LATTICE IMPERFECTIONS AND PLASTIC DEFORMATION IN METALS

I. NATURE AND CHARACTERISTICS OF LATTICE IMPERFECTIONS, NOTABLY DISLOCATIONS.

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Not more than 30 or 40 years ago our knowledge of the physical properties of metals was based almost entirely on experience. After the first world war a change came, brought about by a gradual expansion in the study of the physics of metals, the object of which was to justify theoretically the observed characteristics. Recently, much progress has been made in the field of plastic properties of metals. The new conception, that the plastic deformation of materials is intimately connected with the occurrence and concentration of imperfections in the regular structure of the atoms, that is, of lattice imperfections, has been found to be very fruitful. It can be applied not only to metals, but also to other materials.

Introduction

At the beginning of this century it was still the general belief that the more important physical properties of crystalline substances could be explained exclusively in terms of the periodic arrangement of the component atoms. It was generally thought that any defects in that structure (irregularities in the crystal lattice) had but little bearing on the characteristics, and these irregularities were accordingly disregarded; the crystal lattices were held to be perfect.

Although this standpoint originally met with considerable success, e.g. in the explanation of X-ray diffraction patterns and in the theories of cohesion between the atoms in crystals, it was soon found that there were many phenomena which would definitely not answer to theoretical considerations on that basis. Particularly in the study of transport phenomena such as the conduction of heat and electricity and diffusion, insurmountable difficulties were encountered.

The first and most important step towards a better understanding of these phenomena can be said to have been the recognition of thermal lattice vibrations. The atoms at the lattice points of a crystal vibrate, practically as harmonic oscillators, about their position of equilibrium, and the energy with which they do this, and hence the amplitudes, increase considerably with the temperature.

Although this conception led to important additions to the theory of certain transport phenomena, many effects were still without any satisfactory explanation, in particular, those which relate to the mechanical properties of crystals. Ultimately, therefore, it was found necessary to take into consideration other imperfections in the crystal structure. To enter into a general discussion on the influences of lattice imperfections upon the physical and chemical properties of crystals, would not be practicable within the scope of this article. We shall therefore limit our discussion to two articles dealing with some of the phenomena connected with the plastic deformation of metals. For a review of the many other domains in which lattice defects are of interest, the reader may refer to a recent article in this Review by G. W. Rathenau 1), and to the comprehensive survey by F. Seitz 2).

Some phenomena related to the plastic deformation of metals

Microscopic examination of the surface of metals which have been polished and subsequently deformed, usually reveals a pattern of fine, more or less straight lines (fig. 1a). The more the metal is deformed, the more clearly these so-called slip lines become visible and the greater is their number. The orientation of these lines appears almost invariably to correspond to the planes in the crystal in which the atoms are most closely packed. This relationship is found to be the most marked in metals with the closest packed crystal structure. In metals having body-centred cubic lattices which are not packed as closely as possible, the slip lines are somewhat irregular, but the relationship between the slip lines and the crystal planes still remains.

Examination of the slip lines under high magnification by means of the electron microscope, shows that what appear to be single lines when seen with lower magnification, are in many cases groups of some

1) G. W. Rathenau, Philips tech. Rev. 15, 105-113, 1953 (No. 4).
tens to hundreds of adjacent lines (fig. 1b), which may be termed **slip bands.** Single lines are seen as well, however.

A study of the slip lines or bands has revealed the fact that these are in effect "steps" in the surface of the metal, of which the height may vary from some tens to some thousands of times the atomic spacing. The conclusion to be drawn is clear, viz. that in the deformed metal, translations take place along certain crystallographic planes whose orientation corresponds to the direction of the slip line.

Fig. 2. Diagrammatic representation, on a highly exaggerated scale, of slip in a metal crystal, g slip planes; \( \tau \) shear stress.

This mechanism is sketched in fig. 2, which refers to simple slip lines and is, of course, drawn on an exaggerated scale. This deformation mechanism, which appears to be of a very universal nature, is known as **gliding** or **slip**; the lattice plane along which the slip takes place is termed the **slip plane,** and the direction of the translation, the **slip direction.** The slip plane and direction together constitute the **slip system.**

The theoretical critical shear stress \( \tau_{cr} \), that is, the minimum shear stress necessary to produce a slip translation of this kind, is of the order of 0.1 \( G \), where \( G \) is the modulus of rigidity, or torsional modulus of the material.

This result is obtained in the following manner (vide Frenkel 3)).

