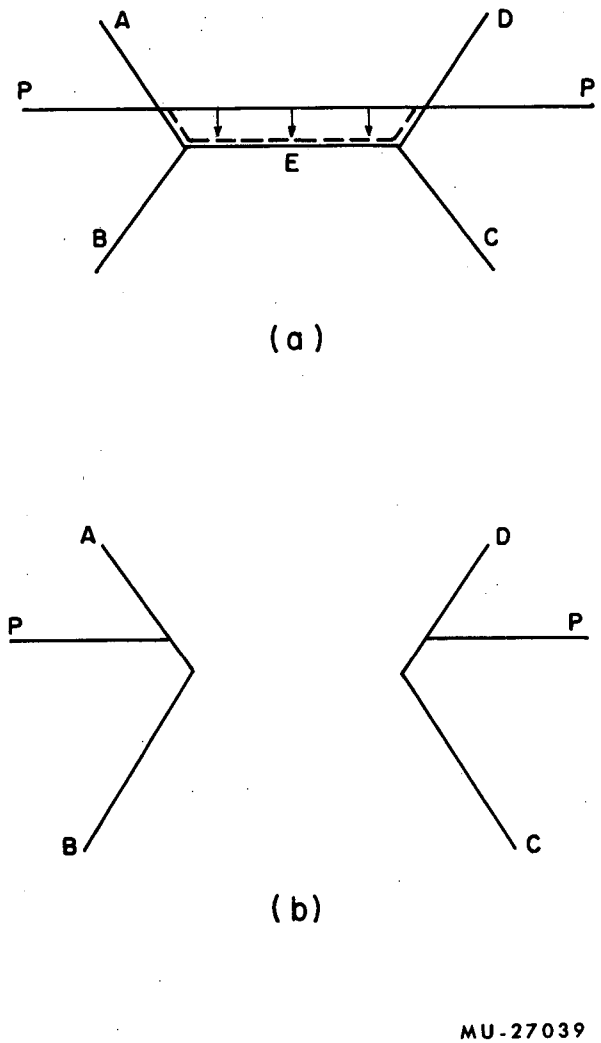

Fig. 6. Cross-slip of a segment of moving dislocation due to interaction with a completely immobile network of grown-in dislocations.

segment, E, and if E and the moving dislocation have opposite Burgers vectors, then a part of the moving dislocation may cross-slip and annihilate the segment E. The final configuration is shown in Fig. 6b. The situation on the glide plane is pictured in Fig. 7a. A segment has been removed from the moving dislocation and two locking points have been established. It will act as a Frank Read source as shown in Figs. 7b, c and d. Therefore, even when the grown-in dislocations are so strongly locked that they are completely immobile, they can play a vital role in the multiplication process. Frank Read sources can be continuously created as soon as there is at least one moving dislocation on a glide plane.

VI. Double Cross-Slip Mechanism of Multiplication

Another way of creating a Frank Read source from a single loop of moving dislocation is cross-slip. A segment of dislocation that lies in pure screw orientation can change its glide plane momentarily to start another loop that expands in a parallel glide plane. The sequence of events is shown in Fig. 8. A screw segment glides into the cross-slip plane and then back into another glide plane parallel to the original one. After one or more loops have been formed in the parallel plane as in Fig. 8b, the small loop that lies in the cross-slip plane may collapse leaving no trace of the origin of the secondary loop. Geometrically this is a very attractive and flexible mechanism for the growth of a slip band. However, it is not easy to see why it should happen. In ionic crystals for which this mechanism has most frequently been proposed,[12] low temperature motion of dislocations is strongly confined to only one glide plane. In the sodium chloride structure a dislocation with Burgers vector

