The Rutherford Memorial Lecture, 1962
Atomic physics and the strength of metals

By Sir Nevill Mott, F.R.S.

Delivered in:
Makarere University College, 23 October
Royal College, Nairobi, 2 November
University College of Rhodesia and Nyasaland, 6 November
University College of Ibadan, 19 November
Ahmadu Bello University, Zaria, 24 November
Kwame Nkrumah University of Science and Technology, Kumasi, 26 November
University of Ghana, Legon Hill, Accra, 29 November

(Received 3 April 1963)

In the previous Rutherford lecture Sir Lawrence Bragg (1961) gave an account of the history and present position of crystallography. Crystallography consists of the study of the regular ways in which atoms or molecules are arranged in solids, and crystallographers now investigate substances of ever-growing complexity. At the same time, alongside crystallography, there has grown up a science which is related to it, namely, a study of the ways in which these regular arrangements can break down. If there are many thousands of crystal structures known in nature, you would guess that there must be tens of thousands of possible defects; this may be so, but in fact the study of defects is confined at the present time to those which are observed in the simpler crystal structures, in metals, oxides and so on, and so we are far from knowing the full complexity that may exist. Actually, however, there is one form of defect which has a certain simplicity and very great theoretical and practical importance, namely, the dislocation. It is the history and present status of this concept that form the subject matter of this lecture.

There is a certain appropriateness in the dislocation as a subject for a Rutherford lecture. It is not only that one of the most important papers on the subject (G. I. Taylor 1934) came from Rutherford's Cavendish, or that the first photographs of moving dislocations were obtained in the same laboratory (and indeed in a room that he formerly used) 25 years later by P. B. Hirsch and his colleagues. It is also because of the similarities and differences between a dislocation and a particle. Rutherford was a man of particle physics; his α-particle, moving up to and repelled by the atomic nucleus, making itself visible by a flash on a zinc sulphide screen or by a track in an expansion chamber, is the most obviously particle-like of particles. Atomic physics was built on the concept of particles with forces between them. A dislocation is not a particle; it is a line. Nevertheless, like a particle in atomic physics, it can move about and be attracted or repelled by other dislocations or by impurities or point defects in the material. It is, as far as I
know, the only example in science of a mobile flexible line, with inertia and momentum, with many of the properties of atomic particles but also with this additional degree of freedom. It shares other properties with atomic particles; corresponding to any dislocation there can exist another dislocation of opposite sign, and when the two meet they can swallow each other up, nothing being left but energy, in this case in the form of heat. A theoretical analysis, too, shows the dislocation increasing its effective mass as its velocity approaches the speed of sound, just as that of a particle does as its velocity approaches the speed of light.

Dislocations were introduced into physics to account for the ductility and the mechanical strength of solids, particularly metals. I believe that the study of this subject illustrates very well the difference between pure and applied research and the importance of both. Here, too, may I refer to Rutherford. We are apt to think of science and particularly of physics as dealing with new and unexpected aspects of nature that our ancestors knew nothing about, electronic and nuclear phenomena, for instance. This was certainly the case with Rutherford’s work. No one suspected until the second half of the nineteenth century that radioactivity existed, let alone that nuclear energy could be released. But there are other much more familiar properties of matter which are just as proper subjects for investigation and may be just as difficult to understand. The fluidity of liquids is one and we do not really understand it yet. The ductility of metals, the subject of this lecture, is another.

Since metals were first used by man, it is as much as anything their ductility which had made them so useful. They can be forged into implements, and it is because of their ductility that they do not have the brittleness of flints and the earlier materials. Two properties of metals were discovered very early in the history of technology. One is that the hardness of most metals can be very greatly increased by any form of cold-working, such as hammering; this property is known as work-hardening. The other is that an alloy, bronze, formed by melting about ten per cent of tin with copper, is considerably harder than pure copper and therefore more suitable for tools and weapons. This is an example of the general principle that alloys usually have a greater resistance to deformation than pure metals.

The subject of this lecture is the attempt to discover, in terms of the movement of atoms, why these facts should be so. In these terms we can now in 1962 understand in some detail why certain alloys are hard, particularly age-hardened alloys. The same cannot be said of work-hardening for which, in spite of much experimental material and many attempts to interpret it, no universally accepted theory exists. But it is remarkable that, although these facts were known before the beginning of history, it is only very recently that any attempt has been made to find out why they are so. Before this was done mankind had radio, cathode-ray tubes and a highly developed electrical industry, and it is as much as anything the development of nuclear power stations, and the entirely unfamiliar conditions that metals have to put up with in reactors, that have stimulated systematic attempts by large laboratories to obtain this understanding.