Take the case of two neighbouring rows of atoms in a simple crystal. Let \( a \) denote the spacing of the atoms in the undistorted condition. The force required to move the one row an infinitesimal distance over the other depends upon the displacement that both rows have already undergone. In the undeformed state, which is the state of equilibrium, this force is zero. Each time the rows are displaced an integral number of times \( a/2 \) from the undeformed state, an equilibrium state is restored, as a symmetrical configuration of the lattice is again reached. It would appear to be a reasonable assumption that the relationship between the shear stress \( \tau \) and the relative displacement \( x \) might be approximated by a simple periodic function:

\[
\tau = k \sin \frac{2\pi x}{a}.
\]

With only a small displacement this becomes \( \tau = k.2\pi x/a \).

In this case also Hooke’s law applies, which states that:

\[
\tau = Gx/a.
\]

From this, \( k = G/2\pi \); hence:

\[
\tau = \frac{G}{2\pi} \sin \frac{2\pi x}{a}.
\]

If \( \tau \) is larger than \( G/2\pi \), it follows from this formula that the displacement \( x \) of the atoms is unrestricted and slip will occur. Closer investigation yields a value for the theoretical critical shear stress slightly lower than \( G/2\pi \), although still of the order of 0.1 \( G \).

For most metals, $G$ lies between some thousands and some ten-thousands of kilograms/mm², from which a theoretical critical shear stress of a few hundreds or a few thousands of kilograms/mm² follows. It is a well-known fact, however, that metals can be very much more easily deformed than would appear from these values. The observed critical shear stress of well-annealed single crystal occurs between 0.1 and 10 kgf/mm²; annealed polycrystalline metals yield somewhat higher values but still very much less than the theoretical shear stress.

To explain this ease of deformation, it was postulated that a particular kind of lattice defect, namely dislocations, occur very frequently in every crystal. It was very soon apparent that such dislocations do play a dominant rôle in the process of deformation of a metal.

We have referred above to the critical shear stress of well-annealed materials. The annealing minimizes the consequences of any previous deformations which would otherwise manifest themselves, for example, by the increase which occurs in the stress required to deform a metal as the deformation itself increases (fig. 3). This effect is known as work hardening (it does not occur in all materials; substances such as pitch, for example, undergo no work hardening). A severely deformed and unannealed metal may have a critical shear stress many tens of times higher than that in the annealed state.

![Diagram showing the relationship between the deformation $\varepsilon$ and the shear stress $\tau$ in a pure metal. The straight part at the commencement of the curve represents the elastic deformation. When the critical shear stress $\tau_c$ is exceeded, plastic deformation sets in, this being accompanied by strain hardening, as a result of which the characteristic assumes a parabolic form (in cubic metals). In practice, irregular curves are often encountered, but the general form remains the same.](image)

As we shall see presently, work hardening in metals is attributable mainly to the mutual interaction of the dislocations associated with the deformation.

The extent to which a metal can be deformed is dependent not only on the previous deformation; variations in the temperature will also affect the ductility, albeit to a relatively small degree. Thus, materials which are brittle when cold are nearly always ductile at elevated temperatures. Foreign atoms or conglomerates in a metal also affect the ductility appreciably; the various hardening processes for metals are based on this fact 4).

Lastly, it appears that the plastic deformation of a metal affects its general physical properties in varying degrees. Amongst other things, the electrical conductivity will change, and it is found that the study of this and other subsidiary consequences of plastic deformation can reveal valuable information about the behaviour of lattice imperfections.

In this section we have surveyed only the ground on which our further considerations are to be based. Before entering into a detailed discussion of the relationship between the mechanical properties of metals and lattice imperfections in crystals, however, it will be necessary for us to say a little more about present-day conceptions of the nature and behaviour of possible kinds of lattice imperfections. The rest of the present article is accordingly devoted to this subject.

Possible kinds of lattice imperfections

Imperfections in a crystal lattice may be one-, two- or three-dimensional. Also, singularities may occur at the lattice points themselves, in which case we might speak of zero-dimensional, or point defects. The three-dimensional defects such as macroscopic holes or inclusions, precipitates etc, and two-dimensional faults which may be taken to include the crystal boundaries and surface layers, can be dealt with quite briefly. In recent years, several articles have appeared in this Review on the subject of three-dimensional imperfections 4); as to the two-dimensional defects, these can often be successfully interpreted as more or less ordered associations of linear and point defects.

Dislocations belong to the category of linear imperfections. Point defects may be taken to include vacancies, i.e. lattice points where an atom is missing, and further, interstitial atoms (atoms at intermediate points in the lattice) and impurities (foreign atoms). Whereas the concept of dislocations was originally introduced to provide an explanation of the mechanical properties of crystals, vacancies etc. have been postulated to promote a deeper insight into diffusion effects and other trans-
port phenomena. It is only in recent years that a close relationship has been found to exist between the different kinds of lattice imperfections.

