Internal Friction Studies of Substructure*

A. V. Granato
University of Illinois, Urbana, Illinois

Abstract

A review is given of the type of information concerning mechanical properties of crystals obtainable from internal friction measurements. The methods of measurement, together with their advantages, disadvantages and limitations are outlined. Results are described within a framework in which defect interactions are considered from the "particle" point of view. A selection of results found by internal friction techniques which are often difficult to obtain by other methods is described. These include recent measurements which have contributed to the understanding of dislocation interactions with point defects and phonons, as well as dislocation distributions in deformed materials.

I. Introduction

A. Concerning the Definition of Substructure. If by substructure, one means structure having to do with small angle boundaries, then it must be admitted at the outset that internal friction methods have so far provided us with very little information about this aspect of mechanical behavior. On the other hand, if the definition of substructure is broadened to include other defects, as has been suggested, then we find that internal friction methods supply us with much information of a type which is often difficult to acquire by other methods. For the purposes of the present article, we shall regard grain boundaries as the primary structure, and all other defects including single dislocations, dislocations in interaction with other dislocations to form small angle boundaries, point defects, etc., as composing the substructure.

B. Mobility of Dislocations. The extent to which internal friction techniques can be used to study dislocations is the extent to which the dislocations are mobile. In a solid containing no defects, ultrasonic plane waves should propagate without attenuation. A plane wave traveling in a solid containing dislocations is attenuated, and the attenuation depends sensitively upon the mobility

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of the dislocations. The mobility, in turn, depends upon the interactions of dislocations with point defects, phonons and other dislocations. Thus, measurements of internal friction, or ultrasonic wave propagation in crystals, are uniquely suited for the study of dislocation interactions with other defects. On the other hand, if the dislocation interactions are such as to render the dislocations immobile, then no internal friction is obtained. Thus the technique is limited to the study of mobile dislocations.

C. Some General Features of Internal Friction Measurements. Advantages of internal friction measurements derive from (1) the sensitivity of the method, (2) the selectivity of the measurements, (3) the fact that the results can be made quantitative, and (4) the fact that the measurements are non-destructive. For example, defect concentrations of the order of 10^{11} cm^{-3} can have an important effect on the internal friction. Also, the internal friction is sensitive only to those point defects which arrive at dislocations, in contrast to properties such as resistivity, which are sensitive to all the defects in the lattice. The method is especially suited to the study of defect interactions, since it is only by these interactions that changes in the ultrasonic wave propagation characteristics are induced. A large amount of detail can be found because of the large number of variables which can be controlled (frequency, strain amplitude, temperature, point defect concentration, ultrasonic mode orientation, purity, deformation, and others). As a disadvantage, a detailed theory or model of dislocation interactions is required to interpret the results. The latter difficulty is one which has held up progress in the use of this technique until quite recently.

Measurements of internal friction tend to complement, and not to compete with measurements made by direct observations and also with other indirect observations. Internal friction measurements differ from other indirect observations in sensitivity and in the linearity of the effects with defect concentrations. For example, properties such as resistivity, lattice parameter, density and stored energy are usually linear in the defect concentration, and this fact leads to simplicity in interpretation of results. However, it is normally necessary to have defect concentrations considerably in excess of parts per million to obtain measurable effects. On the other hand, internal friction measurements require more interpretation (the internal friction depends on the fourth power of the pinning point density in common cases), but fewer defects are required for measurable effects and more detail is obtained in the measurements.

D. Methods of Measurement. A great deal of work so far has been done in the kilocycle and megacyle range, although some work has also been carried out at lower frequencies. The techniques used have been well described in many places.[1] We shall only mention here some of those features of importance in work on the study of defects in crystals. In the 10-40 kilocycle range the technique usually used is to set up a standing wave in a long thin specimen. A typical size of specimen might be about a quarter inch in diameter and two inches in length. Harmonics of the fundamental vibration may also be used but there are difficulties here because different parts of the specimen are excited for different harmonics. The specimen may be driven into resonance by an eddy current drive or by a quartz rod attached to the specimen. The advantage of the former method of coupling is that measurements as a function of temperature are easily made. Even in the latter case, the quartz may be bonded to the specimen at a strain node so that deformation arising from differential expansion at the specimen-quartz interface during temperature changes does not disturb the measurement greatly. The strain amplitude may be varied from below 10^{-8} to above 10^{-5}, depending upon the specimen damping.

Measurements in the 3-300 mc/sec range are made using the pulse technique. Specimens are often in the shape of flat cylindrical disks with typical dimensions of half inch in thickness and one inch in diameter. A thin quartz transducer disk, of diameter half inch or less, is attached by means of an extremely thin bonding layer. The transducer is used both as a transmitter and receiver of ultrasonic pulses. After each round trip in the specimen the pulse is detected and all the echoes are displayed on an oscilloscope, from which the attenuation is measured. Measurements as a function of temperature offer difficulties here, since the differential expansion at the specimen-quartz interface introduces deformation in soft crystals. However, frequency measurements are easier to make here than in the kilocycle region. All odd harmonics of the fundamental driving frequency of the quartz can be used until the attenuation becomes too large to measure. Also several modes of propagation (longitudinal and shear) can usually be used, permitting a study of orientation effects. Normally the strain amplitudes available are low, of the order of 10^{-7} or less. The major difficulty in specimen preparation here is in the requirement that the two reflecting faces be accurately parallel. For measurements above about 50 mc/sec, optical tolerances are often required. It is possible to mount the transducer on the end face of a tensile specimen, so that measurements of attenuation and velocity can be made during deformation.[2] In summary, measurements as a function of temperature and strain amplitude are easy in the kilocycle region and difficult in the megacycle region, but measurements as a function of deformation, frequency and orientation are easy in the megacycle region and difficult in the kilocycle region.

