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Dislocation Movements in Metals

Defects in atomic structure of crystalline solids determine their plastic properties.

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Defects in crystals are responsible for the differences that exist between the properties of any given real crystal, or crystal aggregate, and those calculated for the corresponding "ideal" crystal; that is, calculated from the mathematical model of an ideally regular array of infinitely many atoms. With respect to a number of properties, such as density, elastic constants, specific heat, and melting point, the differences between real crystals and ideal crystals of the same type are minor. For example, if we calculate the temperature dependence of the specific heat of, say, rocksalt, by considering an ideal array of sodium and chlorine ions in a cubic arrangement, we find a satisfactory agreement between theory and experimental fact. However, besides such "structure-insensitive" properties, there are many "structure-sensitive" properties which reveal enormous differences between ideal and real crystals. In particular, this is true for the whole range of plastic properties of crystals and crystal aggregates. Included are all metals and alloys since these, as is well known from x-ray evidence, are crystalline even though their external shape rarely gives any indication of this fact. Metals owe their usefulness for technical applications largely to their peculiar plastic properties. Among these, we find good resistance against impact loading, and we find ductility and machinability, to name only a few of the most important beneficial plastic properties of metals. Ideal metal crystals, by contrast, should be as brittle as glass, and should continue to be absolutely brittle up to the highest temperatures.

Among the various kinds of known crystal defects—that is, among all known departures of actual atomic arrangements from the ideally regular one—it is the class named "dislocations" which is of overriding importance. As we shall see, the motions of dislocations cause plastic deformation, whereas without mobile dislocations, metals would be, and can be demonstrated to be, brittle. Thus the great discrepancy between the mechanical behavior of real metals compared to ideal metal crystals is due to the motions of dislocations. Therefore, in order to understand the plastic behavior of crystalline aggregates in general, and of metals in particular, the properties and the behavior of dislocations must be studied.

In 1934, Orowan (*1*), Polanyi (*2*), and Taylor (*3*) independently introduced the concept of dislocations into the theory of the mechanical properties of crystals, even though no direct proof could be given for their existence. However, since then, an impressive number of methods for the direct observation and study of dislocations has been developed (*4*). These include the formation of etch pits at the points where dislocations intersect surfaces, x-ray techniques, so-called "decoration" techniques in which precipitates are formed along dislocation lines, the observation of stationary and moving dislocations in electron transparent foils viewed in the electron microscope, and finally moiré techniques and field ion microscopy that allow observation of dislocations, and even the atomic arrangement near their axes. With these methods, a comprehensive body of experimental data has been accumulated, which furnishes detailed evidence supporting practically all statements made in this article.

Nature of Dislocations

Dislocations in real crystals are line defects that can be imagined to be the result of cutting and rejoining operations occurring in three steps as follows. (i) A crystal may be sliced part way through, so that a cut is produced that ends in the crystal, or is completely confined within it, no restriction applying to the shape of the cut. Correspondingly, the edge of the cut may be straight, curved in a plane, or curved in space. The edge of the cut will become the axis of the dislocation (Fig. 1*a*). (ii) The two sides of the cut are rigidly shifted relative to each other (Fig. 1*b*). In reality, near the edge of the cut, the displacement must necessarily drop sharply from its full value down to zero. As a consequence, a strongly disordered region arises in the immediate vicinity of the edge of the cut, which is named the dislocation "core." The relative displacement vector, commonly designated with the symbol *b*, is named the "Burgers vector" in honor of J. M. Burgers, who did much of the early fundamental work on the mathematical theory of dislocations (*5*). (iii) The two sides of the cut must then be rejoined. With this last step (Fig. 1*c*), the dislocation is completed. However, in order for this step to be feasible, pieces of the

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crystal will have to be cut away in places, or material will have to be filled in, except in the special case that the surface of the cut is parallel to the displacement everywhere.

From the description of the three steps in making a dislocation, it follows that a dislocation cannot start or end inside a crystal, and that the Burgers vector over the whole length of a dislocation must be constant. However, dislocations can be joined, can branch, and can form networks, inasmuch as consecutive cuts can coincide over parts of their length. It is intuitively obvious, and can be demonstrated by means of elasticity theory, that dislocations generated as described are the centers of strong internal stress fields, the intensity of which is inversely proportional to the distance from the axis.

A dislocation whose direction is perpendicular to its Burgers vector is named "edge dislocation." The name was chosen because dislocations of this type can be formed by inserting an extra slice of material into a plane cut made according to the aforementioned first step. The edge of the inserted slice (which follows the straight or arbitrarily curved edge of the cut) will then coincide with the dislocation axis (Fig. 2*a*).

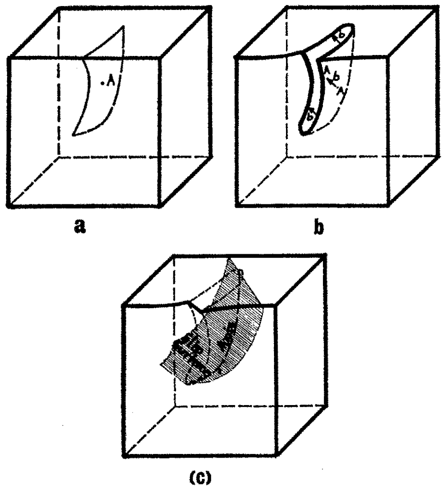


Fig. 1. A dislocation made by a cutting, shifting, and rejoining operation in three steps. *a*, A cut is made partway through a crystal. *b*, The two sides of the cut are rigidly displaced relative to each other, so that any two corresponding points on the surfaces of the cut, such as points *A*, are separated by the same displacement vector *b*, named the Burgers vector. *c*, After appropriately filling-in material, the sides of the cut are rejoined. The surface defined by the dislocation axis and the direction of the Burgers vector, indicated by shading, is the slip surface.

A dislocation with its direction parallel to its Burgers vector arises from a relative shifting operation parallel to the straight edge of a cut. An imagined set of planes normal to the edge of the cut—that is, normal to the dislocation axis—with a uniform spacing of one Burgers vector, is thereby transformed into a continuous screw surface, since, at the cut, each plane is joined to its nearest neighbor. This, then, is named a "screw dislocation" (Fig. 2*b*). Mixed dislocations are all intermediate cases between the two extremes. Thus, in curved dislocations, certain segments may have screw character, other segments edge character, and the remainder may be of mixed type.