The conception of linear lattice irregularities, i.e. dislocations, was introduced in the theory of metals independently by Taylor, Orowan and Polanyi in 1934. Since 1939, when Burgers made the first detailed theoretical study of the behaviour of dislocations, the work of N. F. Mott and his co-workers at Bristol has been mainly responsible for the almost general acceptance of the hypothesis of dislocations. The work of the British scientists gave the impetus to the first direct experimental pointers to the existence of dislocations in crystals.

Dislocations

The dislocation concept links up very closely with the most important deformation mechanism in metals, namely the above-mentioned “slip”. Fig. 4 illustrates the present conception of this mechanism. The two halves of the crystal do not move over each other as a whole; this, as already mentioned, would require a much higher shear stress than that which is indicated by experiment.

The displacement commences at one side (left, in fig. 4) of the crystal and is propagated very rapidly to the other side. A situation whereby all the atoms simultaneously assume non-equilibrium positions never arises; the deformation at a given instant is always localized within a zone of some 3 or 4 times the atom spacing. This zone, of which fig. 4 shows only a cross section, i.e. one atomic layer (this layer should be regarded as arbitrarily extended in a direction perpendicular to the plane of the drawing), contains a linear lattice defect, and it is this that has come to be known as a dislocation.

The displacement of the two halves of the crystal in relation to each other has already taken place to the left of the dislocation in fig. 4d, but not on the right-hand side. The deformation of the crystal as a whole approaches completion as the dislocation moves to the right along the slip plane; once it has arrived at the opposite side of the crystal (fig. 4f) the shear is complete. In this example it is

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Fig. 4. Diagram representing on an atomic scale the occurrence of slip in crystals, according to the dislocation mechanism. The full lines represent the lattice planes. The two parts of the crystal do not slip simultaneously as a whole along the slip plane; the deformation is localized within a small zone (shown in dotted-line rectangles). Only the cross-section in one atomic plane is illustrated, but it should be visualized as extended without limit perpendicular to the plane of the drawing. Movement of the dislocation from left to right (in the sequence of figures a to f) completes the slip.

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viz. the spiral growth of crystals from super-saturated vapour, as well as the recently discovered sub-structures of photo-sensitive silver bromide crystals. Further direct evidence for the existence of dislocations in metals has been gathered by workers in the United States, from the study of crystal boundaries.

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8) F. C. Frank, Advances in Physics. 1, 91, 1952. See also fig. 2 in the article referred to in footnote 1.
seen that plastic deformation due to slip is simply the movement of a dislocation.

The critical shear stress is now the force required to initiate a dislocation and cause it to be propagated. To produce this effect, only a fraction of the total number of atoms in the region of the slip plane need be simultaneously in a condition of increased energy, so that it follows that this force is very much less than in the case of a perfect lattice.

Dislocations occur not only in metal crystals; in principle they may arise in any crystalline substance. Whether or not they play an important part in the behavior of the material, depends very largely on the crystal structure and the cohesion between atoms. Recent investigations have indicated that in ionic crystals and in non-polar semiconductors, dislocations probably play quite an important role, especially with regard to the optical and electrical properties of these materials. The study of dislocations in non-metals is still in an early stage, however, and we shall therefore not concern ourselves with it here.

Types of dislocation

Let us now consider the concept of dislocations in a broader sense. The axis of the linear lattice imperfection (which need not necessarily be a straight line) is called the dislocation axis. The dislocation separates the part of the lattice which has already slipped, from the part which has not yet undergone displacement. The degree and direction of the translation in the displaced zone is represented by a vector \( b \), known as the Burgers vector. Usually, the extent of the translation at a dislocation will be equal to the spacing of the lattice plane in the corresponding crystallographic direction, and the magnitude of the Burgers vector is then equal to this distance. The direction of the vector corresponds to that of the slip direction.

It is thus possible to differentiate between various types of dislocation according to the orientation of the Burgers vector with respect to the dislocation axis. The dislocation referred to in our discussion on the mechanism of slip, is the edge dislocation, as characterized by a Burgers vector perpendicular to the dislocation axis. Fig. 5a illustrates an edge dislocation in perspective, and it will be seen that, in order to complete the translation of the one part of the crystal over the other, the edge dislocation must be propagated in a direction at right angles to its axis, that is, parallel to its Burgers vector. Fig. 5b depicts a cross-section of an edge dislocation perpendicular to the dislocation axis. On close inspection of this diagram (and also of fig. 4), it will be seen that an edge dislocation can be considered as being initiated by the introduction of an additional plane of atoms perpendicular to the slip plane in one of the halves (in this case the upper half) of the crystal. As a result of this, variations in density of the crystal lattice are produced around the axis of an edge dislocation; in fig. 5b the upper half of the crystal in the region of the dislocation is more closely packed than normally, the lower half less so.