In what follows, we give first a discussion of the model used in interpreting internal friction effects, then a framework in terms of defect interactions in which the experiments are to be discussed, and finally, a discussion of a limited selection of experiments as examples of the type of information concerning mechanical properties obtainable from internal friction experiments.

II. The Model

A. Types of Effects. There are a number of ways in which dislocations can contribute to internal friction. A dislocation segment oscillating between two pinning points (vibrating string model) gives one characteristic type of damping. Dislocations which break away from pinning points under stress lead to another type. Dislocations which move by overcoming Peierls barriers are supposed to give rise to the low temperature Bordoni peaks. There are also other mechanisms, models for which have not yet been developed. In particular, there is not as yet a suitable model for the highly deformed state. However, the single dislocation segment model which neglects interactions between dislocations seems to work surprisingly well for moderate deformation (up to about 4%). In what follows we discuss only the vibrating string model since its predictions are definite and many experimental checks of these predictions are now available. We are interested in answering first basic questions such as "What is the physical source of the damping?" and "What is a pinning point?". Then having settled this, we look to see how internal friction measurements can be applied to the study of mechanical properties of crystals.

B. Dislocation Contribution to the Total Strain. The basis for internal friction effects lies in the fact that dislocation motion contributes to the total strain developed in a specimen under stress. For a given applied stress, a solid containing dislocations has a larger strain than a perfect crystal, so that the elastic modulus appears to be lower. Under the action of an alternating stress, the dislocation component of the strain may lag behind the applied stress. This leads then not only to a reduction in modulus but also to a damping of the applied stress. Simple estimates of the magnitude of the expected effect for typical dislocation densities lead to much larger values than the observed effects if it is assumed that the dislocations are perfectly mobile with no restrictions on their motion. It may therefore be concluded that there must be impediments to the motion of dislocations. Generally speaking, the same types of obstacles have been assumed as those assumed in yield stress theories, where a similar problem is faced. These are, for example: atomic pinning points, network points, jogs, other dislocations, etc. But even with such restrictions, one must explain why the dislocation motion lags behind the applied stress. For smooth dislocation motions, one may imagine that the dislocation is viscously damped as it moves through the electron or phonon gas. Also impediments which lead to a jerky motion of the dislocation will lead to a phase lag. Examples of the latter type of effect are provided by motion over Peierls barriers at low stresses, and catastrophic unpinning of dislocations at high stresses.

C. Vibrating String Model. A model, proposed by Koehler,[3/] and developed further by Granato and Lücke,[4/] using an analogy between a vibrating string in a viscous medium and dislocation oscillations has been proven to be correct in most of its particulars. In this model, advantage is taken of the fact that a dislocation has an effective mass per unit length and an effective tension. Thus an equation of motion for small oscillations of a dislocation may be written as

$$A\,Y_{tt} + B\,Y_{t} - C\,Y_{xx} = b\sigma \qquad (1)$$

where A is the effective mass for unit length, Y is the dislocation displacement measured from the equilibrium position as indicated in Fig. 1, B is the viscous damping constant, C is the tension, b the Burgers vector, σ the applied stress, t the time, x a coordinate along the dislocation, and subscripts denote differention. The solution of Eq. (1) together with the boundary conditions $Y(o)$ and $Y(\ell) = 0$, where pinning points are placed at $x = 0$ and ℓ gives the dislocation displacement Y as a function of the frequency ω of the applied stress $\sigma(\omega)$. Using this the dislocation strain and then the effective modulus (or ultrasonic wave velocity) and decrement (or ultrasonic wave attenuation) is easily found. The resulting decrement has a typical resonance type frequency behavior, which depends, however, on the magnitude of the viscous damping constant B. Theoretical estimates of the damping constant to be expected for dislocation interactions with phonons[5/] and electrons[6/] have been made by Leibfried and Eshelby. From these one expects

(1) that the phonon interaction is much larger than the electron interaction, and (2) that the phonon interaction is so large that the dislocation resonance is overdamped. This has the effect of broadening out the resonance and moving it to lower frequencies. Also when account is taken of the fact that not all dislocations have the same length, the expected maximum is broadened out further. For typical expected dislocation segments of length of order 1 micron, the expected resonance frequency is at about 10^9 cycles/sec. However, because of the large damping (we may picture the dislocation as a string moving in heavy molasses), the maximum may be brought down to frequencies as low as 100 kc/sec. At the same frequencies where the decrement goes through a maximum a dispersion in the elastic constant is expected. As is indicated schematically in Fig. 1, the displacement of the dislocation is a function of frequency. (The dislocation strain, and thus the modulus reduction is proportional to the area swept out by the dislocation.) At low frequencies, the velocity of the dislocation is small so that the viscous force is small. The displacement is then limited by the tension forces and is parabolic in shape. At high enough frequencies, however, the viscous forces become dominant, and the displacement of the dislocation cannot achieve its full value. In this case the dislocation moves more like a rigid rod over most of its length, coming down to zero displacement only near the pinning points. Thus we expect the effect of pinning points to be large at low frequencies and negligible at high frequencies.