But this neglect by the pure scientist did not of course prevent a very rapid
development of metallurgy in response to the needs of more conventional engineering. With modern techniques for the control of purity, temperature, crystal structure and so on, metallurgists were enabled to produce enormously harder, tougher and more heat-resistant metals than ever before. But if three decades ago physicists had been asked to say why the addition of a little of one metal improves the properties of another, they would have had to say that they were as ignorant of it as they were of why salt improves the taste of meat. They knew it was so but did not know why.

The history of metallurgy has been the opposite, then, of the history of the newer skills, such as electronics and nuclear energy, in which an understanding of what an electron or nucleus is and why it behaves as it does preceded the technology of how to use it. With metallurgy it was the other way round. Now, of course, things are changing. The engineers, and as I have said particularly the nuclear engineers, are making more and more stringent demands on materials, and that is one reason why it is necessary that theory should catch up with practice and the two should advance hand in hand.

But as so often happens, the first attempt to understand the ductility of metals, their hardening when deformed and the like, was made not in response to an industrial need but because of the inherent interest of a problem—namely, the softness of pure metals. This was discussed independently in three papers published in 1934 by Taylor, by Polanyi and by Orowan. The problem was, how do crystals deform and why do they deform so easily? It was clearly impossible that one plane of atoms should slide over another 'all at once'; if they did the strength would be much higher than observed. They therefore introduced the idea of a 'dislocation'—a sort of ruck in the way one atomic plane fits onto another. Figure 1, taken from Taylor's paper, shows what is meant. Taylor assumed that all deformation occurred through the movement of dislocations of this kind. The figure also shows the

![Diagram of dislocations](http://rspa.royalsocietypublishing.org/Downloaded from)

The Rutherford Memorial Lecture, 1962 151

---

**Figure 1.** Two-dimensional representation of an edge dislocation in a crystalline solid, showing the positive (a, b, c) and negative (d, e, f) forms, which can annihilate each other (from Taylor 1934).

---

This content has been extracted from a document and formatted for readability. The diagram is also included in the form as provided. For a full understanding, the context and the entire lecture might be necessary.
existence of the two kinds of dislocation, and it is easily seen that if one runs into the other mutual annihilation will result.

Given the concept of a dislocation, the all-important question is: do dislocations normally exist in solids? If they do the strength (flow stress) can be equated to the stress required to move them; if they do not, it must be equated to the stress required to create them. I shall come back to these points.

Taylor pointed out that dislocations have an elastic stress field round them, and so would attract or repel each other. He suggested that work-hardening was due to this interaction which would increase with strain if, as he assumed, the number of dislocations increased with strain. Whether the explanation of work-hardening is so simple now appears uncertain. But his idea, that mechanical strength may be due to local internal strains which hinder dislocations from moving, has proved very fruitful. It gives in principle the answer to the question which I posed earlier, why are alloys stronger (more resistant to shear) than pure metals? If the atoms of which the alloy is built up have different sizes, and there is little additional hardness unless they do, as shown in figure 2, the mere fact that they have different sizes will introduce local internal strains. This idea was taken up by Nabarro and myself in 1940 and again in 1948 in papers which attempted to account for the flow stress of alloys in this way, both for solid solutions and age-hardened materials. I shall attempt in this lecture to show what has happened to these ideas.

Another very important concept in the theory of dislocations was introduced by Cottrell (1948, 1953) and his co-workers. This is that impurities in a metal,
particularly carbon or nitrogen in iron, would tend to segregate along the dislocations where the lattice is expanded or strained. These impurities would lock the dislocations, so that a greater stress would be required to move them than would otherwise be necessary. On the other hand, as soon as a dislocation begins to move, it leaves behind it its locking atmosphere and so can move more easily. This leads to the phenomenon of the yield point. Recent researches have shown that a yield point can occur, either by dislocations breaking free or by the creation of new ones, whichever occurs for the lower stress. Very sharp yield points occur in metal whiskers of the kind which do not contain dislocations able to initiate slip (figure 3).

Before going further I must emphasize that a dislocation is a line extending across the crystal, not just a point. A dislocation, being a line, must either terminate at the ends of the crystal or be a closed loop, which separates the part of a slip plane which has slipped through a given distance in a given direction (the slip vector) from that which has not. An early paper by Burgers (1939) described two important kinds of dislocation, the edge dislocation with slip vector perpendicular to the line of the dislocation, and the screw with slip vector parallel to it. But in general a dislocation is curved; its tangent at any point may have any angle to the slip vector. The edge and screw are two important special cases.