In crystals, in order for a dislocation to be stable or metastable, it is necessary that the atoms on the two faces of the cut be rejoined in a position of stable or metastable crystallographic order. This requirement is fulfilled if the Burgers vector is either a lattice vector, or leads from a regular lattice point to a possible metastable atomic position, or leads from one metastable position to another one. In the case that the Burgers vector leads to a metastable atomic position, a fault is generated on rejoining the two sides of the cut. The dislocation will border this fault and is referred to as "imperfect," in contrast to perfect dislocations whose Burgers vectors are lattice vectors. Imperfect dislocations occur only in certain types of crystals and, within these, only on certain planes. The resulting faults are known as "stacking faults."

From the preceding explanations and from Fig. 1, it is apparent that the strains around a dislocation are proportional to its Burgers vector. By Hookes law, stresses and strains are proportional to each other, and hence, the energy of a dislocation is proportional to the square of its Burgers vector. The stresses around dislocations are usually such that dislocations repel each other if their Burgers vectors include angles smaller than 90 degrees between them, but attract if the same angle is larger than 90 degrees. As a result, dislocations tend to have the shortest possible Burgers vectors, since dislocations with longer Burgers vectors can usually lose energy by dissociating into two or more dislocations so that the Burgers vectors of the resultant dislocations add up to the Burgers vector of the original dislocation. Dislocations being mobile, the resultant dislocations then move apart. The most common dislocations are, therefore, those whose Burgers vectors are the shortest vectors between equivalent atoms in the crystal.

The same cause leads to a peculiar effect when metastable atomic positions exist in a crystal. In that case, perfect dislocations can form so-called "extended" dislocations (*6*) by dissociating into two or more imperfect dislocations such that their Burgers vectors lead to and from these metastable positions. The vector sum of the Burgers vectors of the resultant imperfect dislocations, named "partials," then equals the Burgers vector of the original perfect dislocation, but, after dissociation, the sum of the squares of the Burgers vectors, and with it the energy, is smaller than the square of the Burgers vector (and the energy) of the undissociated dislocation. The partials then include ribbons of stacking fault between them, the equilibrium width of which is the larger, the lower the stacking fault energy.

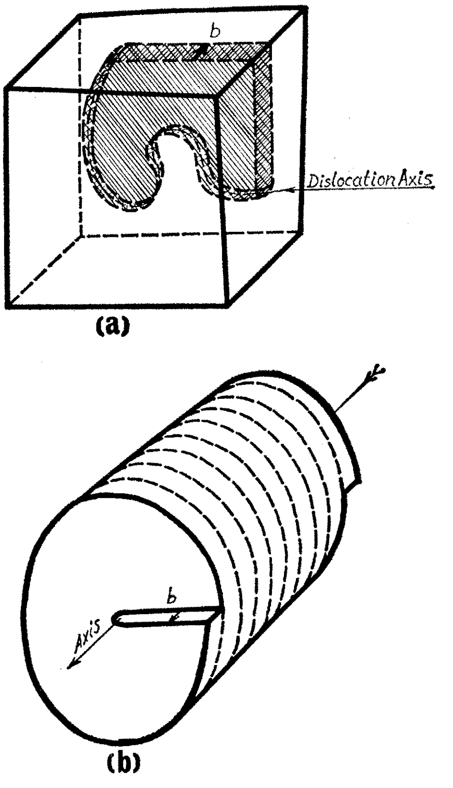


Fig. 2. *a*, Edge dislocation, generated by inserting a plane extra slice of material, bounded by an arbitrary curve, into a specimen. *b*, Screw dislocation, formed by cutting, shifting, and a rejoining operation, whereby the displacement vector *b* is parallel to the edge of the cut. A set of imagined planes normal to the axis is thereby transformed into a screw surface as indicated.

Conservative and Nonconservative Motion

From a geometrical point of view, dislocations in crystals are always mobile. They constitute a disturbance in the crystal which can move as a wave moves over water, or as an elastic shock wave travels through a material. While a dislocation may move large distances, the atoms thereby suffer only small displacements. From the nature of the cutting and rejoining operation that would generate dislocations, it is apparent that the movement of a dislocation can be interpreted as the continuous motion of the edge of the cut. The cutting and rejoining can be effected without bodily transfer of crystal substance to or from the sides of the cut only if the Burgers vector is parallel to the cut.

Thus, that surface which is defined by the direction of the Burgers vector and the axis of the dislocation is the only surface on which the dislocation can move "conservatively" (without the simultaneous bodily transfer of matter, except for the relative displacement through one Burgers vector according to the second step mentioned). Therefore, dislocation motion on this particular surface, named the "slip surface" (see also Fig. 1*c*), is the easiest kind of motion, and causes slip by one Burgers vector over the area swept out by the dislocation axis. For this reason, the direction of the Burgers vector is often called the "slip direction." Hence a dislocation axis can always be regarded as the boundary between two parts of the slip surface, over which the total translation that has taken place differs, in magnitude and direction, by its Burgers vector.

In actual crystals, common slip surfaces are the most closely packed crystal planes. The word "slip system" refers to the combination of Burgers vector and crystallographic slip plane. Evidently, dislocations belonging to a particular slip system can move between each and every close-neighbor pair of the actual atomic planes which are the crystallographic slip planes in that slip system. Crystals usually contain several equivalent slip systems, the multiplicity of slip systems increasing with crystal symmetry.

For example, in face-centered cubic crystals, there are altogether twelve {111}, < 110 > slip systems. On their slip systems, crystals can slip freely, as soon and as long as sufficient numbers of mobile dislocations are available.