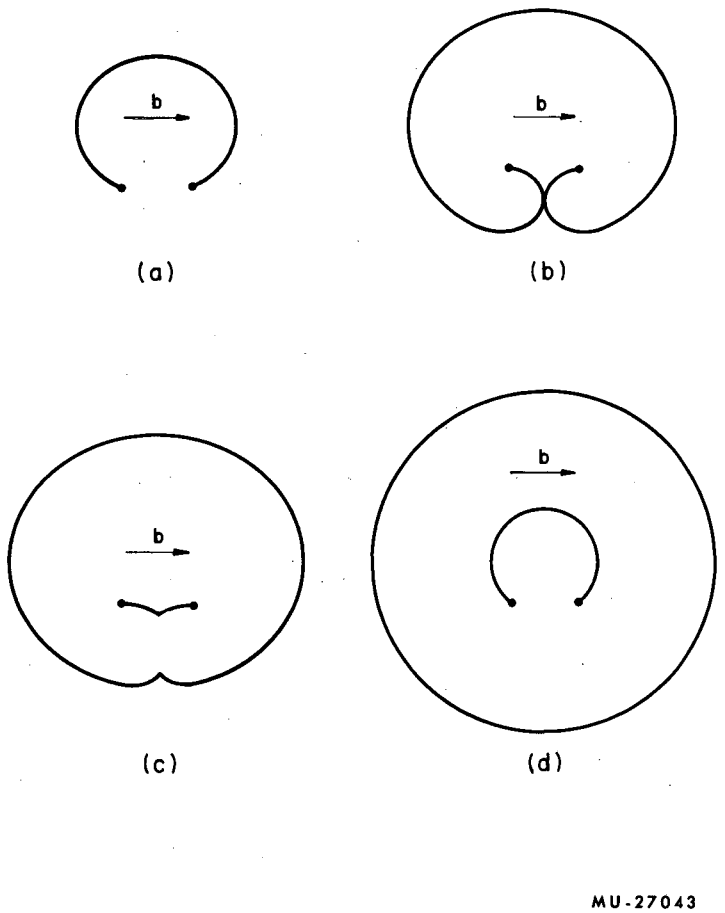

Fig. 7. A Frank Read source can be created by the process shown in Fig. 6 as soon as there is one mobile dislocation on a glide plane.

$\frac{a}{2}$ [110] glides in the $(1\bar{1}0)$ plane at a stress far below that required for glide in (001) which is the only likely cross-slip plane. In magnesium oxide at room temperature, this stress ratio is about 1 to 50.[15] Therefore, double cross slip multiplication would seem to require a rather complex system of local stresses. Small precipitate particles may be able to induce this type of strain multiplication. Direct electron microscope observation of moving screw dislocations is magnesium oxide foils[16] do not show cross-slip of the type sometimes seen in face centered cubic metals where a dislocation glides for easily measurable distances in the cross-slip plane. In magnesium oxide, screw dislocations move off the original {110} glide plane gradually in unresolvably small steps.

In close packed metals there is also a barrier to cross-slip. Even though a cross-slip plane exists in which the dislocation can glide as easily as on the original glide plane, it is different for a dislocation to change glide planes. The reason is the splitting of dislocations into partials with a ribbon of stacking fault between. In order to move into the cross-slip plane it is necessary to first have the partials recombine. Particularly for metals with low stacking fault energy this recombination requires very high local stresses.

VII. Boundary Multiplication

When a crystal contains a continuous network of subgrain boundary tangles like those found in deformed copper, Fig 9, or a well developed network of sharp subgrain boundaries such as those characteristic of heavily deformed aluminum, Fig. 10, it is doubtful that a moving dislocation will find any point along the boundary where it can cut through