An exactly similar distortion of the crystal, having the same slip plane as in fig. 5, and also produced by the movement of an edge dislocation, can be thought of as being initiated by the introduction of an extra plane of atoms in the lower half of the crystal. The direction in which this dislocation would have to be propagated in order to complete the movement is then reversed in sign, and the more densely and less densely packed regions about the dislocation axis change places. We then speak of a dislocation of opposite sign to that of the original.

A second kind of dislocation is the so-called screw-dislocation, the Burgers vector of which is parallel to the dislocation axis. A dislocation of

![Fig. 5. a) Perspective diagram of an edge dislocation which has progressed half way through a crystal. The chain-dot line \( d \) is the dislocation axis; the light lines represent the atomic planes. The Burgers vector \( b \) is oriented in the direction of the shear stress \( \tau \) which strives to complete the slip. To do this, the dislocation must move in the direction of its Burgers vector, along the slip plane \( g \). b) Cross-section of an edge dislocation perpendicular to the dislocation axis as indicated by \( \times \); \( g \) is the slip plane, \( b \) the Burgers vector; \( h \) extra atomic half-plane.](image-url)
this kind is shown in perspective in fig. 6a. The regions in which slip has occurred (right-hand side) and has not occurred (left-hand side) are now separated by a linear lattice imperfection parallel to the direction of slip, in contrast to the edge dislocation, which is oriented perpendicular to the slip direction. Once more, in order to complete the translation, the dislocation must move in a direction perpendicular to its axis, but here this means perpendicular to the Burgers vector. In consequence of the peculiar mutual orientations of the Burgers vector and the dislocation axis, the screw dislocation has no specific slip plane, as this plane is defined as the plane through the dislocation axis and the Burgers vector. Whereas the edge dislocation, in order to accomplish slip, must move over a very definite slip plane, the screw dislocation is free to select its own direction of propagation (provided it is perpendicular to its axis).

This is not the only point of difference between the two kinds of dislocation. From fig. 6a it will be seen that a screw dislocation can be imagined as formed by making an incision of some depth (in the perfect crystal), this being followed by displacement of the part above the incision, parallel to the bottom of the cut, through a distance equal to the atomic spacing. The bottom of the cut is the dislocation axis. If we now describe a path round the dislocation axis, keeping within the same atomic plane, it will be seen that on completion of one whole turn we do not return to the point of origin as in the undistorted crystal, but that a translation has been accomplished equal in extent to the Burgers vector in the direction of the dislocation axis. In other words, a set of parallel atomic planes (perpendicular to the dislocation axis) in the undistorted crystal, forms, on the introduction of a dislocation, a helical surface having the dislocation line as axis. The atomic structure around a screw dislocation is depicted in fig. 6b. It is not possible to visualise the screw dislocation as being produced by introducing an additional atomic plane, and there is accordingly no question of any appreciable variation in density within the crystal, as occurs around edge dislocations.

A sign can also be attributed to the screw dislocation, since here again an equivalent translation can be brought about in two different ways: the screw dislocation can transform the atomic planes into either right-hand or left-hand helical surfaces. Fig. 6 shows a right-hand screw. Right and left-hand screw dislocations must move in opposite senses in order to produce the same translation.

Edge and screw dislocation represent only the extreme instances of the general conception of dislocations in which the angle between the Burgers vector and the dislocation axis is arbitrary.

The character of a dislocation need not necessarily be the same throughout its length: it may vary between one point and another. Moreover, a dislocation element may change its character in the course of propagation through the crystal. A probably common form of dislocation is the dislocation loop depicted in fig. 7, the feature of
which is that within the loop the crystal has slipped and that outside it no slip has occurred, since by definition, a dislocation separates a zone in which slip has taken place from another in which it has not. Everywhere within the loop, the relative amount of slip must be of the same magnitude and in the same direction; otherwise the loop would reveal branches. In other words, around a closed dislocation loop without any branch points, the Burgers vector — which determines the translation — is at all points constant in magnitude and direction. Only certain parts of the loop, as indicated by $E$ and $S$ in fig. 7, conform to the requirement of a purely edge or screw type of dislocation; at all other points the character of the dislocation is composite. The dislocation therefore nearly everywhere includes elements with at least some edge-character, i.e. differences in density are present, and the movement will be confined to a definite slip plane.