The deformation, which takes place by the operation of one slip system, is the same that would result if a stack of corrugated iron sheets were pushed in the direction of the corrugations, or if a pack of playing cards were sheared in a particular direction. This mode of deformation, called "shear," "translation," or "slip," is by far the most important mechanism of plastic deformation in metals. It is illustrated in Fig. 3. The shear deformation, γ, defined as tangential offset divided by normal distance between the slip planes,

$$\gamma = a/h$$

in Fig. 3, is given by the important equation

$$\gamma = \rho\bar{l}b.$$

Here b is the Burgers vector, ρ is the total length of dislocation line per unit volume (measured in cm/cm^3 or in cm^{-2}), and $\bar{l}$ is the average distance moved by the dislocations. For example, in Fig. 3 let the shear $\gamma = a/h$ have been caused by N dislocations parallel to the z-axis, each one having a Burgers vector parallel to the x-axis, and each one moving the distance $\bar{l} = L/n$, where L is the length of the slip plane in the direction of the motion of the dislocations. Then the tangential offset caused by each of the dislocations was b/n, the total offset $a = Nb/n$, yielding the shear

$$\gamma = Nb/nh$$

or

$$\gamma = (N/Lh)\ (L/n)b = \rho\bar{l}b.$$

If in addition to slip, a mechanism operates by which the bodily transfer of matter may take place, then the dislocations can leave their slip surfaces and move "nonconservatively" in any arbitrary direction. Actually, such mechanisms do exist. They are the creation or annihilation of "point defects," that is, of vacant lattice sites as well as interstitial atoms. The motion of dislocations normal to their slip surface is termed "climb."

Forces Inducing Dislocation Motion

1) *Glide forces*: As we saw, dislocation motion causes plastic deformation. As a logical consequence, forces must act on dislocations, causing them to glide when loads are applied to crystals. What then are these glide forces?

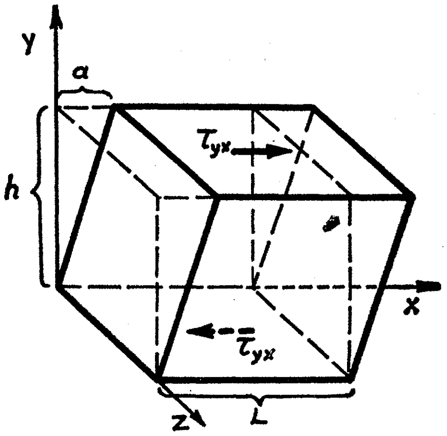


Fig. 3. Auxiliary drawing to clarify the geometry of shear due to dislocation motion. Illustrated is the shear (or slip or translation) which would be caused by the movement of dislocations, with Burgers vectors parallel to the x-axis, on planes normal to the y-axis.

If an infinitesimally small volume element, dv, subject to stresses τ_{ij} (i and j being x, y, or z), is deformed by plastic strains, $d\gamma_{ij}$, then the work done (W) on the volume element by any one of the stress components is

$$dW_{ij} = \tau_{ij}\, d\gamma_{ij}\, dv.$$

Thus, for the case of Fig. 3, where of all strain components only $d\gamma_{yx}$ is finite, only the stress component τ_{yx} does any work on the crystal. Therefore, the glide force on the dislocations must be due solely to this one shear stress component, namely the one which is acting on the slip plane in slip direction, named the "resolved shear stress." If a uniform shear stress throughout the specimen is assumed, a shear increment

$$d\gamma_{yx} = \rho b\, d\bar{l}$$

(Fig. 3), causes the work done on the crystal to be

$$dW_{yx} = V\tau_{yx}\, d\gamma_{yx} = V\tau_{yx}\, \rho b d\bar{l}$$

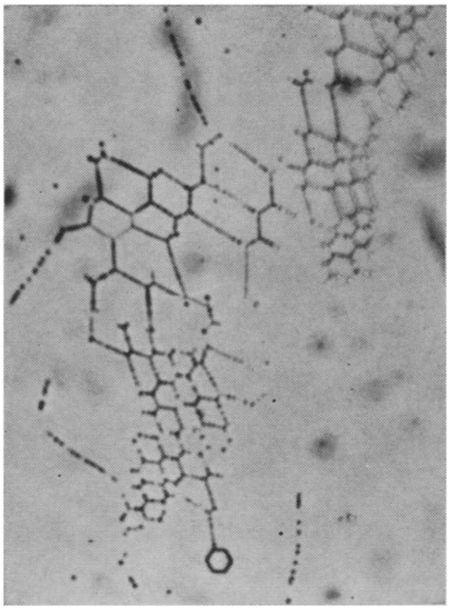
with V the volume of the specimen. This must equal the work done in moving the dislocations (*7*). Hence

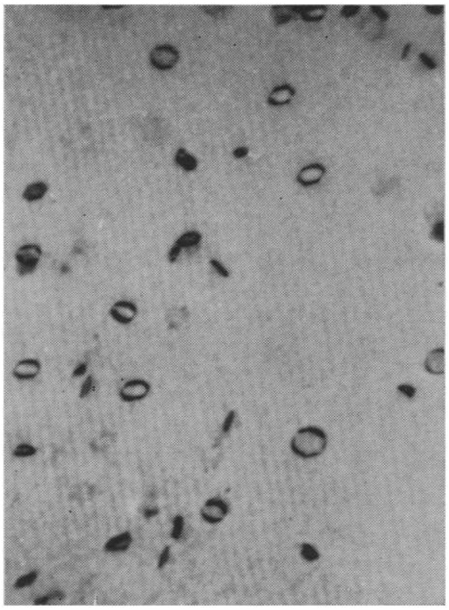
$$dW_{yx} = F_G\, \rho\, V\, d\bar{l},$$

where F_G is the glide force per unit length of dislocation line, acting to drive the dislocations along their slip planes. Thus, one obtains

$$F_G = b\tau_{yx}$$

for the glide force per unit length of dislocation line. This is a quite general result. Specifically if τ_{yx} is a function of position, the value of the resolved shear stress at the dislocation axis must





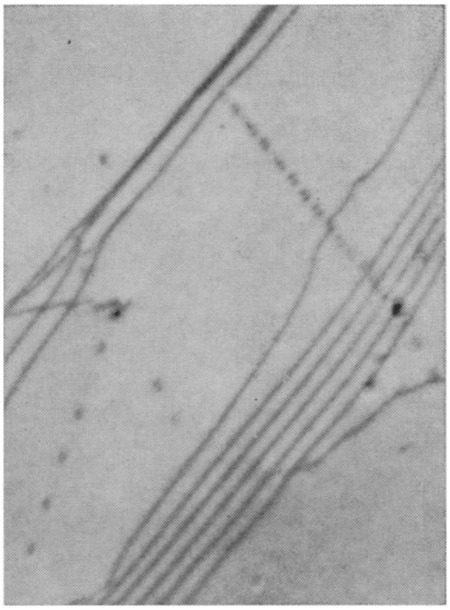


Fig. 4 (left). Networks of "decorated" dislocations in the basal plane of a freshly grown zinc crystal. The dislocations have been made visible by etching. Only parts of the networks are visible because of a small angular misorientation between the specimen surface and the plane of the networks (× 1000). [Courtesy of V. Damiano] Fig. 5 (center). Dislocation loops due to the condensation of vacant lattice sites in the form of close-packed single layers, in aluminum quenched from close to its melting point. The loops outline the boundaries of the condensed layers. Electron micrograph of a foil in transmission about 3000 Å thick (*36*) (× 40,000). Fig. 6 (right). Sequences of very small prismatic dislocation loops radiating out from precipitates in annealed aluminum. The loops are "punched" into the material due to differences in thermal expansion coefficients of matrix material and precipitates. The long smooth lines are glide dislocations, probably due to accidental deformation of the foil. Electron transmission micrograph (× 35,000).

be taken, that is, the force on any given dislocation due to stresses usually changes from point to point.