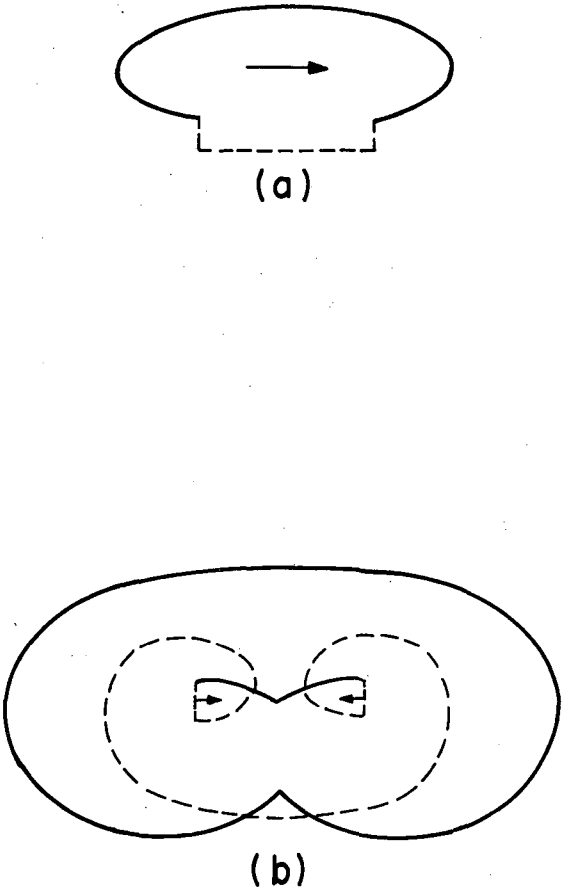

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Fig. 8. Double cross-slip strain multiplication.

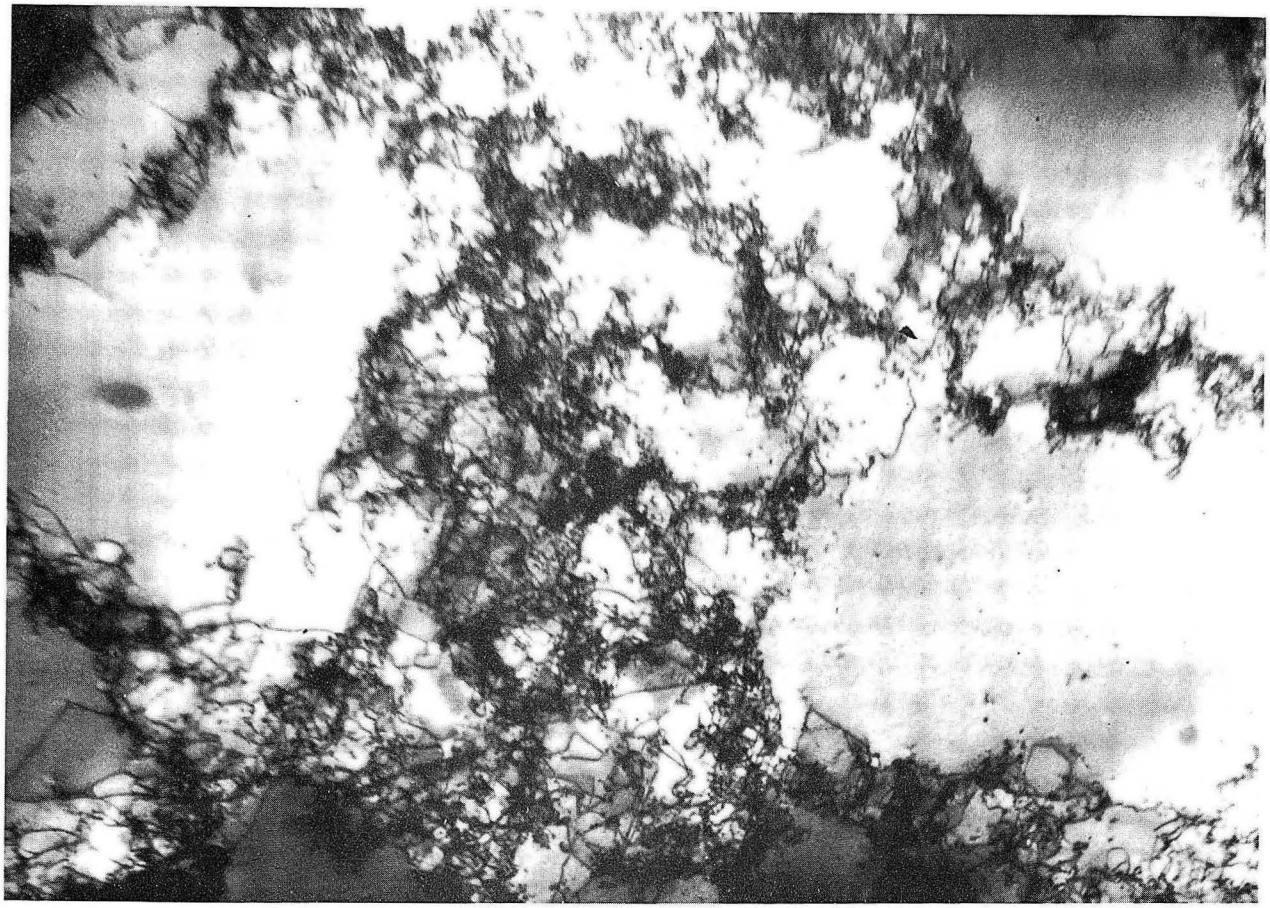

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Fig. 9. Subgrain boundary tangles in deformed copper -- transmission electron micrograph 40 000X.