The dispersion effect is illustrated in Fig. 2 in which the velocity of compressional waves was measured by Granato, de Klerk, and Truell[7] in a sodium chloride crystal as a function of frequency before and after a slight deformation. Before deformation there is only a small dispersion of 0.5% centered at about 75 mc/sec. After the deformation, the magnitude of the dispersion has increased to about 4% and has moved to a lower frequency (35 mc/sec). At room temperature, the dispersion was observed to gradually recover towards the initial condition, presumably as a result of dislocation pinning by deformation induced defects. The interpretation of the effect according to the vibrating string model is as follows. At low frequencies, the dislocations are in phase with the ultrasonic stress. When the stress is applied, the apparent elastic constant (and therefore the ultrasonic velocity) is reduced because the dislocation motion makes the specimen less rigid. However, at high frequencies the dislocations can no longer follow the rapidly changing stress, so that the modulus approaches the true elastic value.

The frequency dependence of the decrement arising from dislocation motion is illustrated in Fig. 3. These are measurements by Stern and Granato[8] showing the effect of cobalt gamma irradiation on the decrement of high purity copper. Before irradiation, the decrement has a maximum at a few megacycles. The gamma rays produce electrons which are energetic enough to displace lattice atoms, giving interstitials which can be effective as pinning points. After 50 hours of irradiation in a 6000 c cobalt source, the height of the maximum decreased and the location increased to about 100 mc/sec. The decrement at low frequencies is much more sensitive to the increased number of pinning points than that at high frequencies, as we expected from our previous discussion of Fig. 1.

According to the theory, the height of the maximum should be proportional to ΛL^2 and the location of the maximum proportional to $1/BL^2$ where Λ is the total dislocation density, and L is the average loop length. At frequencies much lower than that at which the maximum occurs, the decrement should be proportional to $\Lambda L^4 B\omega$ and the modulus to ΛL^2. The predicted dependence on loop length has been confirmed by Thompson and Holmes.[9] Thus measurements of the height and location of the decrement in the megacycle range give the same information as measurements of the decrement and modulus in the kilocycle range. Actually, the two quantities which can be determined from the measurements are the ratios Λ/B and $L/\sqrt{c}$.[8] Before dislocation densities and loop lengths can be determined, the damping constant B and tension c must be known. We shall discuss means by which the damping constant can be determined in a later section.

III. A Formal Framework In Terms of Defect Interactions.

Because the internal friction effects depend entirely upon the interactions of dislocations with other defects, we shall find it convenient to classify the effects in terms of defect interactions using the "particle" point of view introduced by Seitz.[10] The defects to be considered are dislocations, point defects and phonons. This omits but two of the six primary defects in crystals: electrons and excitons. (Foreign atoms have been lumped into the point defect category together with vacancies and interstitials.) We consider first dislocation interactions between pairs of defects.

A. Dislocation-phonon Interations. We first note that phonons interact with dislocations in essentially two ways. Just as in the case of Brownian motion of a particle in an external field, a dislocation moving through a phonon gas under the action of an external applied stress is subject to both a viscous drag and to fluctuations in displacement. The latter effect is thought to give rise to the Bordoni peak, but will not be discussed here.

B. Dislocation-point Defect Interaction. The best known interaction of this type is that first discussed by Cottrell.[11] Because an oversized (or undersized) impurity can relieve the strain energy of the lattice by moving to the dilated (or compressed) region near a dislocation, the dislocation will be bound to the impurity. The binding strength will be small, at most of the order of a few tenths of an electron volt, so that a sufficiently large stress can pull the dislocation away from immobile point defects. Interstitial atoms and vacancies can also act as pinning points. Other possibilities are jogs, dislocation nodes, intersections and places where a dislocation may move out of the slip plane.

C. Dislocation-dislocation Interactions. Dislocations may interact with other dislocations at a distance through their long range stress fields, and also (more strongly) at points of contact. Presumably these effects should become important in deformed materials where the dislocation density is high. A difficulty here is that the predictions of any theory for these effects depend sensitively on the model assumed, but so far no simple model has been established as being representative. Presumably small angle boundaries and pile-ups which are mobile should lead to internal friction with certain special characteristics. Calculations for these configurations have not yet been attempted. Perhaps as a result of the direct observations reported at this conference, we shall learn which of the possible arrangements should be taken most seriously as models for damping in heavily deformed materials. Generally speaking, the measurements show that for high enough deformation, the damping decreases. This shows that the effect of dislocation interactions in inhibiting dislocation mobility more than compensates for the increased dislocation density.

D. Point Defect-Phonon Interactions. By using dislocations as an intermediary defect, diffusion effects can be studied. For example, by measuring the rate at which dislocations are pinned as a function of temperature, the activation energy of migration of defects can be measured.

E. Dislocation-phonon-point Defect Interactions. The combined effect of triple interactions between these basic defects shows up in a striking way in measurements of the effect of thermal fluctuations on dislocation breakaway from pinning points at high strain amplitudes. This is an area which has not yet been exploited.