By considering dislocations as the boundaries of slipped regions on slip planes, it becomes obvious that glide forces must always act normal to the dislocation axes in order to expand regions of favorable slip and to shrink regions in which the slip opposes the applied stresses.

2) *Climb forces:* Since supersaturations of interstitial atoms, vacant lattice sites, and their aggregates, can be eliminated by climbing dislocations, it follows that such supersaturations cause forces on dislocations, inducing them to climb.

The magnitude of these climb forces can be found, according to Bardeen and Herring (*8*) thus. If a dislocation of length l and Burgers vector b, including an angle θ with the dislocation axis, climbs through the distance dx, this corresponds to the transfer of volume

$$dV = lb \sin \theta \, dx$$

of the material and thus corresponds to the generation or destruction of

$$dN = lb \sin \theta \, dx/\Omega$$

vacant lattice sites or interstitial atoms, where Ω is the atomic volume. The Gibbs free energy gained (or expended) thereby is

$$dG = \pm \mu_d \, dN,$$

where μ_d is the chemical potential of the single defects involved, either that of vacancies or interstitials. The chemical potential, in turn, is given by

$$\mu_d = kT \ln (c/c_o)$$

where c/c_o is the supersaturation. Therefore, the climb force per unit length of dislocation is found as

$$|F_c| = \frac{1}{l}\frac{dG}{dx} = \frac{1}{l}\mu_d\frac{dN}{dx} = kT\frac{b \sin \theta}{\Omega} \ln \frac{c}{c_o}.$$

Thus, the climb force only acts on that component of the Burgers vector which is normal to the dislocation axis. It drives the dislocations normal to their slip planes in that direction which tends to establish equilibrium of the point defects.

Frictional Forces

Dislocations will change their positions only so long as the driving forces are larger than the sum of all forces resisting dislocation motion. Some of such forces are quite obvious. For example, internal stress systems of whatever origin, and including those set up by other dislocations, may counteract the externally applied stresses. Furthermore, the motion of a dislocation is blocked at crystal boundaries by incoherent precipitates and by rigid surface layers, as for example oxide layers on free metal surfaces. Moreover, a so-called "imperfect" dislocation, a dislocation whose Burgers vector leads from or to a metastable position, and which, therefore, borders a stacking fault, is attracted toward that fault, and thus cannot be easily moved away from it. But a finite force is even required to move single "perfect" dislocations, (that is, dislocations whose Burgers vectors are lattice vectors and which, therefore, do not border stacking faults), in otherwise perfect crystals.

In a homogeneous continuum, free of relaxation effects, a dislocation would be expected to move in response to infinitely small stresses, since its energy would not depend on position. To a large extent, the same can be said about a perfect dislocation in a crystal with respect to all regions in which linear elastic behavior obtains. In the dislocation core, however, the normal theory of elasticity breaks down, and the detailed positions of the atoms have to be considered. The pattern of atomic arrangement in the core changes periodically with the position of the dislocation, so that the same atomic arrangement in the core is restored every time that the axis moves through intervals of one Burgers vector. Slightly different energies are of necessity associated with these different configurations, that is, the core energy is a periodic function of the dislocation position with a periodicity interval *b*. The force required to overcome the steepest gradient of the core energy as a function of dislocation position is the smallest force which will move a dislocation.

This force has been named the Peierls-Nabarro force for the scientists who made the first calculations of its magnitude (*7, 9*). Another classic paper on this topic was contributed by Huntington, Dickey, and Thomson (*10*). The calculations are most difficult, and the results are at best reliable only to a factor of two or so. This is because the effect is due to only small fluctuations in the core energy, which itself cannot be properly evaluated. The Peierls-Nabarro force may perhaps be given approximately by

$$\tau_D \cong \frac{2}{3}\,\tau^2_{crit}/G$$

where G is the modulus of rigidity and τ_{crit} is the stress at which an ideal dislocation-free crystal would slip on the crystal plane considered (*11*). The value τ_{crit} is always much larger still than the highest fracture stresses of metals, namely on the order of 3 to 10 percent of the numerical value of *G*, depending on crystal type and slip plane. For the slip planes of the most ductile metals, the lower figure applies, but in semi-conductors $G/\tau_{crit} \cong 5$ to 8, and τ_D in these is thus much higher than in metals.

In actual fact, τ_D is an overestimate for at least two reasons. Thermal activation helps dislocations to overcome the Peierls-Nabarro force, and it is believed that, owing to uncorrelated atomic vibrations, including zero-point motion, the positions of dislocation axes are not defined with mathematical precision (*11*). In other words, at no time do the actual atomic positions precisely conform to those which would be calculated for a dislocation. These deviations from the ideal atomic positions have the consequence that the dislocation axis has an "uncertainty," or expressed differently, every dislocation in a real crystal always corresponds to a spectrum of infinitely many dislocations with infinitely small Burgers vectors, distributed over a small region. If the extent of this "uncertainty region" becomes comparable with or larger than one Burgers vector, then the dislocation does in fact straddle a significant part of, or more than, one complete period of the core-energy curve. Then the effective frictional stress resulting from the periodic change of core energy with dislocation position becomes markedly reduced. A theoretical estimate of the effect (*11*) indicates that it operates with moderate strength in close-packed metals — aluminum, copper, nickel, zinc, brass, silver, and gold—even to the lowest temperatures. However, in body-centered cubic metals, particularly iron and simple steels, the effect should cease somewhere near or below room temperature. Experimental evidence concerning the magnitude of the "uncertainty" effect is still lacking.

