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INTERNAL FRICTION AND ULTRASONIC YIELD STRESS OF THE ALLOY 90 Ti 6 Al 4 V

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Abstract—The alloy 90 Ti 6 Al 4 V is used as a mechanical transformer to produce high strains in metal samples. The internal friction, modulus defect and limiting stable strain amplitudes have been studied over wide frequency and temperature ranges. The internal friction Q^{-1} of annealed samples has the low value of 5×10^{-5} which is independent of the frequency and decreases only slightly at high temperature. The result is consistent with a model in which the energy loss is produced by a non-linear motion across the Peierls kink barrier. The value found for the ratio of the internal friction to the modulus change is 0·03 which is intermediate between two theoretical calculations.

This material becomes unstable at strains from 3 to 5×10^{-3}. Evidence is presented which shows that this is a material with a Cottrell type pinning with pinning points on the average about 3 Burgers distances apart. Assuming that on the average one kink occurs between each pinning point, the modulus defect is consistent with about 10^{10} dislocations per cm^2. The frequency and temperature variations are consistent with this model.

1. INTRODUCTION

THE ALLOY 90 Ti 6 Al 4 V has been suggested for use in the supersonic plane. It also has been used as a mechanical transformer[1] for increasing the strain in a sample whose high amplitude and fatigue properties are to be studied. Hence it is important to determine its internal friction and modulus defect at both low amplitude and high amplitude vibration levels.

The internal friction at low amplitudes has been found to be independent of the frequency from 17 KHz to 10 MHz, and to be independent of the temperature (within about 30 per cent) from 4·2° up to 600°K. An impurity type peak is observed which peaks at 180°K for 10 MHz but after this effect is removed, the background internal friction is independent of the temperature. By annealing strained samples and straining annealed samples it is shown that this background internal friction is due to dislocations. The results obtained are in agreement with a 'low frequency' internal friction model[2] for which the damping results from the lattice vibrations generated when the kink crosses the kink Peierl's barrier. Theoretical studies[3, 4] have shown that this energy loss requires a dissipative stress τ_{dp} equal to from 0·01 to 0·1 of the kink stress σ_k, to replace this energy lost to the lattice vibrations. This ratio β can be measured by determining the ratio of the internal friction change ΔQ^{-1} to the modulus defect change determined when a strained sample is annealed. This value was found to be about 0·03 in agreement with the results in copper.

High amplitude measurements show that the internal friction is constant up to longitudinal strains of 4×10^{-3}. This is about half the static strain. The variation of the static stress with temperature indicates that the alloy is a material with Cottrell type pinning with pins on the average separated by about three times the Burger's distance b which is $2{\cdot}95 \times 10^{-8}$ cms for titanium. Assuming one kink on the average for each pinned loop, the internal friction and the modulus defect are consistent

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with a dislocation density of about 10^{10} dislocation density of about 10^{10} dislocations per cm^2.

2. EXPERIMENTAL RESULTS FOR HIGH AMPLITUDE EFFECTS

The low frequency measurements (17 KHz) were performed using a transducer and transformer system described previously[1]. Since the system, the shape of the specimen and the method of calibration have been described previously, the reader is referred to this publication[1]. An evacuated container is used to eliminate the radiation resistance. The modulus change in the sample alone can be obtained by multiplying $2\Delta f/f$ by the ratio $(M_0+M_S)/M_0$ where M_0 is the effective mass of the transducer and transformer and M_S the mass of the sample; Δf is the change in frequency and f the frequency of the sample and system.

The internal friction and modulus change for a typical sample of the alloy are shown by Fig. 1. Instability sets in at maximum strain levels which may vary from 3×10^{-3} to 5×10^{-3}. The instability is accompanied by a sharp rise in the internal friction Q^{-1} and by the modulus defect $\Delta S/S$. After this happens the modulus defect has increased by about three per cent and the internal friction increases by nearly a factor of 10. Furthermore the critical longitudinal stress is reduced by over a factor of 10. It appears that this behavior is connected with the presence of long dislocation loops which are not repinned at room temperature. Specimens annealed in a vacuum at 750°C for 1 hr are sometimes returned to a normal condition but not always. The presence of long dislocation loops is confirmed by the non-linearity of the pick-up voltage as a function of the frequency, as shown by Fig. 2. Instead of a symmetrical response about a resonant frequency, the pick-up voltage builds up to a maximum as the frequency is reduced, after which it discontinuously drops to a low value. On the increasing cycle the response remains low for 34 cycles above the discontinuous decrease frequency after which it discontinuously increases to the value of the decreasing curve. This type of performance is