IV. Some Selected Experiments of Interest in the Study of Mechanical Behavior of Crystals.

A. Dislocation--Phonon interactions. Recently the properties of the internal friction of copper at megacycle frequencies have been studied by Alers and Thompson[12] and also by Stern and Granato.[8] These experiments are complementary since Alers and Thompson studied the attenuation and velocity in a copper crystal before and after neutron irradiation as a function of temperature and orientation, while Stern and Granato studied the attenuation changes during gamma irradiation as a function of irradiation time and frequency. The results will be discussed together. As already noted earlier, and as is easily seen from Fig. 3, at high frequencies, the attenuation depends only on the dislocation density and the damping constant, and not on the loop length or dislocation line tension. By making an independent count of the dislocation density, the magnitude of the damping constant could be determined. Furthermore, the damping constant was found to be linear in temperature. The latter fact is in accord with what is to be expected if scattering by phonons is the source of the damping, since the damping constant should then be proportional to the phonon density according to Leibfried or linear in temperature (at not too low temperatures). The magnitude of the damping constant found was somewhat in excess (about a factor of 4) of that given by Leibfried's estimate. A similar discrepancy is found in the ratio of the observed to the calculated thermal resistance at low temperatures caused by dislocations. Both Alers and Thompson and Stern and Granato concluded that the physical source of the damping at megacycle frequencies was the scattering of phonons by the moving dislocations and that the vibrating string model is applicable. In addition, by extrapolation of megacycle results into the kilocycle range, and vice versa, Granato and Stern[13] were able to show that the same mechanism accounts for the part of the damping observed in the kilocycle range which can be removed by irradiation pinning. Thus it seems safe to say that the question as to the source of the damping is now understood.

An interesting side result here is that from these results, a question that arose early in dislocation theory can now be answered. The question is: "Can relativistic velocities of dislocations be achieved at stresses near the yield stress?" The answer is no. From the value of the magnitude of the dislocation-phonon interaction strength deduced ultrasonically, one finds that in copper at room temperature the relation between velocity and stress[8] is

$$v/c = \sigma/(75G) \tag{2}$$

where c is the shear wave velocity and G is the shear modulus. Since the theoretical yield stress of a perfect crystal is of order G/30, stresses of the order of the yield stress of perfect crystals would be required for relativistic velocities.

The question which remains is: "What is the mechanism of the scattering of phonons by dislocations?" It has not yet been determined whether this is due to scattering by the strain field (which changes the elastic constants in the vicinity of the dislocation) or whether the scattering is due to a reradiation of sound waves by the dislocation under the influence of the incident phonons.

It has been pointed out by Mason[14] that the same mechanism limits the velocity of dislocations at high stresses as found by Johnston and Gilman[15] in direct observations of the motion of etch pits. From these direct observations, the damping constant for LiFl is found from the relation

$$b\sigma = Bv \tag{3}$$

to be 7.0×10^{-4} (c.g.s. units). This is the same value found for copper ultrasonically. An ultrasonic experiment in LiF is now in progress at Illinois which should check these independent methods. If the ultrasonic determination agrees with that found from the Johnston-Gilman technique, then it should be possible to determine the dislocation-phonon interaction strength for other materials by the (simpler) ultrasonic method.

B. Dislocation-pinning point interactions. Recently, a significant step forward was taken in this area by Bauer and Gordon,[16] who showed that it is possible, by combining ultrasonic and optical measurements, to identify the atomic configuration which is effective in pinning a dislocation in NaCl. Bauer and Gordon found that dislocation pinning in x-irradiated NaCl proceeded at the same rate at low temperatures as it did at room temperature. From this they concluded that diffusion of point defects is not involved in the pinning process and that the pinning points must be produced at the dislocation core. Further, they discovered that dislocations pinned by irradiation at low temperatures can be unpinned by light. This effect is shown in Fig. 4. Bauer and Gordon note that a model used to explain dislocation pinning must satisfy many conditions to be in agreement with their observations. First, defects must be created at, or in the immediate vicinity of, free dislocation segments; in addition, these defects must act as strong pinning points. Furthermore, the pinning defect must possess a characteristic optical absorption band and must be simple enough to be "dissolved" when it is ionized or excited. In the case of rock salt it is found, for example, that unpinning is produced only by light with wavelengths within a fairly narrow band centered about 6300Å. Finally, the defect must occur generally in the alkali halides (with the possible exception of LiF). The model must also be capable of explaining how dislocation pinning can be reversed at low temperatures but converted to a permanent type of pinning if the crystal is warmed to room temperature, and how unpinning illumination is capable of unpinning dislocations at low temperatures while F illumination can cause additional pinning at all temperatures. The model put forward by Bauer and Gordon which fits all these experimental facts is one in which the pinning point is identified as a complex consisting of a jog on a dislocation formed by a Cl-ion and a F-center located one atom distance away and below the slip plane. Because this F-center is in a region where the crystal structure is dilated its absorption band is shifted toward the red, i.e., from 4500Å to 6300Å, an amount which is in agreement with the shift calculated from the strain field about a dislocation in rock salt. Ionization of the F-center by 6300Å light results in electrostatic attraction between the negative-ion vacancy so formed and the Cl-ion forming the jog on the dislocation; recombination of this Cl-ion and its neighboring vacancy causes the pinning point to "dissolve". The model predicts that if a crystal irradiated at low temperature is warmed up in the dark to a temperature where F-centers can diffuse and then cooled down again, it should no longer be possible to remove the pinning points by illumination, as is observed.