Additional frictional forces act on dislocations in impure metals and in alloys, since even very small concentrations of impurity atoms can effectively anchor dislocations. At least three different mechanisms are responsible:

1) Cottrell (*12*) was the first to point out that impurity atoms in nominally "pure" metals must interact with the normal stresses of dislocations. He argued that, during suitable annealing treatments, all interstitial atoms and those substitutional impurity atoms which are larger than the atoms of the matrix are attracted to regions of dilatation, while substitutional impurity atoms smaller than the matrix atoms diffuse to regions of compression. "Impurity atmospheres" should thereby form, reducing the energy of the dislocations by the amount of the binding energies. Dislocations thus anchored are practically immobile until a sufficiently high stress, often combined with thermal activation, frees them abruptly. The discontinuous yielding of iron has been explained on this basis, with carbon and nitrogen causing the anchoring.

2) Interstitial impurity atoms of a kind which impose a tetragonal strain on their immediate surroundings can interact with the shear stresses of a dislocation as substitutional atoms of different sizes interact with the normal stresses. This type of anchoring, first investigated by Nabarro (*13*) and later by Schoeck and co-workers (*14*), seems to be of particular importance in iron, with carbon and nitrogen again as the active impurities.

3) It was pointed out by Suzuki (*15*) that substitutional atoms in nominally "pure" metals as well as in alloys would have to have either a higher or a lower free energy within stacking faults than they have in the rest of the matrix. Therefore, just a few atomic jumps by substitutional atoms close to extended dislocations will cause the stacking faults between the partial dislocations either to become enriched or to be depleted of atoms of solute. In both cases, the energy of the extended dislocation would be reduced, and the dislocation would be anchored as in the other two cases.

The three anchoring mechanisms are largely responsible for the hardening which results when pure metals are alloyed. That is, they greatly contribute to one of the technologically most important effects by which the properties of metals can be manipulated.

Origin of Dislocations

The line energy of stable perfect dislocations in metals amounts to about 2 ev/Å. Dislocations, being line defects, are generally hundreds or even thousands of angstroms long and thus have energies of many electron volts. This means that dislocations never exist in thermal equilibrium. Furthermore, when a dislocation loop is generated on a slip plane, the work done by applied shear stresses of magnitude up to the fracture stress is small compared to the dislocation energy, except for loop diameters very much larger than one Burgers vector. Consequently, externally applied stresses of usual magnitudes cannot possibly generate dislocations in metals that previously were dislocation-free. Instead, stresses in the order of τ_{crit} are required for the initiation of dislocations.

These facts were realized as soon as the concept of dislocations had been introduced into solid-state theory. However, valid answers to the problem of how dislocations originate in undeformed crystals were comparatively long delayed. By now, various kinds of mechanisms operating during crystal growth are known to produce dislocations, mainly in the form of regular networks (Fig. 4). Also, the condensation of vacant lattice sites in the form of coherent layers, in effect eliminating parts of atomic planes, leads to dislocation loops. These outline the condensed sheets of vacancies after the adjoining atomic planes have collapsed to fill-in the disc-shaped cavity. Figure 5 gives an example of such loops in quenched aluminum as seen with the electron microscope. Finally, localized stresses, such as those that arise at the boundaries of precipitates during heating and cooling, because of differences in thermal expansion coefficient, may become large enough for the generation of sequences of prismatic dislocations (Fig. 6), as was first shown by Jones and Mitchell (*16*).

Geometry of Glide Motions

In the absence of stresses and of supersaturated vacancies and interstitials, dislocations shorten as far as is possible. Correspondingly, dislocations in newly grown, then slow-cooled, undeformed (colloquially "as - grown") crystals are present in the form of networks with straight links (Fig. 4). Similarly, line tension keeps gliding dislocations on slip planes straight, or only slightly curved, unless these dislocations meet localized obstacles, such as precipitates, and intersecting dislocations, between which they will be driven forward by the resolved shear stress while held back, locally, at the obstacles. The nature of the obstacles determines whether the dislocations will intersect or circumvent (*17*) the obstacles (Fig. 7). Circumventing of obstacles occurs in metals containing incoherent precipitates. The finer the dispersion of the precipitates, the higher the stress necessary to move the dislocations past them. This is the major cause of precipitation hardening.

A dislocation whose axis lies completely in one plane with its Burgers vector, which at the same time is an easy slip plane, often is referred to as "glide" dislocation. Besides these, real crystals contain "prismatic" dislocations whose axes do not lie in a plane with their Burgers vectors, so that their slip surfaces are general prisms—whence the name. These usually result from vacancy condensation, or from high localized stresses at interfaces between matrix and precipitates (see Figs. 5 and 6). In general, the resolved shear stresses on prismatic slip surfaces differ from place to place, and different sections of the slip surfaces offer different resistance to glide motion, depending on their crystallographic orientation. Therefore, prismatic dislocations do not execute the same relatively simple motions as glide dislocations, but parts of them may be driven in one direction, other parts in the opposite direction, and some sections will be immobile.

Figure 8 illustrates the motion resulting when a shear stress acts on a straight link of a prismatic dislocation formed through vacancy condensation, as in Fig. 5, while the neighboring parts of the loop remain immobile. Successive stages are labeled *a, b, c,* and so forth. From this figure it is apparent that dislocation multiplication is effected, since the original dislocation link is restored every time a cycle is completed and an expanding closed loop has been generated.

This is but one example of dislocation multiplication by the so-called Frank-Read mechanism (*18*). There is now no doubt that dislocations are multiplied by this mechanism during plastic deformation.

The case shown in Fig. 8 is simpler than many others, since the dislocation segments which stay fixed, the so-called "pole" dislocations, have a Burgers vector lying in the plane in which the multiplication takes place. Although this is not necessary, the sweeping dislocation may be anchored by dislocations having a Burgers vector component normal to the plane of multiplication. In such a case, each revolution of the sweeping dislocation moves it perpendicular to its original slip plane on to the next atomic plane (*19*). Still other forms arise if one side of the sweeping dislocation is attached to pole dislocations which cause the sweeping dislocation to move down, while the pole dislocations on the other side cause a planar motion, or a motion in the upward direction. Finally, the sweeping dislocation can spin around only one pole if it is somehow anchored at its other end. The last case results in a single spiral, and an operating source of this type was directly observed in the electron microscope as shown in Fig. 9 (*20*).