into the next subgrain. The critical step for growth of a slip band may then be bowing out of a new segment from the opposite side of the boundary, Fig. 11. If a dislocation, A, moves up against one side of a subgrain boundary, the stress is increased that acts on those elements of the subgrain boundary network capable of bowing out into the neighboring subgrain and thereby continuing propagation of the shear. It is quite possible that new loops, (B, C in Fig. 11a) originating in this way, will be formed at more than one place along the boundary. These loops will usually lie near but not exactly on the original glide plane. Since they are formed independently, adjacent loops can either annihilate, form a pair or, if their glide planes are widely enough separated, cross-over as in Fig. 11b. If they cross-over, a shear displacement of b in the first subgrain becomes 2b in the second subgrain. Repetition of the process can produce a slip band. The result of the process is to leave pairs of dislocations of opposite sign that enter the subgrain boundaries from the two opposite sides. These will be incorporated into the tangle through dislocation reactions, cross-slip and climb. Eventually, they may at least partially annihilate each other within the boundary.

The critical stress for this type of multiplication should depend on the lengths of individual dislocation segments in the tangle. There is considerable experimental evidence that subgrain boundaries do play an important role in determining flow stress. For example, the stress to produce slip bands in crystals of zinc that contain small angle boundaries does depend on the nature of the boundary and on boundary angle.[17] Also one of the most reproducible correlations that can be made between critical resolved shear stress for flow and the dislocation substructure in copper as observed by transmission electron microscopy is a correlation