Dislocation loops are probably of frequent occurrence, since dislocations cannot be terminated somewhere in the middle of a perfect crystal. Along the dislocation there are invariably two zones which are displaced a definite amount with respect to each other. Where the dislocation is terminated, this relative displacement must disappear, and this can only happen on the surface of a crystal, at a crystal boundary, or at points where another dislocation is present. In the absence of any of these essentials, that is in the centre of a perfect crystal, only continuous, e.g. loop dislocations are possible.

Mechanical model of a dislocation

To study the characteristics of dislocations, it is usually sufficient in principle and without sacrifice of generality to confine our investigation to pure edge and screw dislocations. Using a mechanical model \(^{11}\) such as that shown in fig. 8, it is possible to illustrate the occurrence and movement of the dislocations.

A large number of permanent magnets are arranged on a board in a square pattern, to represent a single lattice plane of the crystal (which for convenience is supposed to be simple-cubic). Above this board a rack is mounted which is movable in either of the square-directions and which carries a number of pendula with “bobs” of soft iron. These iron balls also represent a lattice plane. In the state of equilibrium, each pendulum hangs over a magnet.

Whereas the “atoms” in the lower plane have fixed positions, those in the upper plane are capable of movement. To represent the inter-atomic forces, all the pendula are interconnected by small helical springs which exert no tension in the state of equilibrium, but which somewhat impede the approach or separation of any two of the pendula. When the rack of pendula is pushed to one side by hand, the magnetic forces between the magnets and iron balls is at first sufficient to resist the resultant “shear stress”. When the stress reaches a certain appreciable value, first one row of balls will be seen to jump over a distance equal to one lattice spacing in the direction of the stress; the application of a very slightly larger stress causes neighbouring rows in succession to carry out the same movement until finally the whole system of

pendula has moved up a distance equal to the lattice space. Thus a dislocation is seen to migrate through the lattice.

It is possible that the first row of balls to jump over will be perpendicular to the direction in which the stress is applied, as shown in fig. 8; in this case we have an edge dislocation. Again, the first row of "atoms" executing the movement may be parallel to the direction of the applied stress; in this case it is a screw dislocation which is seen to develop and move through the lattice (fig. 9). Which particular kind of dislocation will occur depends entirely on accidental factors.

Fig. 9. As fig. 8, but showing a screw dislocation. Movement of the pendulum rack causes the dislocation to move backwards or forwards through the model.

Movement of dislocations

Although the model provides only a very rough approximation to the real conditions, it is possible to note at once two important properties of dislocations. Firstly, the stress needed to initiate a dislocation is very much greater than that required to set it in motion. Secondly, under the influence of the stress, the dislocations are indeed propagated in such a way that the deformation of the crystal by slip is completed. Edge dislocations move in the direction of the stress and screw dislocations perpendicular to it.

It can be said, then, that a force operates on the dislocation. The magnitude of this force can be computed in the following manner (vide Mott and Nabarro 12).

Let us assume a dislocation of which the Burgers vector is $b$, running from one side of a cube-shaped crystal of dimension $L$ to the other side (cf. fig. 5). When it reaches the opposite side, one part of the crystal will have been displaced in relation to the other through a distance $b$. Let the components of the shear stress in the slip direction be $\tau_g$; the crystal area on which this operates is $L^2$, so that the work performed by the external force in producing the slip is:

$$W = \tau_g L^2 b.$$ 

The length of the dislocation line is $L$, as is also the distance travelled by the dislocation. If we now denote the force acting on a unit length of the dislocation by $F$, it may be said that the work done must be equal to:

$$W = FL^2.$$ 

(This is actually the definition of a "force acting on a dislocation").

The result of a shear stress whose component in the direction of the Burgers vector is $\tau_g$ (and it can be shown that this is universally valid) is therefore a force $F$ per unit length of the dislocation equal to

$$F = \tau_g b,$$

which tends to move the dislocation in the sense of completing the slip movement.

Only a slight force is needed for the propagation of the dislocation along its slip plane. If the whole dislocation is to be displaced by atomic distance, it is only necessary for the atoms themselves to move a fraction of this distance (fig. 10). In doing this, half of the atoms move under the influence of an attracting force and the other half under a repelling force, so that, to a first approximation, the forces counterbalance each other and the dislocation moves easily. It is owing to the discontinuous structure of the crystal that a dislocation cannot move in its slip plane entirely without effort, as might be inferred by the above remarks. Different positions of a dislocation axis between two lattice planes are not exactly equivalent.