Another experiment by Baker[17/] establishes the result that the velocities observed by Johnston and Gilman are not the velocities of the dislocations at a given stress level, but only the velocities of the pinning points. Baker observes periodic dislocation motion of amplitude 1000b at stress levels an order of magnitude below the macroscopic yield stress. The velocity of the dislocations is then of order 1cm/sec ($v \sim dw$, where d is the amplitude of motion and w is the frequency of oscillation). This result demonstrates that the Peierls force is not effective in limiting dislocation motion. The dislocations oscillate at high speeds between pinning points. The overall motion of the dislocations is limited by the speed of the pinning points. This result is a good example of ways by which internal friction measurements can help to distinguish between various postulated deformation mechanisms.

The fact that the dislocations in NaCl are not Peierls stress limited can also be seen from the low temperature elastic modulus measurements of Bauer and Gordon[16] shown in Fig. 5. The fact that the modulus of unirradiated NaCl is lower than that in the irradiated state shows that the dislocations are mobile even at helium temperatures.

As a final example of the study of dislocation-pinning point interactions we may note the interesting observation made by Hikata and Tutumi[18] shown in Fig. 6. They find a striking similarity in the curves of creep rate and ultrasonic attenuation in aluminum at room temperature. In this case the pinning points must be carried along with the dislocations as they move through the lattice. This suggests that jogs may be the effective pinning points in these measurements. The similarity in the shape of the curves may be understood if both depend primarily on the distance between pinning points and this distance is assumed to decrease with time at constant load. This is so because both attenuation and creep rate depend sensitively upon the distance L between pinning points. The former depends upon the 4th power of L, whereas the latter should depend exponentially upon L. Presumably, L decreases because of jog formation in dislocation intersections. Measurements of both attenuation and velocity should permit one to compute the creep rate curve completely from ultrasonic data.

C. Point defect-phonon interactions. An example of the way in which point defect diffusion migration activation energies can be determined is given by the analysis by Granato, Hikata and Lücke[19] of the recovery data of A.D.N. Smith.[20] Smith deformed copper specimens by 1 percent and measured the recovery of the modulus as a function of temperature and recovery time. By assuming that the recovery mechanism is the pinning of dislocations by deformation induced defects, an activation energy of 1 eV was found, which was assumed to be that for vacancy migration energy. At the time of this assignment, all other assignments for this quantity were either near 0.8 eV or less or 1.2 eV or more. In the meantime, the migration energy has been determined by Simmons and Balluffi,[21] and the ultrasonic value is the only previous assignment in agreement with their value. This suggests that the ultrasonic method may be a useful one for such studies.

A second example of the way in which ultrasonic effects can be used in the study of point defect motion is provided by the measurements of Thompson, Blewitt and Holmes,[22] shown in Fig. 7. In this experiment the modulus (or frequency) of a copper specimen was measured as a function of temperature after a neutron bombardment. The modulus at first decreases with temperature in the normal way, but at about 40°K, this process is interrupted. Interstitials are known to move at this temperature, and presumably the normal decrease of modulus is being compensated by dislocation pinning by interstitials. After the pinning is complete, the modulus continues to decrease in the normal fashion. When the specimen is recooled, no anomalous effects occur, indicating that the dislocations are now fully pinned and no longer contribute to the modulus. This measurement provides a good example of the selectivity property of ultrasonic measurements. Only those interstitials which migrate to dislocations are detected. Electrical resistance measurements show that annealing occurs at temperatures below 40°K, but these defects do not travel to dislocations.

D. Dislocation-dislocation interactions. It may be expected that, with increasing deformation, dislocation-dislocation interactions should become important. A maximum in the damping as a function of deformation is often observed.[2,23] However, for deformations of less than a few percent, the observed results seem to be understandable on the basis of the vibrating string model, neglecting dislocation interactions. Some interesting results concerning the distribution of dislocations on various slip systems have recently been obtained by Hikata, Chick, Elbaum and Truell.[24] In these experiments, ultrasonic attenuation and velocity as well as stress were measured continuously as a function of strain. The aluminum specimen was oriented for single slip, and ultrasonic waves of two different orientations were used. The results are shown in Figs. 8 and 9. In Fig. 8, results are given for the case where the ultrasonic wave had no shear stress component in the primary glide system. After an initial rise, the stress-strain curve clearly indicates the existence of easy glide for approximately 0.2 percent tensile strain. The corresponding attenuation change is very similar to the stress-strain curve, with little increase of attenuation during the easy glide region even though the dislocation density is increasing greatly as can be seen from the results of Fig. 9. The behavior of longitudinal waves, which have shear stress components in the primary glide system, is quite different. As seen in Fig. 9, the attenuation increases quite rapidly with increasing strain, from the beginning of the deformation, and does not exhibit any special characteristics associated with easy glide. These results show clearly the sensitivity of the ultrasonic measurements to the distribution of dislocations. With further development of the technique, we may look forward to the possibility of having continuous plots of dislocation density and loop lengths as a function of strain in various slip systems.

A second point of much interest in these measurements concerns the interesting variations in the ultrasonic velocity. This is of great usefulness in checking certain fine details in the predictions of the theory,[25] but will not be discussed here.

V. Conclusions.

It is clear that ultrasonic measurements provide us with much useful information concerning defect interactions of importance in understanding the mechanical behavior of crystals. In this review we have discussed qualitative features of a few selected results, but it should be emphasized that results in quantitative form are obtained. Some of the conclusions which have been noted are:

1. The physical source of the sound damping is the dislocation-phonon interaction. Relativistic dislocation velocities cannot be achieved in copper and LiF at room temperature at the yield stress.

2. The vibrating string model is confirmed by the measurements. This model appears to apply even for moderate (up to a few percent) deformation.