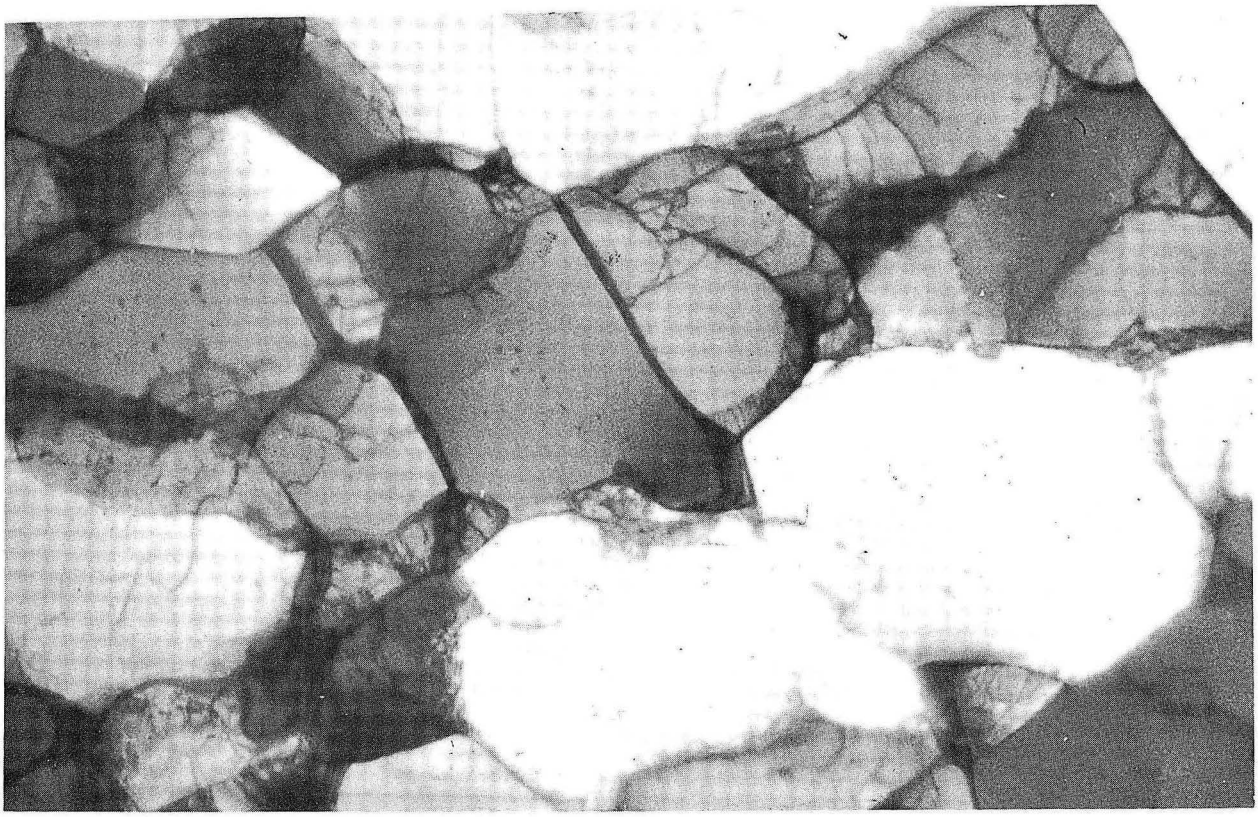

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Fig. 10. Sharp subgrain boundaries in heavily deformed aluminum -- transmission electron micrograph 40 000X.

between stress and the length of individual segments of dislocation in the tangles.[7]

VIII. Damage in Slip Bands

All of the strain multiplication mechanisms involve the meeting of two dislocation segments of opposite sign that either annihilate as in the classical Frank Read mechanism or cross-over each other. In every case the alternative to strain multiplication is the start of a close pair of dislocations. Therefore, the multiplication of strain and a concomitant increase in length of immobile dislocation line are due to the same processes. The maximum shear displacement that can take place on a given glide plane may depend on the number of close pairs formed per unit area swept by a dislocation. Although close pairs and elongated prismatic loops formed from them[6][14] are observed both in metals and ionic materials, slip bands are densely populated with pairs and loops of all sizes, some probably even too small to "see" by transmission electron microscopy. In metals, pairs and loops are relatively rare and are mostly associated with subgrain boundary tangles. In copper and aluminum there are large regions within slip bands, greater than one micron in diameter, that contain no visible damage.

The basic difference between these two types of materials may be the much greater glide mobility of dislocations in the metals not only on the normal glide plane but also on possible cross-slip planes. The ratio of the stress to just move dislocations to that required for slip band growth is much greater for metals than for ionic crystals. A substructure consisting mostly of prismatic loops and pairs is not stable in metals.[13] For example, a close pair or prismatic loop formed in one of the perfect regions in the center of a subgrain may always be eliminated by the next

dislocation that comes along. This process is illustrated by Fig. 12. The shear stress exerts no net force on the close pair, AB. However, if a third dislocation, C, approaches it on a glide plane close to that of A and having the same Burgers vector as A, it will not be able to pass the pair and the group of three dislocation lines may be moved to the right into the subgrain boundary. If AB is a closed loop, then parts of it must glide on a cross-slip plane.

When the damage can be collected into subgrain boundaries in this way, there is a much better chance for annihilation of segments of opposite sign and, therefore, less damage may be left in the crystal per unit area of glide plane swept. When the stress to just move dislocations is high, as in magnesium oxide, the damage remains uniformly distributed in the slip band and the rate of strain hardening (as measured by the small slip band strain) is high.

Body centered cubic metals may be an intermediate case. However, much more experimental work needs to be done concerning the growth of individual slip bands in different materials before the differences in behavior can be fully understood.