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energetically. In metals of cubic and hexagonal structure, this results in only a slight reduction in the mobility of the dislocation in its slip plane. The mobility is in any case quite high and, as the atomic arrangement and hence the lattice vibrations in the crystal affect it but little, it is not very dependent on temperature. It is for this reason that dislocations play such an important part among lattice defects.

In the foregoing we have referred to motion within the slip plane of the dislocation. For a screw dislocation, every plane of movement can be considered as a slip plane. For dislocations not bearing the character of purely screw dislocations, however, we should also consider movements which do not meet this condition.

Dislocations with some edge-character are always associated, in the manner described for pure edge dislocations, with an extra half plane of atoms.

![Fig. 11 illustratng non-conservative movement of an edge dislocation (vide A. H. Cottrell, Dislocations and plastic flow in crystals. Oxford Univ. Press, 1953).](image)

Fig. 11 illustrates this situation diagrammatically. The plane of the drawing represents a cross-section perpendicular to the dislocation axis A. The extra half plane is represented by the line AB, and the Burgers vector lies in the slip plane of which the line CAD is the cross-section. With movement out of the slip plane, for example towards the point A', the length of the extra half plane changes from AB to A'B'. This means that a number of extra atoms have been removed (or added), and this can happen only if accompanied by a simultaneous transport of atoms to or from other parts of the crystal, or, if this is impossible, at the expense of considerable local distortion. Both processes necessitate activation energy (e.g. for diffusion); so also, therefore, does the movement of a dislocation beyond its slip plane.

Accordingly, whereas movement within the slip plane needs very little energy and is hardly dependent on temperature (conservative movement), movement outside the slip plane (non-conservative movement) demands very much more energy, which cannot in general be derived from the externally applied forces; furthermore, owing to the fact that an activation energy is involved, the latter is highly dependent on the temperature.

Clearly, a screw dislocation performs only conservative movements, since each crystallographic plane containing its axis can serve as slip plane. Needless to say, screw dislocations in crystals as well as other kinds of dislocation can be propagated only over crystallographic planes. Chalmers and Martius [13) have pointed out that it makes a difference which planes are involved. They showed that during its propagation, a dislocation will reveal a preference for planes containing the most closely packed layers of atoms. The slipping movement therefore occurs mainly along such planes.

Since the slip lines in a distorted crystal are the markings of the slip planes, this explains why, as mentioned above, the slip lines in many metal crystals run along the most closely packed crystal planes.

**Origin of dislocations**

Dislocations represent a considerable amount of energy by reason of the distortion in the crystal lattice that accompanies such lattice defects. The energy of a dislocation is defined by its stress field, the nature of which is rather intricate and would take us too far afield to consider here. It appears that the stresses decrease in inverse proportion to the distance from the dislocation axis, in consequence of which, in a continuous medium, the energy of a dislocation would be infinitely high. This is not the case with crystals because of their discrete atomic structure; it is found that the energy of a dislocation per atomic plane is several times \( Gb^3 \), i.e. in most metals, several times \( 10^{-12} \text{erg} \).

A proper dislocation is at least some tens of times the atomic spacing in length. The energy of such dislocations, that is, the energy required to produce them, is thus always \( 10^{-10} \text{erg} \) or more.

The presence of a dislocation increases the entropy of the crystal. It can be shorn that this effect is negligible compared to the energy of formation, which means that dislocations cannot possibly be initiated thermally, or exist in thermal equilibrium with the crystal lattice. In a metal crystal plastically elongated about 1%, we must nevertheless accept the occurrence of dislocations

with a total length in the crystal of at least $10^7 \text{cm}^2$ (i.e. a dislocation density of $10^7 \text{cm}^{-2}$), as demonstrated by experiments which will be discussed in the second part of this article.

This apparent contradiction has been solved by adopting the view that dislocations occur as a result of the actual process under consideration, viz. the plastic deformation. In principle this can be explained in the following manner.

If inhomogeneities occur somewhere within the crystal or at its surface, e.g. small cracks or inclusions, applying a stress will result in local stress concentrations, i.e. zones in which the stresses are much higher than elsewhere in the crystal. It is plausible that the formation of one or more dislocations in such zones might lead to a reduction in energy. If this is so, such zones will be capable of acting as sources of dislocations.

Which particular kind of inhomogeneities are likely to act best as sources, is not yet known with certainty. It is very probable that the most likely source of dislocations will be found at the crystal boundaries.