A sequence of dislocations emitted from the same source may form a so-called "pile-up" when the leading dislocation is held up at some obstacle and the trailing ones pile up behind it. Such pile-ups are an important feature in the plastic deformation of metals. After this was first clearly recognized (*21*), the concept of dislocations, emitted from easily acting sources and piled up behind obstacles, has been incorporated into many theories of

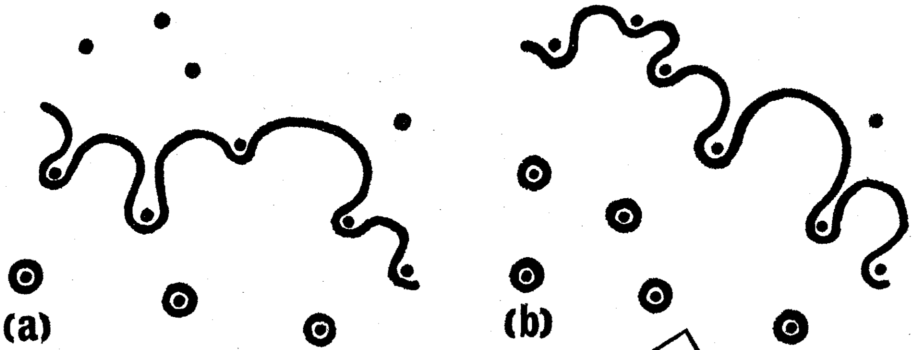


Fig. 7 (above). Two successive stages in the movement of a glide dislocation, circumventing and encircling obstacles dispersed over its slip plane.

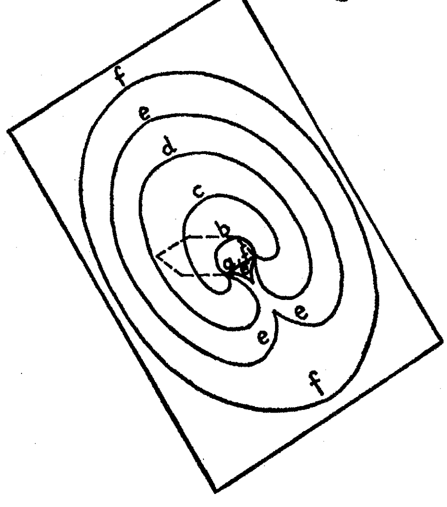
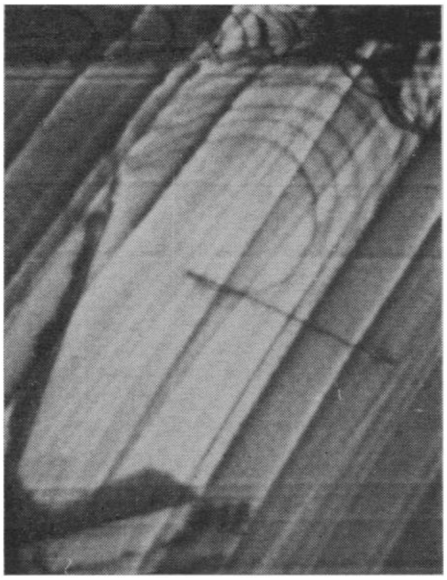


Fig. 8 (right). Dislocation multiplication, arising when one link in an otherwise anchored prismatic dislocation loop (of hexagonal shape) moves under the influence of a resolved shear stress acting on it. Successive stages are labeled *a, b, c,* and so forth. This is one example of the so-called "Frank-Read" mechanism.

plastic deformation and "workhardening." The term "workhardening" refers to the phenomenon that the applied stress must be increased continously in order to deform a metal, since metals normally offer an increasing resistance to further deformation, that is they "workharden."

Where sequences of dislocations reach surfaces, surface steps or "slip lines" are formed. The investigation of slip lines is a useful tool for the study of the distribution of mobile dislocations in crystals, their mean free paths, and other characteristics. However, it must be realized that slip lines show where dislocations have moved, not where they are. Furthermore, slip lines are, by definition, surface phenomena, and care must be exercised in deciding to what extent the conclusions based on slip-line investigations may be applied to dislocation behavior in the bulk of specimens. Figure 10 is a micrograph of slip lines on α-brass.



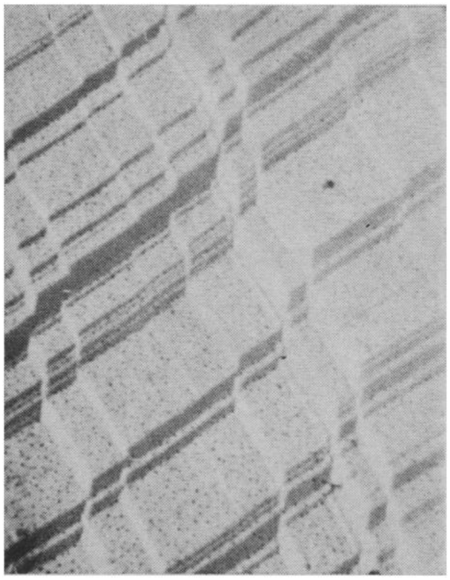


Fig. 9 (left). Frank-Read dislocation source of a type in which the sweeping dislocation spirals around a pole dislocation (*20*) ($\times$ 45,000). Fig. 10 (right). Stepped surface due to slip on atomistic planes of two slip systems in α-brass. Such steps are named slip lines ($\times$ 9000).

For screw dislocations, that is, for dislocations having a Burgers vector parallel to the axis, no specific slip plane is defined, but all planes containing the dislocation axis can act as slip plane, at least from a purely geometrical viewpoint. Therefore, gliding screw dislocations can move from one slip plane into another one, inclined to the first, in a motion named "cross slip" (*22*) (Fig. 11*a*), provided that at least two planes intersecting the screw dislocation have a low Peierls-Nabarro stress. "Double-cross slip" (Fig. 11*b*) is repeated cross-slip motion, first from one slip plane into another slip plane intersecting the former in a direction parallel to the screw dislocation, and then back into a plane parallel to the original slip plane. It is generally believed that cross slip and double-cross slip play an important role in the behavior of actual dislocations in real crystals.