3. The atomic configuration making up a pinning point has been identified in NaCl.

4. Deformation in common alkali halides is determined by the motion of pinning points. The dislocation is free to move between such points. Similar considerations seem to apply for creep in aluminum.

5. The migration activation energies of point defects in small concentrations can be determined ultrasonically.

6. A detailed description of dislocation densities and loop lengths in various slip systems may be obtained ultrasonically.

Work currently in progress on the effect of thermal fluctuations on dislocation unpinning promises to provide us with useful information concerning the temperature dependence of the yield stress. There is a need for models suitable for describing the internal friction of heavily deformed materials.

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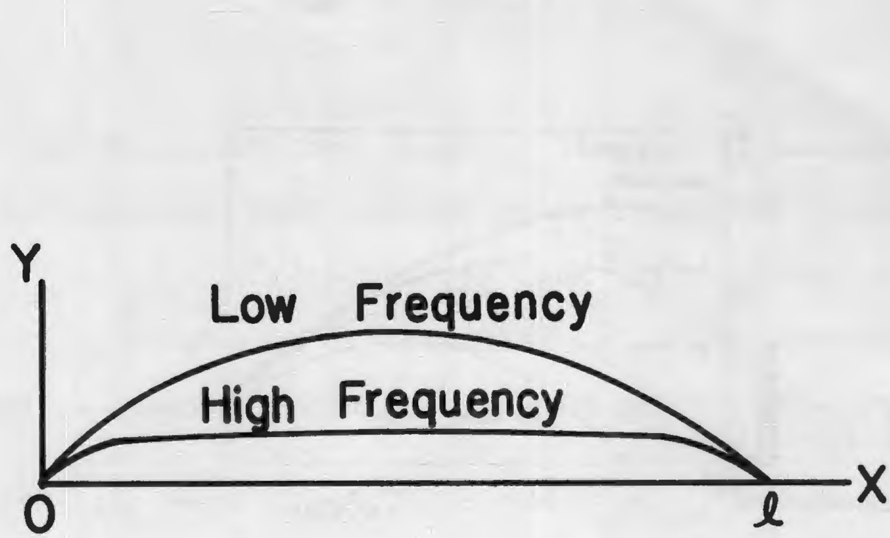

Fig. 1 Schematic dislocation displacement y(x) as a function of coordinate x for a) low frequencies and b) high frequencies. At low frequencies the displacement is limited by tension forces. At high frequencies, the displacement is limited by viscous forces.

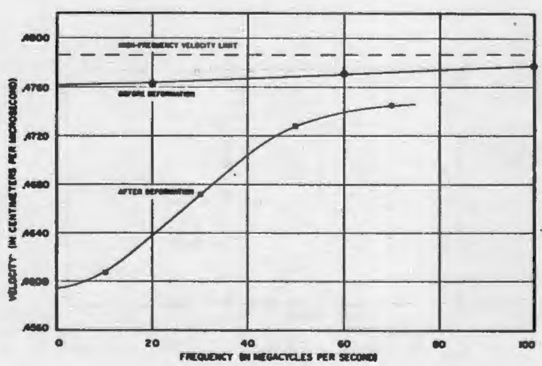

Fig. 2 Velocity dispersion for compressional elastic waves propagating in the (100) direction in NaCl. Deformation increases the magnitude of the dispersion from 0.5% to 4% and moves it to lower frequencies. (After Granato, de Klerk and Truell.)

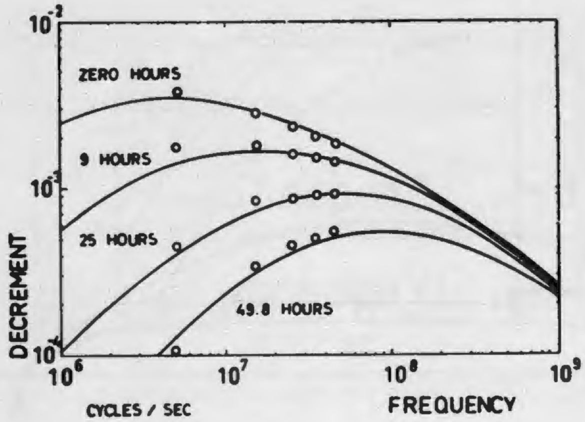

Fig. 3 The dislocation decrement as a function of frequency for several times during cobalt gamma irradiation in a 6000 curie source. The solid curves are theoretical. (After R. M. Stern and A. Granato.)

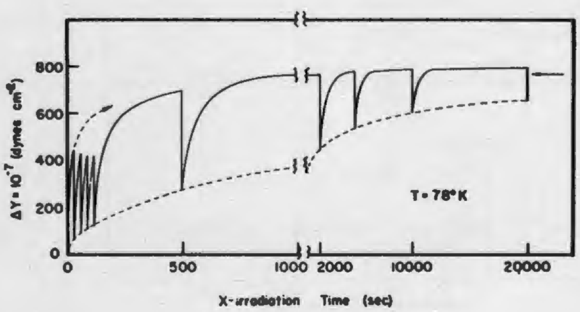

Fig. 4 Behavior of the modulus change of a deformed NaCl crystal, ΔY, during successive x-irradiations and exposure to visible illumination. (After Bauer and Gordon.)