The geometry of dislocations has been a subject in which theoretical investigation has been ahead of theory. Frank (cf. Mott 1951) was the first to point out that in many structures three dislocations can meet in a point and that a hexagonal chicken netting arrangement might be particularly stable. Heidenreich & Shockley (1948) predicted that in close-packed planes in the face-centred and hexagonal
close-packed structures a dislocation would split into two halves, separated by a stacking fault. This has frequently been observed by transmission electron microscopy. Lomer (1951) and Cottrell (1952) showed how two dislocations on different planes could run into each other and become immobile; this has been observed too.

An important early application of the dislocation concept was the realization that the mosaic boundaries existing in imperfect crystals, known to occur from X-ray analysis (see, for instance, Bragg & Bragg 1933), are simply rows or networks of dislocations. A row of dislocations has lower energy than a random arrangement, because when lined up in a row the compression (or other strain) due to one neutralizes the strain energy resulting from the expansion due to the next. This is most vividly shown by the bubble model (Bragg & Lomer 1949). In real crystals, given the chance, dislocations will line up in this way.

Although much was deduced about the properties of dislocations from geometrical considerations before they were actually observed, they have more recently been made visible and their real existence established by several techniques. It was first shown by Lacombe (1948) that the point where a dislocation meets the surface is susceptible to certain etching reagents, so that an etch pit can be formed there. Rows of dislocations show up particularly well in this way. The technique has recently been applied by Gilman and his co-workers at the General Electric Laboratories to the investigation of the movement and multiplication of dislocations in lithium fluoride, for which very detailed information has been obtained (Gilman & Johnston 1962). The next technique to be used for making dislocations visible is by ‘decoration’, a method only available for transparent materials. Almost any impurity in a crystal, if present in supersaturation, tends to segregate out along dislocations. This was first observed by Hedges & Mitchell (1953), who found that when silver bromide is exposed to light the silver formed photoptically separates out there. The silver is opaque, and the dislocations then show up as lines within the crystal, visible under the microscope. Other techniques, particularly in the hands of Amelinckx (1956), have involved heating the crystal in the presence of a metal which can dissolve interstitially, and then cooling so that the metal precipitates out along the dislocations.

The first method which made dislocations visible without decorating them was that of transmission electron microscopy, first developed by Hirsch and his colleagues in the Cavendish Laboratory. The method is in principle simple; the beam of electrons is passed through a thin foil of metal or some other substance, and any dislocation lines not parallel to the beam show up on the screen of the microscope as lines, the two ends of which are projections of the points where the dislocation meets the top and bottom of the foil. The arrangement is shown schematically in figure 4. Slip of the kind shown in the figure occurs in metal foils as a consequence of thermal stresses which result from heating by the electron beam, so that dislocations and arrays of dislocations can be seen in movement on the screen of the microscope, or in a ciné film taken of it. I remember vividly the day when Dr Hirsch and his colleagues arrived in my room in the laboratory and said ‘Prof.—come and see some moving dislocations’.
not sure that at that time we expected to be able to see dislocations with the electron microscope, though subsequent experimental and theoretical work has interpreted the observation in terms of the diffraction of the electron beam by the strained lattice round the dislocations.

The observations have shown, among other things, how easily a dislocation in a close-packed plane splits into two partial dislocations with a stacking fault between them. An example is shown in figure 5, plate 3, the stacking fault producing an interesting interference effect which gives the impression of shading. Photographs, particularly those with a ciné camera, show what a large effect the stacking fault energy has on the kind of slip which occurs. When the stacking-fault energy is low, as in stainless steel or alpha brass, the dislocations dissociate into two partials with a wide ribbon of stacking fault between them, and they are then seen to move forward with a jerky motion along an active slip plane, usually with a considerable number of dislocations in each plane. An illustration is provided in figure 6, plate 3. On the other hand, when the stacking-fault energy is high, as is the case particularly in aluminium, the phenomenon of ‘cross-slip’, the movement of a dislocation from one plane to another, occurs very frequently. An example is in figure 7, plate 4, where the trace left by a dislocation which has cross-slipped twice is easily visible.*

All these methods, as well as the new information that they provide, give conclusive evidence that the deformation of crystalline solids is in fact normally due to the movement of dislocations. We have now, also, a good deal of information on whether mechanical strength is related to the difficulty of forming dislocations or of moving these that are already there. Gilman & Johnston (1962) find that for lithium fluoride the dislocations already present are frozen-in so firmly by impurities that crystals initially deform by forming new ones at defects on the surface. Similarly iron may yield by the formation of new dislocations, particularly

* These traces fade after they are formed, and are thought to be due to strains originating in the oxide layer on the metal foil.
at low temperatures when those present already are securely locked. But in all materials after some yielding and in metals and alloys which do not show a yield point, the strength depends on the stress required to move a dislocation.