Dislocations may also be formed during the growth of the crystal. According to a theory put forward by Frank\(^8\), screw dislocations can very appreciably accelerate the growth of a crystal. This point has already been mentioned in the article referred to in footnote\(^1\), so that there is no need to dwell further on this very remarkable and interesting point here. So far as we are concerned here, we are interested only in the fact that a crystal grown in accordance with the mechanism described by Frank will invariably contain dislocations.

Although the two formative mechanisms described throw some light on the origin of certain dislocations distributed at random throughout the crystal, they by no means explain the observed phenomenon that, in a given slip plane, slip often occurs over a number of atomic spacings simultaneously, whereas in other crystallographic planes no slip takes place at all. This phenomenon can be due only to the movement of large numbers of dislocations all lying in the same slip plane. A possible way out of the difficulty is offered by the hypothesis of a dynamic multiplication mechanism for the dislocations initiated in one of the ways outlined above. A condition for the occurrence of this mechanism (a Frank-Read source, so named after the originators of this conception\(^14\)) is the presence of dislocation lines in the crystal, anchored at two points, so that they can deflect as a result of a shear stress without being able to move as a whole. Mott\(^15\) has shown the plausibility of the existence of such anchored dislocation elements in metallic crystals. He pointed out that the dislocations in non-distorted metals with cubic symmetry very probably form a space network, containing many nodes in which three dislocation lines, not located in the same plane, meet (fig. 12). If the orientation of the applied stress is such that a considerable force acts on one of the three converging dislocations, this usually means that the other two dislocations are subjected to only a slight force, or even to a force in the opposite direction. The nodes (e.g. A and B in fig. 12) will then constitute more or less fixed points in the dislocation network, and hence act as anchoring points for those dislocations which would otherwise tend to move under the influence of one force or another.

An anchored dislocation element (1) is illustrated in fig. 13. For convenience we will suppose that this

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\(^{15}\) N. F. Mott, Phil. Mag. 43, 1151, 1952.
is a purely edge-type dislocation. If a shear stress $\tau$ is applied, the dislocation will be deflected; it thus becomes longer, and the dislocation energy accumulated within it increases. Deflection is continued until a position is reached (e.g. position 2 in fig. 13) whereby the gain in potential energy by the dislocation in the stress field $\tau$ is balanced by this increase in dislocation energy. If the stress increases, position 3 will ultimately be reached, when the dislocation will have assumed the form of a semi-circle. This is a critical form. It can be shown that if the stress is increased gradually from this point, the energy required to produce further deflection of the dislocation (and hence to increase its length) is always less than the gain in potential energy that accompanies this further deflection. The stress required to produce the critical semi-circular configuration can be computed, being approximately:

$$\tau_0 = \frac{Gb}{l}, \quad \ldots \ldots \ldots (2)$$

where $b$ is the Burgers vector of the dislocation and $l$ the length of the original straight anchored element. If the critical value $\tau_0$ of the shear stress is exceeded, the dislocation expands ad lib without any further increase in the stress. Those parts of the semi-circle which are now practically perpendicular to the original edge dislocation now have the character of a screw dislocation (the Burgers vector, which remains constant, is at those points almost parallel to the dislocation axis). These screw-like parts of the dislocation thus move laterally outwards, perpendicular to the direction of the strain. Those parts which are practically parallel to the original dislocation, progress as an edge dislocation parallel to the direction of the strain. This being so, positions 4 and 5 are traversed, the strain being the same all the time, until position 6 is reached. Here, two parts of the screw of opposite sign face each other and, the movement being continued, will meet. They then cancel each other, since two dislocations of opposite sign coming together result once more in a perfect lattice. Finally, position 7 is assumed, this being none other than the condition at 1, plus a dislocation loop enclosed by the "source" and moving outwards under the influence of the applied stress.

In principle, an unlimited number of loops could emanate from the source, assuming that the applied stress is at least equal to the critical stress $\tau_0$.

This would provide a simple explanation for the observed translations of several hundreds of times the atomic spacing along a slip plane.

It is an obvious step to identify (in order of magnitude) the observed critical shear stress with the critical stress $\tau_0$. If this is done, it will be found from formula (2) that in most metals the length $l$ of the source in fig. 13 is several times $10^{-4}$ cm. Of course, longer sources will start working at lower stress levels than short ones, but it must apparently be accepted that the length of most sources is about $10^{-4}$ cm; this is of the same order of magnitude as the estimated average length of the network elements in Mott's model.