In discussing the nature of dislocations, reference has been made to the fact that "perfect" dislocations may split into "partials," including ribbons of stacking fault between them whenever metastable positions of low energy exist on the slip plane. Such extended dislocations are quite common in layer crystals composed of close-packed planes with comparatively weak binding between the planes. In face-centered cubic metals, the ordinary glide dislocations on $\{111\}$ planes with ½ $<110>$ as Burgers vectors can split up in that same manner. Stacking faults formed in this way, dragging behind moving partials, are seen in Fig. 12 (lower right). Two partials result whose Burgers vectors are of type ⅙ $<211>$, but in face-centered cubic metals, the stacking fault energies are generally much higher than in layer crystals, and the equilibrium widths to which the dislocations split are consequently much smaller. Glide dislocations on basal planes in hexagonal metals can split into partials in much the same manner.

Although the motion of isolated dislocations on slip planes is hardly affected by the presence or absence of splitting into partials, the formation of extended dislocations is important in at least three ways. (i) Intersections between extended dislocations are more difficult than between unextended dislocations (*6*). (ii) Unless the Burgers vectors of the partials happen to be parallel, (*23*) cross slip of extended dislocations can take place only if the two partials are first compressed to eliminate the stacking fault. This process may require high stresses and thermal activation. (iii) Climb of extended dislocations similarly requires the prior elimination of the stacking fault ribbon over the length of the climbing sections.

Since all three mechanisms, dislocation intersections, cross slip, and climb, play a considerable role in the mechanism of plastic deformation, the magnitude of the stacking-fault energy of metals is an important parameter influencing the mechanical properties of crystals. Seeger (*24*) was the first to draw attention to this fact. Usually, the stacking-fault energies of alloys are lower than those of pure metals. Thus, the dislocations in alloys tend to be more widely split than in pure metals. A great part of the differences in mechanical behavior between pure face-centered cubic metals and alloys can be explained on this basis.

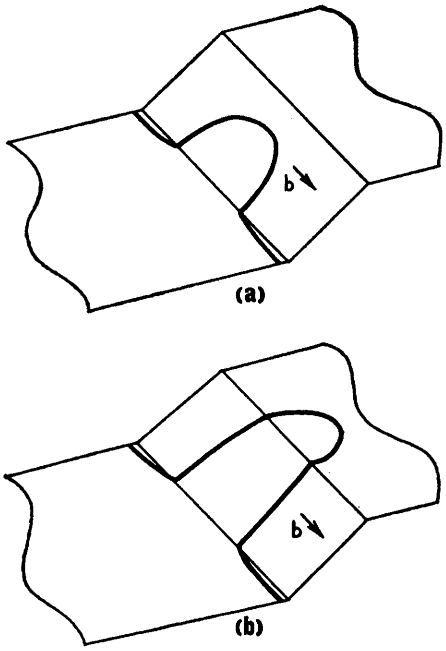


Fig. 11. Schematic representation of the geometry of cross slip (*a*) and double-cross slip (*b*) by dislocation parts having screw orientation at the moment of transfer into the intersecting slip plane.

Velocity of Dislocations

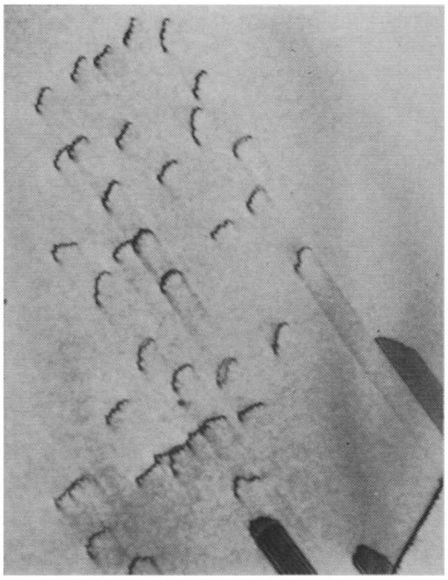
That the speed of sound plays the same role in dislocation theory as the speed of light in relativity theory was first shown by Frank (*25*) and Eshelby (*26*). Dislocations are elastic disturbances in crystals, which cannot be propagated faster than the speed of sound in the same material. This fact expresses itself so that the energy of moving dislocations rises with the dislocation velocity, v, as $\{1-v^2/c^2\}^{-\frac{1}{2}}$ where c is the speed of sound in the material in which the dislocations are situated. Hence, the dislocation energy tends toward infinity as the velocity tends towards the speed of sound,—about 2 kilometers per second in metals. However, dislocations in actual crystals typically move much more slowly, namely, several microns per second at stresses close to the yield point, from observations on moving dislocations in the electron microscope. However, v, the velocity of dislocations, rises sharply with increasing stress, τ, namely as

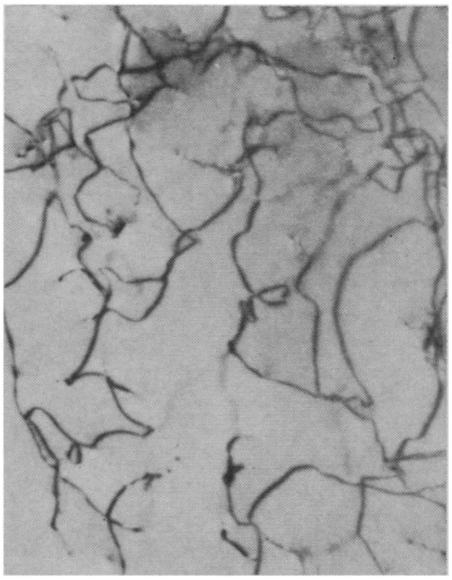
$$v = v_0\,(\tau/\tau_0)^n.$$

The first direct measurements of dislocation speed as a function of stress which yielded the named relationship have been performed by Johnson and Gilman (*27*), who revealed and studied dislocation etch pitting on the surfaces of lithium fluoride crystals. Subsequently, similar measurements were performed on iron-silicon alloy (*28*) and on semiconductors (*29*). These measurements showed the exponent n to be near 1.5 for the case of semiconductors, while n is about 25 for lithium fluoride and above 40 for silicon-iron. The reason for this enormous difference in the dependence of dislocation velocity on stress is not yet perfectly understood. However, the pronounced differences between mechanical behavior of the brittle semiconductors, such as germanium and silicon, and that of ductile metals can be understood from the great differences of Peierls-Nabarro stress on the one hand, and of stress sensitivity of dislocation velocity on the other hand. The semiconductors are brittle because their Peierls-Nabarro stress is high, while at the same time dislocations in them cannot multiply at a rapid rate. Thus these materials cannot deform at a rapid rate when subjected to increasing stresses, and therefore, the semiconductors fail by brittle fracture at all except quite high temperatures. In ductile metals, by contrast, the Peierls-Nabarro stress is low and the dislocation velocity rises very steeply with increasing stress. Thus dislocation motion, and with it dislocation multiplication, is always fast enough to prevent brittle failure.