There have been many investigations of the factors which determine this stress. One has to ask first of all what this stress is for the motion of a dislocation in an otherwise perfect lattice (the so-called Peierls–Nabarro stress; for a recent review see Kuhlmann-Wilsdorf 1960). The evidence seems to suggest that this is negligible on close-packed planes of the face-centred cubic, but that for other structures a highly temperature-sensitive viscous resistance to movement exists.

In the time available, I will limit myself to some recent investigations for face-centred cubic metals of one example of resistance to flow. These are the experiments that have been made recently by Kelly and his colleagues in the department of Metallurgy in Cambridge on the origin of the strength of age-hardened alloys, of which aluminium with 4% of copper is perhaps the best-known example.

![Figure 8](http://rspa.royalsocietypublishing.org/) The critical resolved shear-stress of a copper single crystal containing 2 at. % of cobalt, quenched and aged at 600 °C as a function of the time of ageing (Kelly & Nicholson 1936).

It has been known for some time that these alloys, if quenched from a high enough temperature, are supersaturated solid solutions, but that during ageing precipitates of the dissolved metal or of an intermetallic compound begin to form. Initially they are often coherent with the surrounding matrix and are then known as Guinier–Preston (G.P.) zones, after their discoverers. On continued ageing, they increase in size, decrease in number and become incoherent with the surrounding lattice.

The property of these alloys that needs to be explained is that the resistance to flow first increases during ageing, and then decreases. This is shown in figure 8, in which is plotted the variation of critical resolved shear stress of single crystals of copper containing two per cent of cobalt, quenched and aged at 600 °C. The problem then is: why do a large number of small precipitates (or G.P. zones) harden the material, whereas a small number of large precipitates do not have as great an effect?

There is nothing surprising in the idea that obstacles, particularly of some hard material, could hold a dislocation up. The method of transmission electron microscopy has provided many examples of this. One of the clearest examples is shown in figure 9, plate 4, obtained with magnesium oxide—though here the obstacle is probably an irregularity in the dislocation itself, of the kind known as
Figure 5. Showing stacking faults in a material of low stacking-fault energy (Cu + 7\% by weight of Al, single crystal); electron micrograph by Howie (1961). (Magn. × 55000.)

Figure 6. Dislocations in the same alloy as figure 5, showing a considerable number on each slip plane, with little if any cross-slip. Deformation in stage II of work-hardening (Howie 1961). (Magn. × 32000.)
**Figure 7.** Showing trace left by moving dislocations in aluminium, heavily cold-worked with trace of multiple cross-slip. A network is also shown (Whelan 1958). (Magn. x 33000.)

**Figure 9.** Cusp-like formation of dislocations in MgO (Washburn et al. 1960).
**Figure 10.** Dislocations in a CuCo crystal aged to its maximum strength (Bonar & Kelly 1963).

**Figure 13.** Distribution of dislocations in an over-aged material that has been deformed (Bonar & Kelly 1963).
a jog. The question at issue for the age-hardened materials is how does a dislocation get through the obstacles and what stress is required to push it through?

According to the recent analysis of Kelly and his colleagues in the Department of Metallurgy at Cambridge, in the hard materials the dislocations go right through the G.P. zones. The resistance may be partly a consequence of strains due to coherency—as postulated in the early work by Mott & Nabarro (1940, 1948), but is more likely to be caused simply by the energy required to introduce a dislocation into the material of the precipitate when it has a higher shear modulus (e.g. CuAl₂ in Al, Co in Cu). The movement of dislocations in materials aged to

<table>
<thead>
<tr>
<th>Sample</th>
<th>Age</th>
<th>Temperature</th>
<th>Flow Stress (Kg/mm²)</th>
<th>Work-Hardening %</th>
</tr>
</thead>
<tbody>
<tr>
<td>C1</td>
<td>2h</td>
<td>190 °C</td>
<td>25.94</td>
<td>33.56 %</td>
</tr>
<tr>
<td>C12</td>
<td>30h</td>
<td>190 °C</td>
<td>28.23</td>
<td>31.75 %</td>
</tr>
<tr>
<td>D2</td>
<td>100h</td>
<td>190 °C</td>
<td></td>
<td></td>
</tr>
<tr>
<td>D10</td>
<td>300h</td>
<td>190 °C</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Figure 11. Work-hardening of Al 1.6 at. % Cu single crystals after ageing at 190 °C for the times shown, aged and over-aged (Kelly 1962). All specimens before ageing were solution-treated for 15 h at 545 °C and water-quenched.

