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Dislocation relaxation*

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Abstract

An analysis of the literature on dislocation relaxation yields some promising subjects for further investigation together with two criteria which may be profitably followed when planning experiments.

1. Introduction

The first chapter of the book *Dislocations in Crystals* by Read [1] has the title "Why discussions?". This question was very appropriate in 1953 when the book was published, because dislocations, after their "birth" due to Orowan [2] and Polanyi [3] in 1934, were just coming out of minority and were therefore still little known, conserving a sort of "hypothetical" life. Moreover, the relaxation effect which indicates their existence and discloses some aspects of their motion had been first observed only in 1949 [4] and consequently was hardly taking its first steps.

Since the publication of Read's book, much has been done and written about this effect, as evidenced by the review of Niblett [5], Chambers [6], Nowick and Berry [7], De Batist [8] and Fantozzi *et al.* [9]. Presently, the amount of experimental and theoretical information available on it is so large that, when reading the title of the present paper, one may be tempted to ask "Why dislocation relaxation again?". An attempt to give a satisfactory answer to this question is made in the following sections, by indicating two general criteria which have often been adopted, without an explicit formulation, in past experimental investigations and have proved as essential as they appear for present ones in order to obtain fundamental information on dislocation dynamics. A list is successively given of some aspects of dislocation relaxation which appear worthy of further investigation and may be the themes of promising research.

2. Structural simplicity and constancy

When the experiments which showed the existence of dislocation relaxation [4] were planned, it was reasonably assumed that the interpretation of their future results would be particularly easy and physically significant if the measurements were made on solids whose structures were (1) chemically and physically simple and (2) constant during the measurements.

Following the criterion (1) of *structural simplicity*, the choice of materials fell on chemical elements, as pure as possible, with the same crystal lattice to facilitate the comparison of results.

It was already known [7, 8] from the earlier work of Koester [10] that a sudden plastic deformation increases immediately the dissipation of elastic energy which takes place in a solid, indicating a corresponding increase in geometrical imperfection content. Successively, if the temperature is not varied, the dissipation decreases gradually in time, approaching asymptotically a constant final value. This indicates that the structure of the deformed solid, under the action of thermal agitation, evolves in an irreversible way towards states of lesser energy dissipation, tending to a final state of statistical equilibrium. When the dissipation is near its final value, it may be further reduced by increasing the temperature and consequently by altering the statistical equilibrium structure. In contrast, no change in dissipation or structure takes place when the temperature is reduced.

On the basis of the above results and the criterion (2) of *structural constancy*, during the original experiments every plastic deformation was immediately followed by a treatment of structural stabilization, in which the solid was kept at a constant temperature not lower than any of those which were to be investigated. This treatment lasted until the dissipation appeared constant in time, within the limits of experimental accuracy, indicating that the final equilibrium structure had been sufficiently approached.

The convenience of following the above criteria appeared already clear from the original experiments and

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has been fully confirmed by the considerable amount of experimental work reviewed in refs. 5–9. Owing to the structural stability, when the dissipation coefficient was measured at different temperatures, the results repeated themselves with great accuracy in successive experimental runs, made either by cooling or by heating the solid. Consequently, it was possible to detect with certainty relaxation peaks not higher than 10^{-5}, which is a rather small value even for present experimental facilities. Moreover, it was comparatively easy, owing to the simple structures of the materials chosen, to give a structural interpretation of the relaxation effect observed, by ascribing it to the motion of straight dislocations in the neighbourhood of their equilibrium positions [4], according to a model first outlined by Mason [11] and successively improved and refined by Seeger [12], who later gave a number of essential contributions to dislocation dynamics. According to this interpretation, the effect has an *intrinsic* character, because it is due only to dislocations, and would be present also in an ideal solid which contained no other imperfections. Moreover, the same effect appears as characteristic of the simple lattices in which it has been observed rather than of any particular species of atoms.

For present work it is important to observe that, when the dislocation relaxation has been investigated in several different solids with simple and constant atomic structures – as is actually the case – the experiments can be profitably extended to more complex ones, interpreting their results with the help of the simpler ones obtained preliminarily. Similar extensions have already been made in the past, for instance when the investigations on copper, silver and gold were successfully followed by similar ones on some of their alloys [8, 13, 14].

With regard to the opportunity of measuring dissipation *immediately* after plastic deformation, when the structure of the solid is still in full evolution, the experimental evidence available shows that the dissipation–temperature curves determined in this way generally show a larger number of dissipation peaks than the analogous curves obtained when the atomic structures, after a stabilization treatment, have reached a statistical equilibrium. However, this abundance of peaks does not necessarily correspond to an increase in our knowledge of the solid state, because some of the peaks observed, in particular those which disappear at higher temperatures, may be due not to relaxation effects but to *conflicting changes in relaxation time and strength*, owing to their dependence – physically not very significant – on both temperature and *time*. It may be added that even the *relaxation* peaks eventually present cannot give reliable quantitative information on the effects to which they are due, owing to the unknown variations in relaxation time and strength which take place during the experiments.

3. Relaxation spectrum and subsidiary peak

The first mathematical model adopted for dislocation relaxation was the simple one introduced by Zener [15] for thermally activated effects, with a single activation enthalpy and a single high temperature limit of the characteristic time. According to this model, the enthalpy can be easily computed from the variation observed in the dissipation peak temperature when the vibration frequency of the solid is changed. From the value obtained in this way it is also possible to compute the width of the dissipation–temperature curve corresponding to the same model.

