Measurement of the Force-Distance Profile for the Interaction between a Dislocation and a Point Defect*

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It is demonstrated that the force-distance profile for a dislocation and a single point defect can be determined from amplitude-dependent internal-friction measurements.

The strength of many crystalline materials is determined in principle by the details of the interaction between dislocations and point defects. Theories of the force-distance law for the interaction between a dislocation and a single point defect have been available for many years,[1] but this force law has never been measured. We demonstrate here that important features of this force can be measured ultrasonically.

We consider a straight dislocation segment of length $2l$, interacting at its center with a single pinning point. The true interaction force-distance law is not known. However, it would be expected to have the same general features as the Cottrell expression,[2] which is shown in Fig. 1. F_m is the maximum value of the interaction force F as a function of the distance s of the dislocation from the projection of the point defect on the dislocation glide plane. The stress necessary for mechanical breakaway in the absence of thermal fluctuations is $\sigma_m = F_m/lb$, where b is the magnitude of the Burgers vector. For an applied stress $\sigma < \sigma_m$ and long enough segment lengths l,[3] the activation free enthalpy $H(\sigma)$ required for thermal breakaway from a single pinning point is given by

$$H(\sigma) = \int_{s_1}^{s_2} [F(s) - F_m \sigma/\sigma_m] ds$$
$$= lb \int_{\sigma}^{\sigma_m} (s_2 - s_1) d\sigma, \quad (1)$$

where s_1 and s_2 are the roots of $F(s)/F_m - \sigma/\sigma_m = 0$.

Only the difference $s_2 - s_1$ enters in the expressions for $H(\sigma)$. A shear of $F(s)$ parallel to the s axis, so that the abscissa becomes $s_2 - s_1$, defines a "force-width profile,"[4] $P(s)$, also shown in Fig. 1. $P(s)$ contains the same information relevant to the calculation of $H(\sigma)$ as does $F(s)$. Only $P(s)$ can be derived from the experimental measurements. $F(s)$ can then be obtained by reversing the shearing process, only after assuming, on the basis of a theoretical model, the shape of $F(s)$ between $s = 0$ and s corresponding to F_m.

If $F(s)$ is smooth near its maximum, it can be approximated by a parabolic curve (dashed curve in Fig. 1) over a range near σ_m. The corresponding width profile is also shown in Fig. 1 (dash-dotted curve). For the parabolic force, the activation energy has the simple stress dependence

$$H_p(\sigma) = K(1 - \sigma/\sigma_m)^{3/2}, \quad (2)$$

where K is a constant determined by F_m and the curvature of $F(s)$ near its maximum. For a constant probability of overcoming the barrier, $H_p(\sigma)$ can be taken as a multiple of kT. It then follows that the applied stress should have a temperature dependence

$$\sigma = \sigma_m - AT^{2/3},$$

where A is a constant determined by K and the effective attack frequency.[5]

The form of Eq. (3) was first obtained by Mott

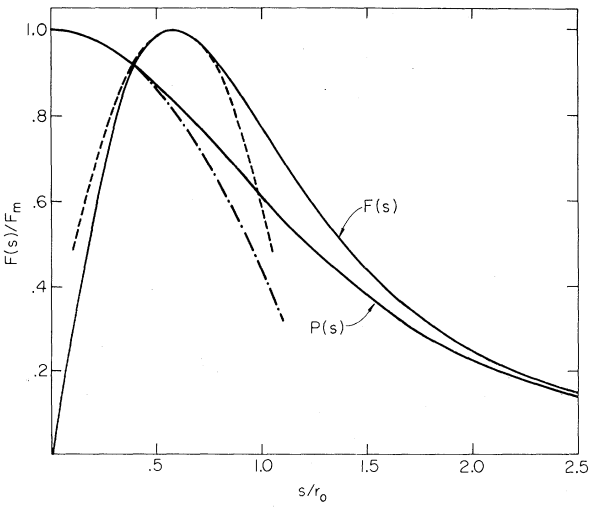

FIG. 1. Force-distance curve $F(s)$, and its corresponding width profile $P(s)$, for an edge dislocation interacting with a point defect located a distance r_0 from the dislocation glide plane. The dashed curve is a parabolic fit to $F(s)$ near its maximum and the dash-dotted curve is its corresponding width profile.