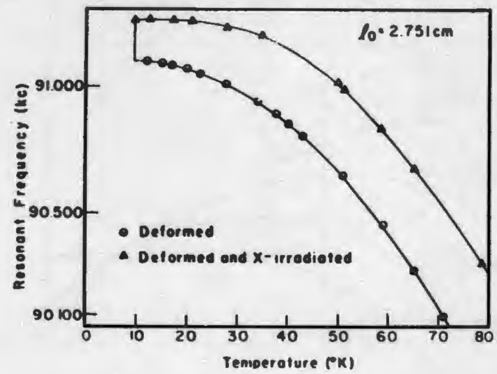

Fig. 5 Dependence of the resonant frequency of a deformed NaCl crystal, before and after x-irradiation, on temperatures between liquid-helium and liquid nitrogen temperature. l_o represents the room temperature specimen length. (After Bauer and Gordon.)

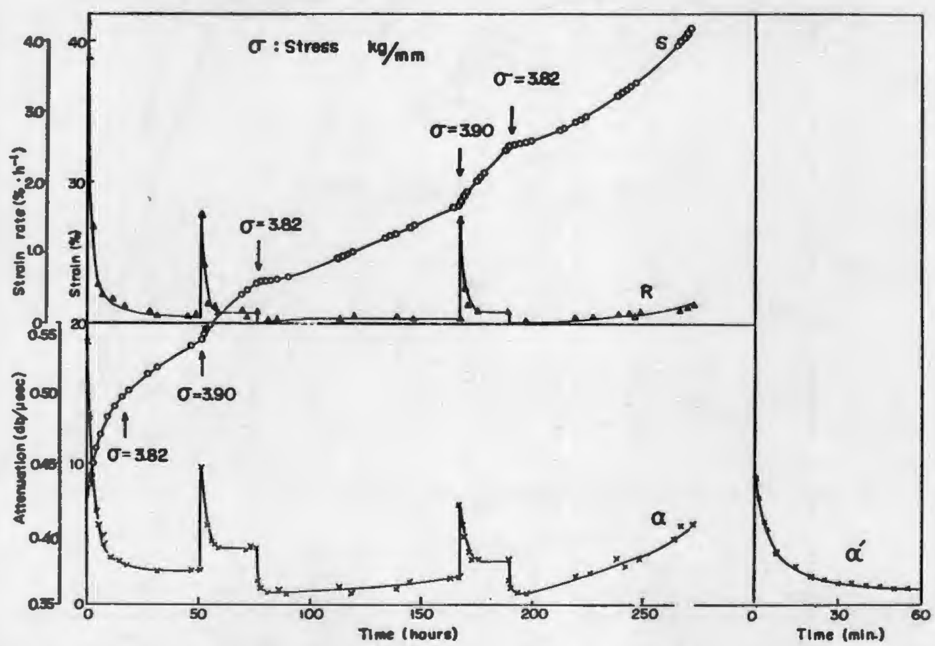

Fig. 6 Curve S: Creep strain-time relation.
Curve α: Attenuation-time relation.
Curve R: Creep rate-time relation.
Curve α': Recovery of attenuation.
(After Hikata and Tutumi.)