When the above computations are made, it is always found that the experimental dissipation–temperature curve due to dislocation relaxation are two or three times wider than those which correspond to the simple Zener model. This proves the existence of a *relaxation spectrum** which, owing to the thermally activated character of the effect, may vary between two extreme types: a spectrum of different limit times, all associated with the same enthalpy (*time spectrum*), and a spectrum of different enthalpies, all associated with a common limit time (*enthalpy spectrum*).

Many attempts have been made to decide which type of spectrum better approaches the experimental results obtained for dislocation relaxation, but no satisfactory conclusion has yet been reached. This point therefore appears to deserve further investigation owing to the important information on dislocation content which could be derived from the type of their relaxation spectrum.

In connection with the shape of the dissipation–temperature curve, it must also be observed that, when the relaxation effect due to dislocations has an intrinsic character, the main peak is accompanied, on the low temperature side, by a less evident subsidiary peak, first detected by Niblett and Wilks [5, 7–9, 16]. It has been proposed to explain the simultaneous presence of two relaxation peaks by ascribing them to different types of dislocations whose identification is, however, still controversial [17–19]. Moreover, no attempt appears to have been made to justify the complete absence of the subsidiary peak in relaxation effects due

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*As is done in Zener's book [15] and all substantial contributions to dislocation relaxation [5–9] the term "spectrum" is employed here in its proper meaning suggested by the analogy with electromagnetic and in particular optical phenomena. The use of the same term to denote the dissipation–temperature curves may generate a confusion which does not contribute to a satisfactory understanding of relaxation effects.

to the interaction of dislocations with point imperfections [19], as will be discussed in more detail in the next section. The structural origin of the Niblett and Wilks peak is therefore another point which deserves further investigation.

4. Intrinsic and interaction effects in palladium and b.c.c. solids

The relaxation effect first observed in plastically deformed palladium [20] was similar to those previously or simultaneously found in other deformed f.c.c. solids [4–9, 13, 16, 20, 21] and was structurally interpreted in the same way, as an intrinsic effect. However, palladium, differently from other f.c.c. solids, can accommodate considerable amounts of interstitial hydrogen. Making use of this property, it was soon found that, after a plastic deformation combined with the introduction of hydrogen, a relaxation peak was observed which was clearly due to a hydrogen–dislocation interaction owing to its dependence on the contents of both these imperfections [22]. It was then suspected that even the first peaks observed in the same material [20] could not have an intrinsic character but were rather due to an interaction with some *residual* hydrogen. The doubt was not well founded, because the dissipation–temperature curves obtained after plastic deformation *alone* were qualitatively different from those which corresponded to an interaction effect. In particular, the former exhibited clearly a subsidiary Niblett and Wilks peak, which was completely absent in the latter [22]. Moreover, the dissipation observed after plastic deformation alone could not be due to an interaction with residual hydrogen, because, at low temperatures, the same dissipation *decreased* instead of increasing after the introduction of further hydrogen. These differences, together with other less evident but substantial ones, were later evidenced [19] and the controversy on the character of dislocation relaxation in palladium was settled in a reasonable way by showing that this character is intrinsic for a small residual hydrogen content, as in the first measurements [20], while the interaction predominates gradually when further hydrogen is introduced [22].

The behaviour of b.c.c. solids with respect to interstitial impurities is similar to that of palladium, but, differently from this metal, the structural interpretation of the relaxation effects observed in a considerable number of experiments [6–9, 23, 24] is still controversial. Further investigations are therefore required in order to separate eventual intrinsic effects from interaction ones and to extend to non-compact lattices our knowledge of dislocation dynamics.

5. Hexagonal lattices

Several attempts have been made to find whether plastic deformation of hexagonal solids can produce a dislocation relaxation effect as in the f.c.c. materials. It was preliminarily observed [25] that temperature changes of hexagonal polycrystals give rise to large intergrain stress due to the anisotropy of thermal expansion. This stress generally exceeds the plastic limit, increasing the dislocation content of crystal grains. Consequently, in order to follow the criterion of structural constancy, the dislocation relaxation in hexagonal solids must be investigated by measuring the dissipation as a function of temperature in plastically deformed *single crystals.*

Such investigations have so far been made on zinc [25] and magnesium [7, 26] and have shown that an intrinsic effect due to dislocation relaxation is also present in hexagonal crystals.

For other hexagonal solids (zirconium, titanium, rhenium) the experiments have been done only on polycrystals and the results are less clear, as would be expected owing to the structural changes which take place during the measurements. However, rather satisfactory results have recently been obtained on cobalt and cadmium polycrystals [27], in which the relaxation effect appears very similar to that due to the intrinsic dislocation motion in the f.c.c. lattice.

The comparison between the respective dislocation behaviour in f.c.c. and hexagonal crystals has considerable interest, because the fundamental difference between these two structures lies in the stacking sequence of their close-packed planes. Limiting to single crystals the experiments on dislocation relaxation, for the reasons indicated above, it appears therefore desirable to extend them to other hexagonal elements.

6. Conclusions

Following the two criteria of structural simplicity and constancy, further basic information on dislocation dynamics can be obtained by investigating the time or enthalpy spectra of their relaxation effects; the experiments may also be profitably extended to b.c.c. solids and hexagonal single crystals.

In some cases the simplicity criterion may be neglected, extending the investigations to complex structures and interpreting their results with the help of presently available information on simple solids.

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