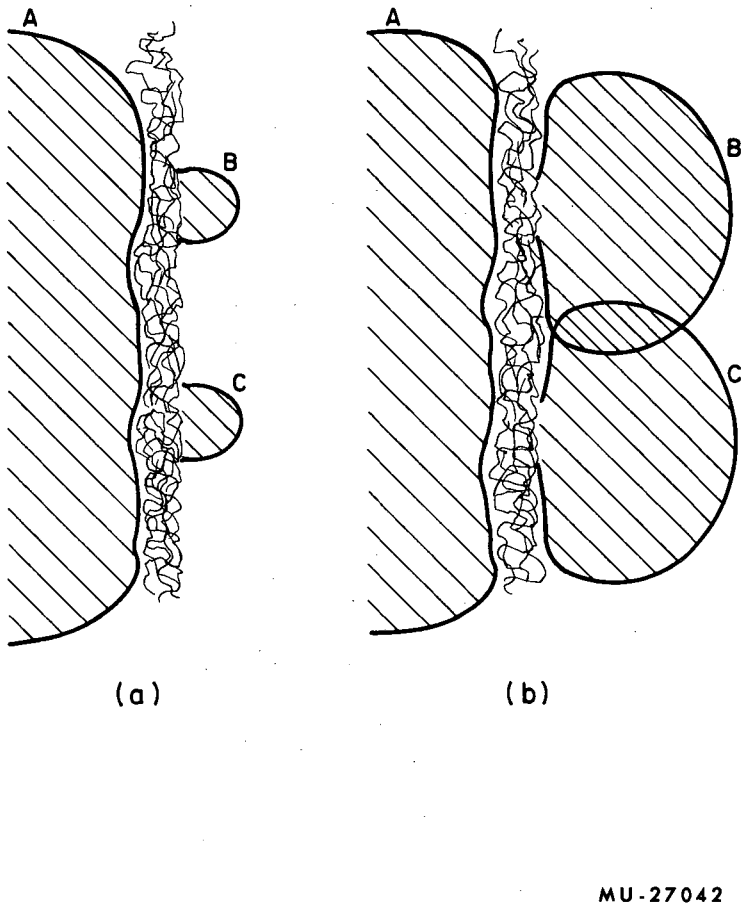

Fig. 11. Strain multiplication at a subgrain boundary.

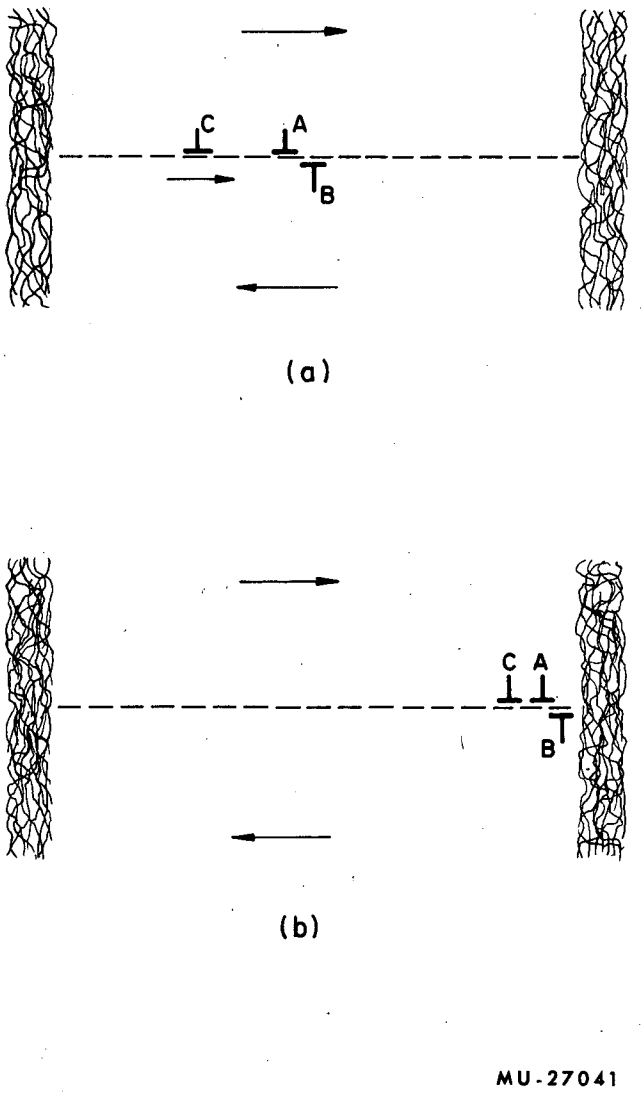

Fig. 12. Sweeping of a dislocation pair or prismatic loop by a moving dislocation.

ACKNOWLEDGMENT

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