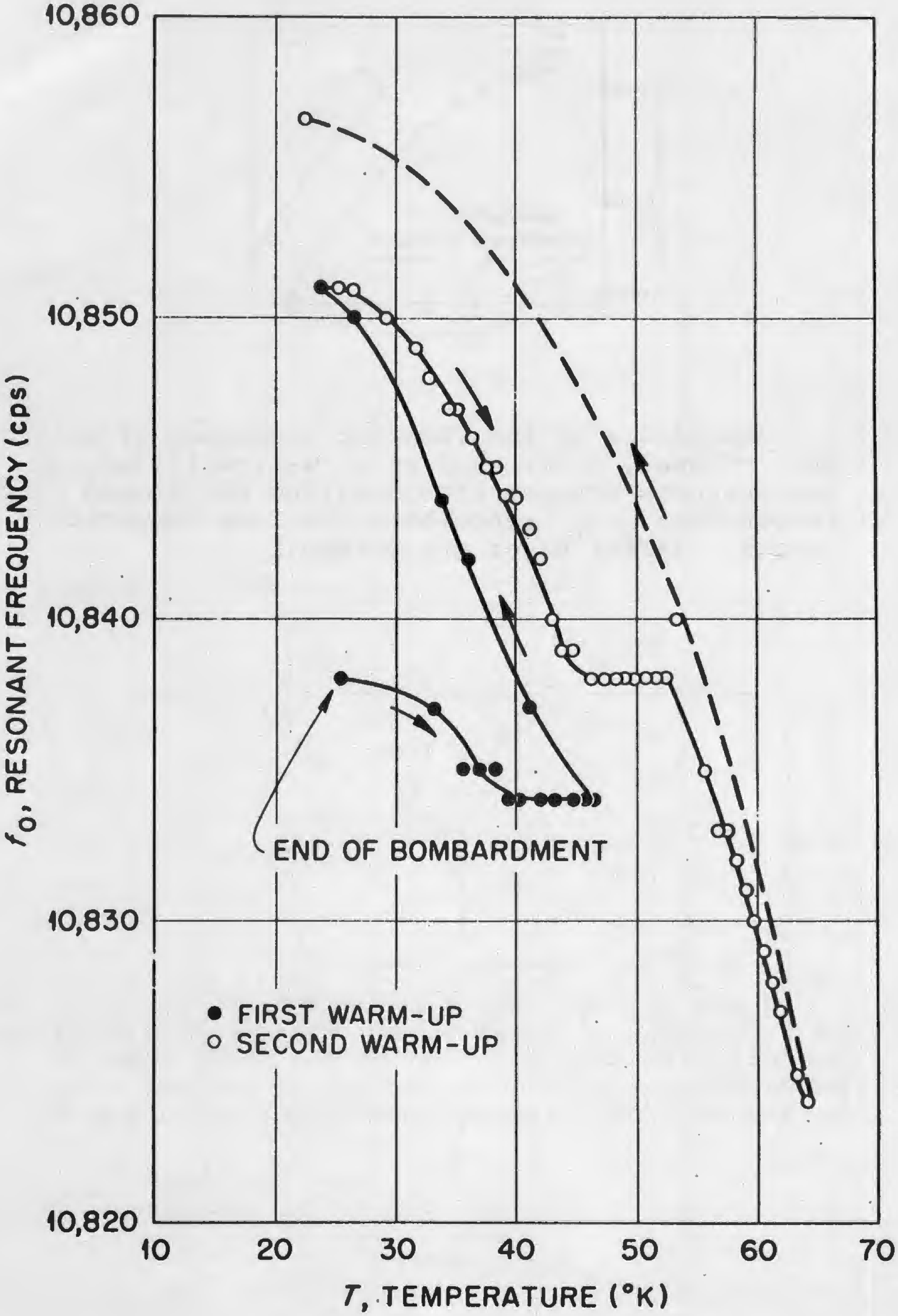

Fig. 7 **The charge in resonant frequency with temperat upon warming up from fast neutron irradiation at 20°K. (After Thompson, Blewitt and Holmes.)**

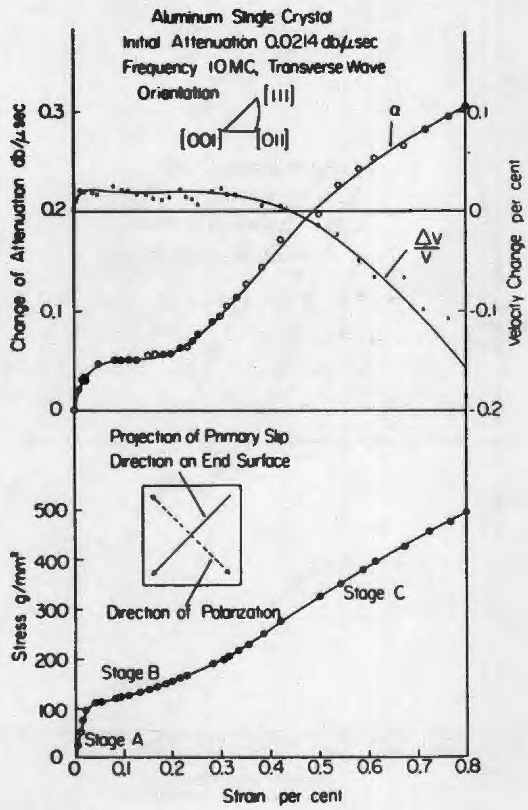

Fig. 8 Stress, shear wave attenuation and velocity change as a function of total strain for <0.5> orientation. The polarization direction of the shear wave is perpendicular to the projection of primary slip direction on end surface. (After Hikata, Chick, Elbaum and Truell.)

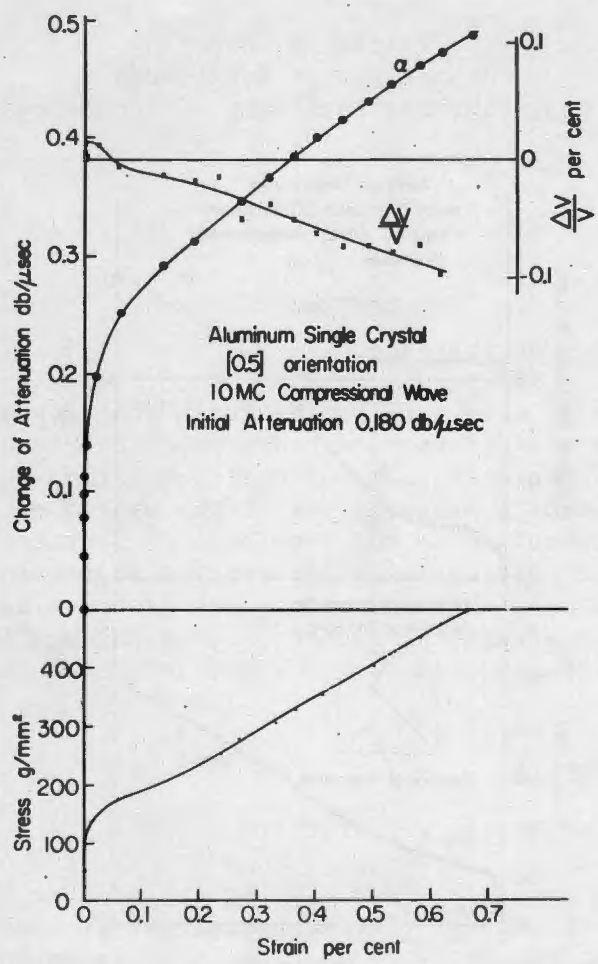

Fig. 9 Stress, longitudinal wave attenuation and velocity change as a function of total strain for <0.5> orientation. (After Hikata, Chick, Elbaum and Truell.)