**Formation of vacancies and interstitial atoms**

If the energy of formation of a lattice imperfection and the change in entropy of the crystal associated with it are known, it is possible to estimate the thermal equilibrium concentration of such imperfections as a function of temperature. This concentration is proportional to the Boltzmann factor

$$e^{-\frac{U}{kT}},$$

where $U$ is the (free) energy of formation, $T$ the absolute temperature and $k$ Boltzmann's constant. At room temperature, the value of the product $kT$ is $4 \times 10^{-10}$ erg.

The formation energy of vacancies and interstitial atoms in metals is not known with any degree of accuracy. Huntingdon and Seitz have computed the energy for a vacancy in a metal such as copper to be roughly $10^{-12}$ erg. The formation energy of interstitial atoms will probably be several times higher. These values agree fairly well with the conclusions drawn from experimental work, and with their aid it can be computed that a concentration of vacancies or interstitial atoms of some appreciable magnitude can exist in equilibrium with the lattice only at temperatures near the melting point of a metal. If any indications are found of the formation of such lattice defects at normal temperatures, it must be concluded that their origin is not thermal. In fact, there are indications that during the process of deformation of a metal at ordinary temperatures, very large numbers of vacancies and interstitial atoms do occur, and we shall refer to this again in our next article.

The obvious course is to seek a mechanism that is responsible for the occurrence of lattice defects during deformation, i.e. during the formation and propagation of dislocations. Several mechanisms have been proposed, of which only one will be discussed here.

References

We have already seen above that the non-conservative movement of a dislocation (i.e. of an edge type of dislocation outside its slip plane) is necessarily accompanied by a variation in the length of the “extra half plane” of atoms involved in the dislocation. A number of extra atoms is withdrawn or added. If the temperature is high enough, or if the movement takes place very slowly, these atoms can be transferred to other parts of the crystal by diffusion. If these conditions are not fulfilled, however, a shortage, or surplus, of atoms occurs in the neighbourhood of the dislocation; in other words, there will be a number of vacancies or interstitial atoms. The question is, then: can non-conservative movements of dislocations occur during plastic deformation at low temperatures? Seitz and Mott \(^{17}\) have shown that such movements do indeed take place on a large scale.

Dislocations emanating from a Frank-Read source travel through the crystal, and it is inevitable that they encounter parts of the original and still existent network. The further movement, as a result of the applied stress, of a dislocation loop straight across the network element, results in the formation of discontinuities in both dislocation elements at the point where they intersect. Consider the simple case of two screw dislocations at right angles to each other (fig. 14). If dislocation 1 is intersected by dislocation 2, the relative translation within the crystal produced by the latter will result in a displacement of the two parts of dislocation 1 on both sides of dislocation 2. A “jog” is thus formed in dislocation 1 equal in length to the Burgers vector of 2, and in the same direction as this vector. Conversely, a jog occurs in dislocation 2 in accordance with the Burgers vector of dislocation 1. These jogs behave like small sections of dislocation, with the same Burgers vector as that of the original dislocation, but of course with a different orientation of axis. In this particular case, the jogs constitute small edge dislocations, since their Burgers vectors (those of the original screw dislocations) are perpendicular to their axes. Further movement of the screw dislocation is in effect non-conservative movement of these edge dislocations. The movement of the jogs would be conservative only if it took place in their own slip planes, that is, the planes through the jog and the axis of the screw dislocation. Actually, however, these jogs move together with the screw dislocation, i.e. they move in a plane perpendicular to that of the axis of the screw dislocation. Consequently, with further rapid movement as produced by the applied stress, the jogs leave behind them a row of vacancies or interstitial atoms.

Not only will the perpendicular intersection of screw dislocations almost always produce a jog with a non-conservative movement impressed on it: in general, the intersection of any two dislocations will also do this. In this way it is possible for very large numbers of vacancies and interstitial atoms to occur during plastic deformation.

We have now discussed both the formation and the principal characteristics of dislocations and other lattice defects in metals. In the next article, the effect of such lattice defects on the deformation and other related phenomena will be investigated.

**Summary.** The theoretical necessity for the existence of lattice imperfections in crystals, such as vacancies, interstitial atoms and dislocations, is clarified in the light of a number of phenomena related to the plastic deformation of metals, e.g. the occurrence of slip lines, work hardening and variations in the electrical conductivity. This article is devoted mainly to a discussion of the more important characteristics of dislocations, and deals successively with the geometrical aspects of various kinds of dislocation, their behaviour under shear stresses in the crystal as illustrated by means of a model, and the movement of dislocations, taking into account the difference between conservative and non-conservative movements. Lastly, the formation of lattice defects of different kinds as the outcome of interaction between existing defects is examined. The effects of these imperfections on certain mechanical properties will be dealt with in a subsequent article.

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