and Nabarro[6] in 1948 who, however, considered a different model. They calculated the temperature dependence of the yield stress to be expected for dislocations held up by the stress field around a precipitate by making the approximations that a constant segment length had to be thrown over a sinusoidal stress barrier. That the form of this result had more general validity was anticipated by Mott.[7]

Measurements of the temperature dependence of the yield stress at constant strain rate have often been fitted by theoretical curves based on different assumed force-distance laws for the dislocation–point-defect interaction, but the measurements are not sufficiently precise to distinguish between the theories or to determine the temperature law. Amplitude-dependent internal-friction measurements also depend upon the dislocation–point-defect interaction. These have the advantage that the measurements do not change the state of the system. Furthermore, for low stress amplitudes the dislocations need to overcome only the weakest point defects. The nature and density of these can be experimentally controlled, for example, by dilute alloying. This is in contrast to the case for the macroscopic yield stress, where the dislocations may be forced to overcome other obstacles as well, which are often out of experimental control and are not of primary interest here.

A zero-temperature theory for amplitude-dependent damping was given by Granato and Lücke,[8] according to which the damping results from the unpinning of dislocations from pinning points under the action of an external stress. The theory was extended to finite temperatures in a series of articles by Granato, Lücke, Teutonico, Schlipf, and Heuser.[3,5,9-12] Blair and co-workers[13,14] have also tried to find analytic solutions of the basic equations,[9] while Peguin and Birnbaum[15] have calculated some numerical solutions. A physical discussion and summary is given in Ref. 12. For the low-temperature region, the principal results are as follows:

(a) The stress amplitude $\sigma_\Delta(T)$ required to produce a constant Δ, at a temperature T, corresponds to a constant fraction of dislocation segments being broken away from point defects and is directly related to calculations for the yield stress of macroscopic flow measurements.

(b) For stress amplitudes $p\sigma_m < \sigma_\Delta < \sigma_m$ and relatively pure materials, the stress σ_Δ should have the temperature dependence of Eq. (3), regardless of the form of the dislocation force-distance law, provided only that the force has a smooth maximum. If the force-distance law is given by the Cottrell expression, then by comparing the parabolic approximation to $P(s)$ in Fig. 1, one would expect $p \simeq 0.85$.

(c) For stress $\sigma^* < \sigma_\Delta < p\sigma_m$, deviations from Eq. (3) should occur which depend on the detailed force-distance law between a dislocation and a single pinning point. The stress σ^* depends upon the density of pinning points, being lower for purer materials.

(d) For $\sigma_\Delta < \sigma^*$, the activation process no longer depends only on the details of the force-distance law. For such low stresses, breakaway from a single pinning point becomes ineffective because the broken-away state is a relatively high-energy state, and the repinning rate is higher than the unpinning rate. Breakaway to a lower energy state can now occur only by simultaneous unpinning from more than one pinning point, but the activation energy for this is higher and the process occurs only at higher temperatures.

An observation of the $T^{2/3}$ law of Eq. (3) would be of especial significance since this form is required theoretically on very general grounds and measurements which purport to measure the force-distance profile between a dislocation and a single point defect should meet this requirement as a necessary and crucial test. There are many measurements of the decrement as a function of applied stress amplitude and temperature, but the $T^{2/3}$ law has not previously been observed. However, the conditions prescribed above (high purity and $p\sigma_m < \sigma < \sigma_m$) have not been met in previous measurements. Also the accuracy required to determine the temperature dependence in this small stress interval has not previously been achieved. These conditions are realized in the following experiments.

Internal-friction measurements were performed between 1.7 and 20 K in undeformed single crystals of Al and Cu, having a total impurity content of approximately 40 ppm. The crystals were spark cut and chemically polished in the form of square-sectioned samples with ⟨321⟩ axes. Each sample was then carefully epoxied to a longitudinal quartz resonator and the resulting composite oscillator was cooled with a double-wall He cryostat. Decrement measurements were taken at different temperatures during the free-amplitude decay in a vacuum of 10^{-3} Torr from a fixed excitation at a frequency of 49 kHz.[16] The strain-amplitude–independent decrement was subtracted, at each temperature, from the total decrement