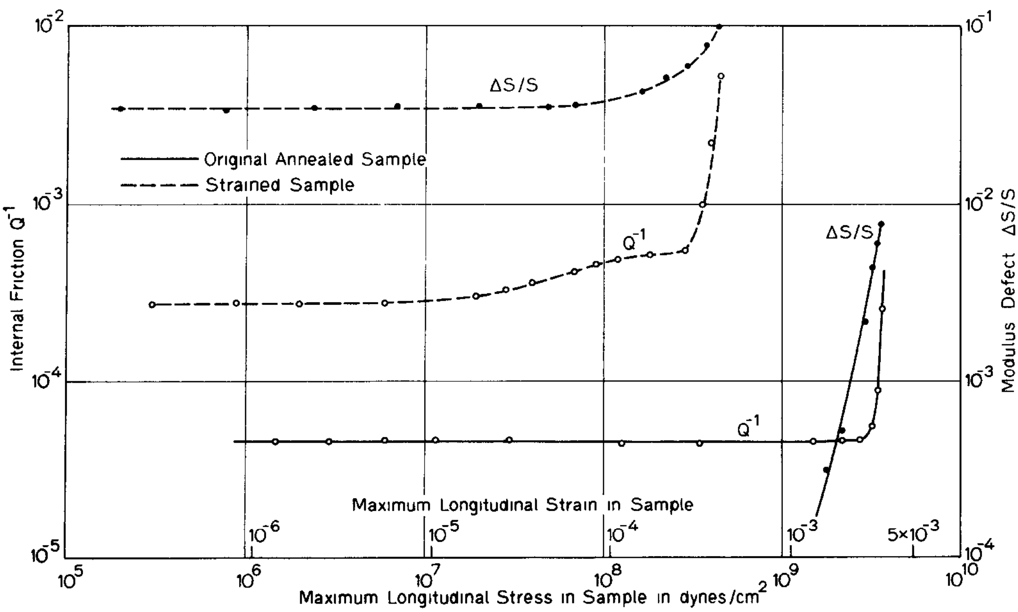

Fig. 1. Typical internal friction and modulus defect for an annealed sample of the alloy 90 Ti 6 Al 4 V. Dashed line shows the result obtained after the sample has become unstable.

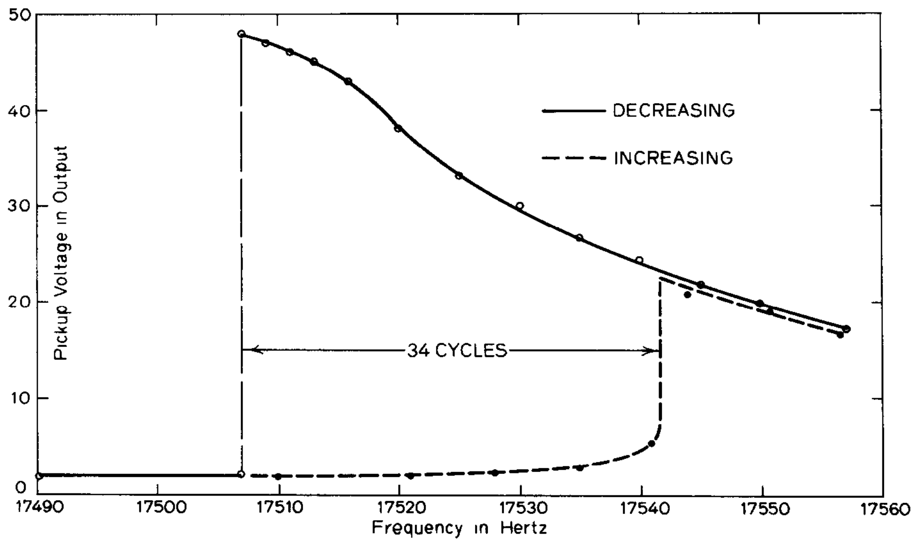

Fig. 2. Pick-up voltage plotted against the frequency for an unstable sample near its critical point. This result is indicative of a non-linear soft spring characteristic.

characteristic of a non-linear spring which is softer at high amplitudes. This performance is in agreement with the existence of long dislocation loops which bow out and become less stiff as their displacement increases.