near their maximum hardness thus behaves as though an additional frictional force had been introduced. This force is temperature-dependent because the G.P. zones are so small that temperature fluctuations are large enough to help a dislocation to get through them. Also the work-hardening proceeds at the same rate as in the unhardened material, because nothing is changed in the process of work-hardening except that an extra stress must be applied to force the dislocations through the precipitates. Figure 10, plate 5, shows some dislocations in an alloy at its maximum strength; the precipitates are too densely distributed for cusps on the dislocations to be visible. Figures 11 and 12 show the work-hardening curves and the temperature-dependence of the flow stress, contrasted with these in the over-aged alloys.

In the over-aged alloys after some deformation, the dislocations appear as in figure 13, plate 5; clusters of dislocations round the precipitates are observed. It is believed that for these materials the flow stress is the stress required to force dislocations between the obstacles. The successive forms of the dislocation will be as in figure 14 (c), so that each dislocation line which passes an obstacle leaves
a loop behind. These are the sources of the clusters of dislocations which we see round each obstacle in figure 13.

In straining an over-aged material, the model based on these photographs leads one to expect

(a) that the flow stress $S$ (or rather the ratio $S/G$, where $G$ is the shear modulus) should be independent of the temperature, because no interaction with obstacles of atomic size is involved;

(b) there should be rapid work-hardening, since during plastic flow the material is filled up with these dislocation loops;

(c) the flow stress will drop as the distance between the obstacles increases (figure 15).

All these predictions are confirmed by experiment, as figures 11 and 12 and 15 show.

I have been able to show you, I hope, a part of this subject in which mechanical strength is fully or almost fully understood in terms of the movements of dislocations, and thus in terms of the movements and properties of the atoms of which the alloy is made up. I would not like you to think that the same detailed understanding exists in the whole field. In particular, we have no universally agreed theory of work-hardening. The method of transmission electron microscopy shows us clearly that, as the strain in a material increases, so does the density of dislocations within it, though how they are arranged certainly varies from one substance to another. Almost any description of slip in terms of the movement of dislocations will predict that the density of dislocations will increase with strain, essentially because dislocations are rarely straight and a curved dislocation will bow out into a loop; the special mechanisms of Frank and Read, by which a number of dislocation loops are produced in one plane, is an example of this. Moreover, it has been
clear since Taylor's paper of 1934 that dislocations can interfere with each others' motion, and that a greater stress is required to move a dislocation through a material containing other dislocations than through a material which is free of them. What is not clear is how far each dislocation moves before it is stopped,

![Diagram A](image1)

![Diagram B](image2)

![Diagram C](image3)

**Figure 14.** Schematic behaviour of a dislocation in aged material (A, B) and over-aged material (C). The last diagram shows how dislocations can leave loops behind, leading to rapid work-hardening in over-aged materials (from Cuff & Schetky 1955).

![Graph](image4)

**Figure 15.** Dependence of the resolved shear stress on distance between precipitates, for Al 1.7 at. % Cu. Aged at 190 °C to contain θ' precipitates of various sizes (Bonar & Kelly 1963).

what stops it and how to predict the kind of stress–strain curve that is observed in practice. Even less do we have a clear understanding why a high density of dislocations leads to fracture in fatigue tests.

I hope then that I have shown you in this lecture that the application to one of the oldest problems of technology of the concepts of atomic physics and the advanced techniques of electron microscopy has produced a branch of science
which bids fair to be very useful and which at the same time has some of the
elegance and fascination of particle physics.

I should like to thank Dr P. B. Hirsch and Dr A. Kelly, both of whom have pro-
vided me with photographs and have shown me some of their recent results
before publication.

References

Amelinckx, S. 1956 Phil Mag. 8, 269.
Bonar, L. G. & Kelly, A. 1963 (Private communication.)
Brenner, S. S. 1958 Growth and perfection of crystals (ed. by Doremus, R. H., Roberts, B. W.
Cottrell, A. H. 1952 Phil. Mag. 43, 645.
London: Physical Society.
Kelly, A. 1962 Theories of precipitation and dispersion hardening, Symposium on Electron
Microscopy and Strength of Crystals. J. Wiley.
Lomer, W. M. 1951 Phil. Mag. 42, 1327.
Physical Society.
Orowan, E. 1934 Z. Phys. 89, 605.
Polanyi, M. 1934 Z. Phys. 89, 660.