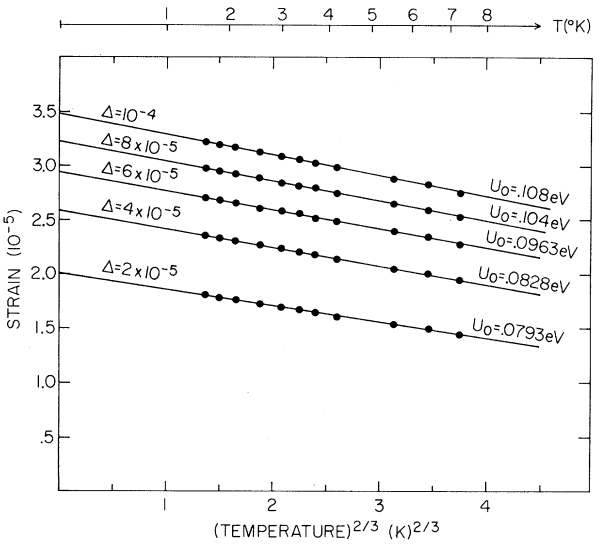

FIG. 2. Measured values (dots) of the strain amplitude as a function of (temperature)$^{2/3}$ required to produce a constant value of the decrement Δ for different decrement levels in Al. The values of U_0 shown are obtained assuming that the force-distance law follows the Cottrell expression, shown in Fig. 1.

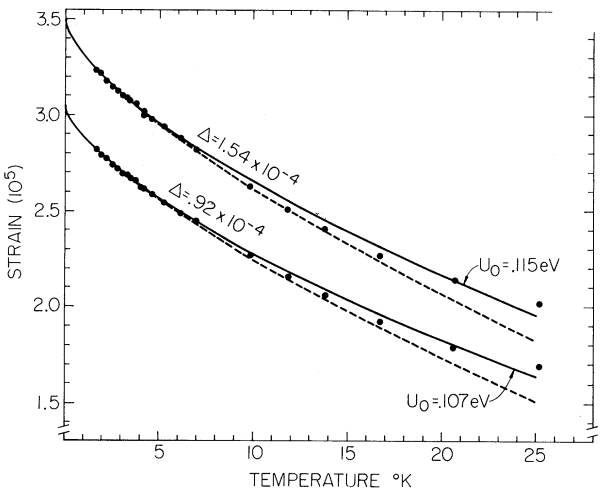

FIG. 3. Measured values (dots) of the strain amplitude as a function of temperature for two constant values of the decrement Δ in Al. The solid curves are the calculated strain-amplitude temperature dependence, assuming that the force-distance law follows the Cottrell expression with an interaction energy chosen so as to give the best fit to the data below 6 K. The segmented curves are the calculated strain-amplitude temperature dependence for the parabolic fits to the Cottrell expressions for the force-distance law used to calculate the solid curves.

in order to obtain the amplitude-dependent part, which was then corrected against the inhomogeneity in the applied strain.[17] The resultant set of decrement-versus-strain curves were used to obtain the strain-versus-temperature curves for constant values of breakaway-induced losses. The results for Cu are similar to those for Al, which are shown in Figs. 2 and 3.

Figure 2 shows the strain amplitude measurements for constant values of the decrement versus $T^{2/3}$ between 1.7 and 7.2 K. This exponent gives the best linear computer fit to the data. Measurements in three samples gave almost identical results.

The maximum interaction energy U_0 between the dislocation and point defect cannot be determined from these results alone. Measurements over a wider range are needed to determine the full shape of the force-distance curve. If the Cottrell[2] form for the force law is assumed, the values of U_0 shown in the figure are obtained. These would be expected to be the same if all dislocations were of the same type and only one type of impurity were present. The values found are nearly constant, but there is a systematic change with decrement level. More detailed studies with control over dislocation and impurity type will be required to pursue this further.

Figure 3 shows the strain amplitude for two values of the decrement Δ, on a linear and more extended temperature scale. The solid curves show the numerically calculated strain temperature dependence[18] which should be expected for the Cottrell force-distance law with an interaction energy chosen so as to give the best fit to the data below 6 K. The segmented curves show the numerically calculated strain temperature relation for the parabolic fits to the Cottrell expression for the force-distance law used to calculate the solid curves and, as expected from Eq. (3), they follow a strict $T^{2/3}$ dependence.

For the two crystals studied, where the interaction energies were of the order of 0.1 eV, the increase in the temperature to 7 K meant a reduction in the breakaway stress by approximately 15% from that necessary to produce mechanical breakaway. It is then within this range from the top that the force-distance law between dislocations and point defects can be considered parabolic. Above approximately 7 K neither the Cottrell nor the parabolic force-distance laws seem to represent accurately the true operating force. As long as the temperature is low enough and the dislocation loops long enough so that only isolat-

ed breakaway processes occur along the pinned dislocations,[3] a computer fitting method[18] can be used to map the true force-distance profile for the dislocation–point-defect interaction.