It appears that in the annealed state the dislocations are closely pinned in an atmosphere approximating a Cottrell type pinning [5]. This is shown by the yield stress curve for static stresses, Fig. 3, plotted as a function of the temperature. According to the Cottrell atmosphere concept the concentration along the dislocation is given by

$$c = c_0 e^{U/kT} \tag{1}$$

where c is the concentration along the dislocation, c_0 the density of the impurities in the body of the material, U the binding energy to the impurity, k Boltzmanns constant and T the absolute temperature. If we apply this concept to the data of Fig. 3, we find that U is about 0·5 eV and the average separation between pinning points is 10^{-7} cm or about 3 Burgers distances. From internal friction measurements it appears that on the average there is one kink between each pinning point.

Measurements were made of the ultrasonic yield stress as a function of the temperature. While the scatter is quite large, it appears that the ultrasonic yield stress is about half the static yield stress. Similar results are found for other metals [6]. This may be caused by the change in the positions of the pinning points with ultrasonic stress as suggested by Alefeld [7].

3. INTERNAL FRICTION MEASUREMENTS

Another important property of the alloy is its internal friction. This is connected with the motion of dislocations. This can be verified by annealing and straining experiments. When the sample is initially formed by the turning process its Q is lower and its frequency lower than found after being annealed for one hour at 760°C. Table 1 shows the resonance frequency and Q in the as formed condition and in the heat treated condition. The Q increases and the resonant frequency increases. If we take the average of the increase in Q^{-1}, i.e. ΔQ^{-1}, and the average increase in the modulus change $\Delta S/S$ the ratio is about 0·03 as shown by the last column. Similarly for a strain of

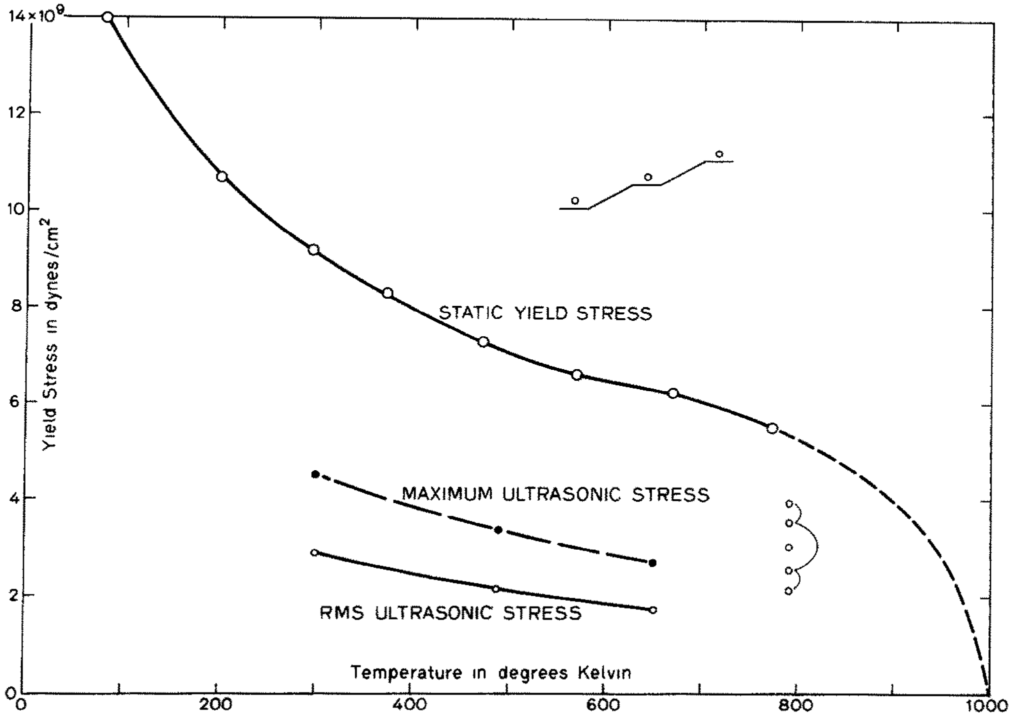

Fig. 3. Yield stress for static and dynamic conditions.

Table 1

Sample	As formed		Heat treated		Δf	ΔQ^{-1}	$\Delta S/S$	$\frac{\Delta Q^{-1}}{(\Delta S/S)}$
	f_R	Q	f_R	Q				
1	17326	1920	17418	6960	92	$8{\cdot}61\times10^{-4}$	$2{\cdot}48\times10^{-2}$	0·0346
2	17377	2480	17468	8750	87	$6{\cdot}56\times10^{-4}$	$2{\cdot}34\times10^{-2}$	0·0282
3	17335	2480	17415·5	9200	80·5	$6{\cdot}7\times10^{-4}$	$2{\cdot}17\times10^{-2}$	0·0308
4	17392	2900	17460	10900	68	$5{\cdot}55\times10^{-4}$	$1{\cdot}83\times10^{-2}$	0·0304

Average 0·0308.