Jogs and Tangling of Dislocations

Since a moving dislocation causes a translation over the area swept out by its axis, other dislocations, when intersected, are correspondingly sheared. Thereby "jogs" are formed on the intersected dislocations, whose length and direction equals the length and direction of the Burgers vector of the cutting dislocation. The problem is symmetrical in the sense that the moving dislocation also acquires jogs, where, in turn, the length and direction of the jogs are equal to the Burgers vectors of the intersected dislocations (*30*). The slip surface of any dislocation, generated as we saw by all lines parallel to the Burgers vector intersecting the dislocation axis, thus becomes stepped if jogs are formed on the dislocation. Therefore, in order for conservative motion of a jogged dislocation to take place, the jogs must slip along the steps in the slip surfaces, that is, they must shift tangentially along the gliding dislocation, except in the case of jogs on edge dislocations. However, it is believed that the jogs will not always follow the directions of the aforementioned steps in the slip surface, particularly not those on screw dislocations (*31*), which means that the jogs will not necessarily move "conservatively," and thus will generate





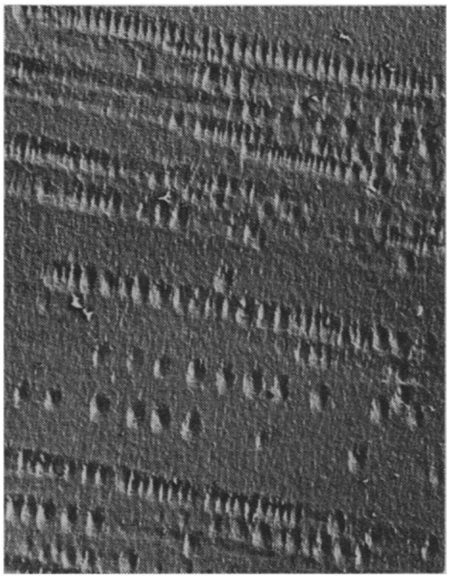


Fig. 12 (left). A group of dislocations moving away from a low-angle boundary in an electron transparent foil of stainless steel viewed in the electron microscope. The direction of motion of dislocations can be recognized from the trails left behind by them. The latter are a diffraction effect which fades after a little while. Stacking faults recognizable by striped contrast are dragged out behind moving partial dislocations in the same group (*20*) (× 60,000). Fig. 13 (center). Tangled dislocations in an electron transparent foil gained from aluminum, lightly deformed in bulk, then thinned down (*37*) (× 25,000). Fig. 14 (right). Sequences of etch pits, formed at the points of emergence of dislocations, arranged in piled-up groups. The material is deformed α-brass, lightly doped with cadmium to "decorate" the dislocations, and thus to make them susceptible to preferential-etching attack. The polished and etched surface was replicated with silicon monoxide, shadowed with tungsten oxide, and examined in the electron microscope (× 25,000).

point defects. Only intersections between dislocations with Burgers vectors each in the slip plane of the other dislocation do not cause permanent jogs; in this case, shearing of the dislocations takes place in their own slip surfaces, and thus the jogs are immediately eliminated by simple glide. The stress required for jog formation, and the drag on dislocations by jogs, have often been cited as mechanisms of work hardening, particularly by Basinski, van Bueren, Hirsch, and Mott (*32*).

While there is little doubt that point defects are generated during glide, and while there is also little disagreement that the nonconservative motion of jogs is at least partly responsible, considerable disagreement exists concerning the precise mechanisms operating. In particular, Seitz (*33*) and Koehler (*34*) have proposed several different geometrical possibilities for point defect formation during slip and, more recently, an additional mechanism was suggested (*35*).

Namely, as already explained dislocation axes in actual crystals are not mathematical lines but are better described by their "uncertainty regions," that is, we may think of the axes as smeared out within their uncertainty regions. Correspondingly, the slip surfaces of dislocations are also not defined with mathematical precision, and the dislocation must, so to speak, diffuse away from their original slip surface in the course of their slip motions, always shifting their imprecisely defined slip surface according to their momentary axis position. True, this effect will be large only for screw or near screw dislocations, but, in close-packed metals, the effect is believed to remain noticeable down to the lowest temperatures. The slight statistical departures of dislocations and jogs from mathematically precise planes must, of course, generate vacancies or interstitials, or both.

Regardless of which process or processes of defect generation should eventually be proved to be the most important, the defects formed must be in considerable supersaturation and must be eliminated (*33*). The most obvious sinks (sites of elimination) for these vacancies and interstitials are stationary or slow-moving dislocations which climb on absorbing these defects. We believe that this is the most important reason why dislocations in most pure metals, deformed at intermediate temperatures, do not form regular arrays, but form three-dimensional "tangles" of dislocations, interspersed with prismatic loops (Fig. 13). By contrast, most scientists working on this subject believe that cross slip or dislocation interactions rather than this peculiar kind of climb is responsible.

The fact that face-centered cubic alloys are not as susceptible to dislocation tangling as are pure face-centered cubic metals, but tend to form dislocation pile-ups (Fig. 14), would follow naturally from both explanations, since in alloys the stacking-fault energy is usually lower than in the pure metals, and thus the glide dislocations in alloys are definitely split. Unless the two partials are somehow pushed together, which requires considerable stresses or energies, the dislocations are bound to specific atomic planes and therefore cannot cross slip, cannot climb, and do not have a noticeable uncertainty normal to the slip planes associated with themselves. The origin of the dislocation tangles, then, is one of the problems in the theory of moving dislocations that is still under dispute.

Summary

Motions of dislocations are by far the most important cause of plastic deformation in crystals, and thereby determine the mechanical properties of crystals in general, and of metals and their alloys in particular. Our understanding of the plastic behavior of metals, as distinct from technical know-how, be it in rolling, bending, wire-drawing, turning, filing, or whatever, extends as far as our understanding of both the forces on dislocations and the motions executed in response to these forces, taking due account of the most complicated interactions between the dislocations.

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