We believe that this technique can be used to measure the dislocation–point-defect interaction force under a wide variety of controlled conditions.

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[1]For example, F. R. N. Nabarro, *Theory of Crystal Dislocations* (Oxford Univ. Press, London, 1967).

[2]A. H. Cottrell, in *Report of a Conference on Strength of Solids, H. H. Wills Physical Laboratory, University of Bristol, 7–9 July 1947* (The Physical Society of London, London, England, 1948), p. 30.

[3]L. J. Teutonico, K. Lücke, F. W. Heuser, and A. V. Granato, J. Acoust. Soc. Amer. 45, 1402 (1969).

[4]G. B. Gibbs, Phil. Mag. 20, 611 (1969).

[5]A. V. Granato, K. Lücke, J. Schlipf, and L. J. Teutonico, J. Appl. Phys. 35, 2732 (1964).

[6]N. F. Mott and F. R. N. Nabarro, in *Report of a Conference on Strength of Solids, H. H. Wills Physical Laboratory, University of Bristol, 7–9 July 1947* (The Physical Society of London, London, England, 1948), p. 1.

[7]N. F. Mott, Phil. Mag. 1, 568 (1956).

[8]A. V. Granato and K. Lücke, J. Appl. Phys. 27, 583 (1956).

[9]L. J. Teutonico, A. V. Granato, and K. Lücke, J. Appl. Phys. 35, 220 (1964).

[10]K. Lücke, A. V. Granato, and L. J. Teutonico, J. Appl. Phys. 39, 5181 (1968).

[11]F. W. Heuser and K. Lücke, to be published.

[12]A. V. Granato and K. Lücke, to be published.

[13]D. G. Blair, T. S. Hutchison, and D. H. Rogers, J. Appl. Phys. 40, 97 (1969), and Can. J. Phys. 48, 2943, 2955 (1970), and Can. J. Phys. 49, 633 (1971).

[14]D. G. Blair, J. Appl. Phys. 43, 37 (1972).

[15]P. Peguin and H. K. Birnbaum, J. Appl. Phys. 39, 4428 (1968).

[16]K. Hultman and R. B. Schwarz, to be published.

[17]H. Schaefer and F. Povolo, Scr. Met. 7, 445 (1973); S. Asano, Scr. Met. 8, 947 (1974).

[18]R. B. Schwarz and A. V. Granato, to be published.

Origin of Very High-Energy Cosmic Rays*

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It is suggested that a single source of cosmic rays (supernovas) occurring in all galaxies can produce the observed spectrum and the observed anisotropy, and is predictable from supernova shock theory. Below 10^{13-14} eV the source and observed spectrum are the same, $N(>E) \propto E^{-\alpha}$, $\alpha \simeq 1.75$. Above 10^{14} eV I predict a source with $\alpha \simeq 1.2$. Galactic leakage above 10^{15} eV is linear so that $\alpha \simeq 2.2$ as observed. Above 10^{19} eV cosmic rays fill the metagalaxy to a flux several times the anisotropic residual flux from a few events in our galaxy as observed.

Cosmic rays are assumed to be either galactic or extragalactic in origin. The total energy required for filling metagalactic space with all cosmic rays is improbably high and, in addition, very heavy nuclei and electrons must then have a separate origin.[1] On the other hand, the near isotropy at energies above where any reasonable galactic magnetic field could affect particle containment dictates an extragalactic origin.

I propose instead that the identical source mechanism occurs in all galaxies and that we observe at very high energy a combination of our own statistically distributed sources as well as an isotropic flux from all other galaxies. This assumption requires a source spectrum that is flatter by one power of E than the observed spectrum for $E > 10^{15}$ eV. The observed spectrum is then a result of leakage from our galaxy as well as attenuation in metagalactic space. The consequences of this assumption are as follows:

(1) The resulting spectrum agrees with observations, including the "bump" at $10^{13} < E < 10^{15}$ eV.

(2) The anisotropy becomes an increasing function of energy above 10^{15} eV, and the degree of anisotropy agrees with the recent large anisotropy interpreted[2] for all measurements $E > 10^{19}$ eV.

(3) The energy spectrum of the local anisotropy should be flatter than the isotropic flux as observed.[2] That is, the anisotropy should consist of a few local peaks of flux (not valleys) in solid