2 per cent, the internal friction increases to about 5×10^{-4} and the frequency drops by 160 cycles. Both of these experiments are indicative of a dislocation background.

Since the average pinning length is very small any dislocation dissipation due to a damping proportional to the dislocation velocity is negligible and all the damping must be due to the 'low frequency' component discussed in the next section. Since any other type of damping is negligible, this material provides an opportunity for evaluating the frequency and temperature variation of this 'low frequency' type of damping. Measurements were made on a number of unannealed samples with the result shown by Fig. 4. In order to cover a wide frequency range some shear wave measurements were made at 10 MHz. This raises the question of whether some of the attenuation was due to scattering losses from the polycrystalline grain structure of the material. This alloy has a fine grain structure with grains from 10^{-3} to 10^{-4} cm. Measurements were made over a frequency range and it was found that grain scattering effects did not produce an appreciable loss until a frequency of 30 MHz. Hence the values measured at 10 MHz are due to dislocations. Since the internal friction scatters over a range, it appears probable that the values are substantially independent of the frequency as shown by Fig. 4. Lower values still can be obtained by annealing the materials.

Measurements were then made to evaluate the internal friction as a function of the temperature. Since the pulsed 10 MHz system can be taken down to liquid helium temperature, measurements were made on two samples as shown by Fig. 5. Both samples showed an impurity type peak occurring at about 180°K. This did not appear to be due to dislocations since straining the sample changed the dislocation background but did not change the difference of the peak height from the background. If we analyze the peak as a single relaxation process, as shown by Fig. 6, it appears to be somewhat broader than a single relaxation and with an average relaxation energy of about 0·1 eV. If we subtract out the peak, the dislocation loss appears to be independent of the temperature.

Since it was not possible to go higher in temperature with this arrangement, measurements were made with the transducer transformer arrangement, using a furnace around the specimen with the combination mounted in an evacuated chamber. A number of different samples were taken up to 500°K with similar results. When a sample was taken up to 600°K an apparent increase seems to occur

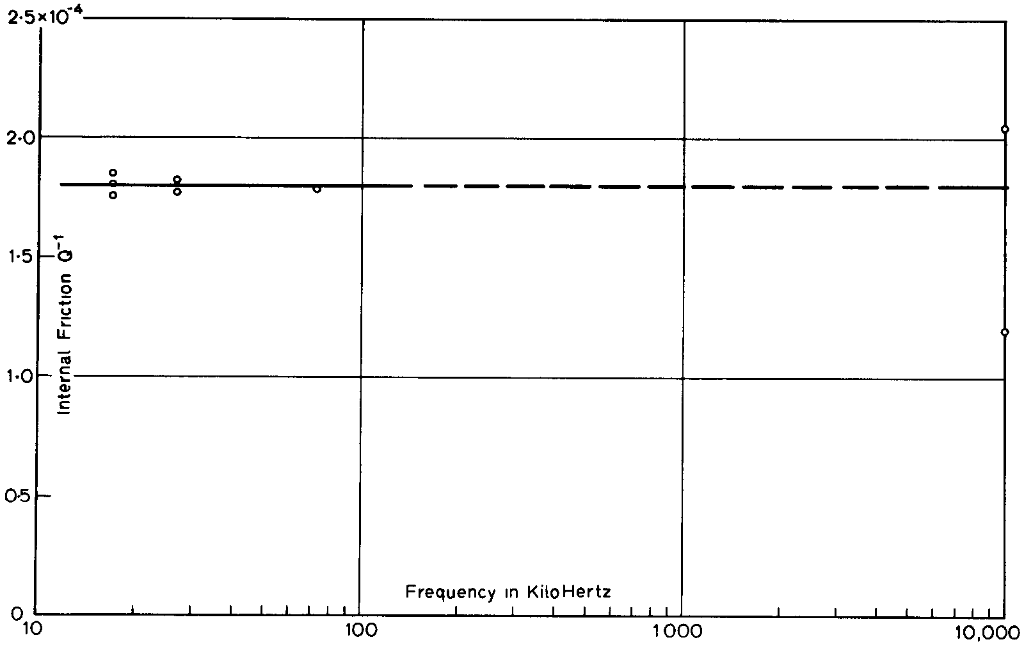

Fig. 4. Internal friction for unannealed samples as a function of the frequency. Measurements up to 70 kHz were made with longitudinal vibrations. Measurements at 10 MHz were made by a pulsing method using shear waves.

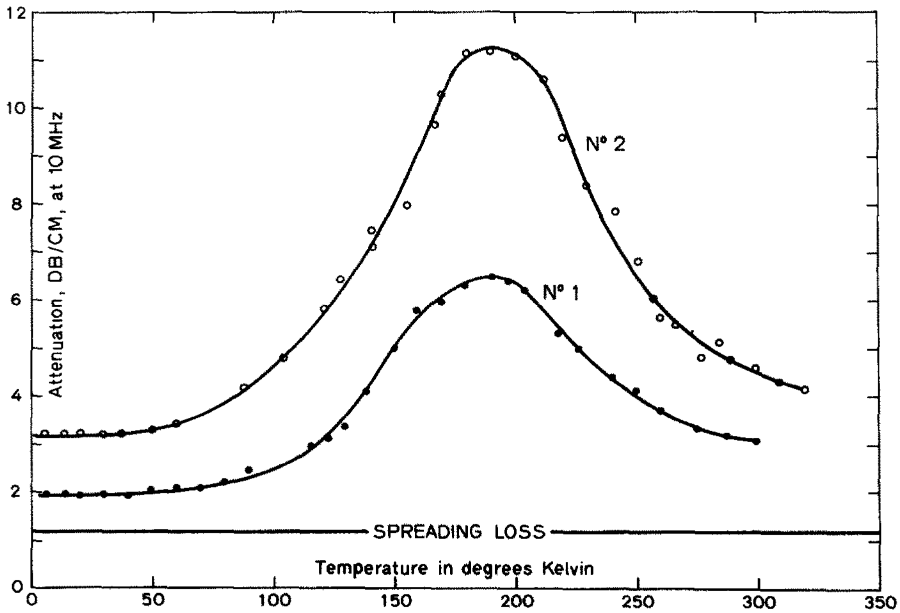

Fig. 5. Attenuation for shear waves at 10 MHz plotted against the temperature for two unannealed samples.

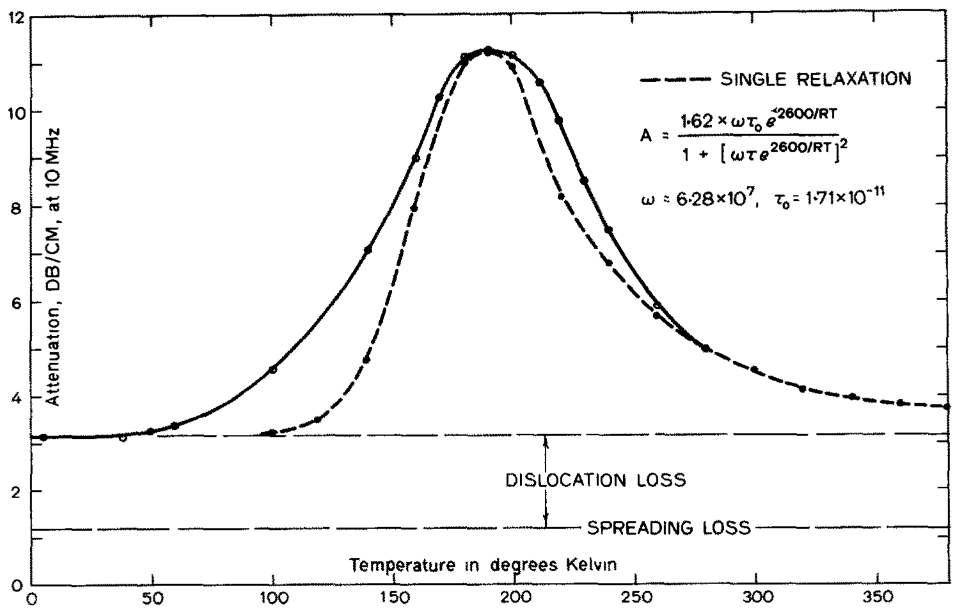

Fig. 6. Analysis of impurity peak in terms of a single activation energy. After subtracting this loss the dislocation component is independent of the temperature.

which may be connected with a partial unpinning. Hence up to 500°K there is a decrease in the internal friction of about 20 per cent. Figure 7 shows the measured variation of the internal friction with temperature for two well annealed samples.

4. THEORETICAL INTERPRETATION

The internal friction due to a drag coefficient proportional to the velocity has been worked out by Oen, Holmes, and Robinson. For a single loop length as would be consistent with the present case, the internal friction from this source will not be more than

$$Q^{-1} = 6\times 10^{-9} \tag{2}$$

which is negligable compared to the measured value of $1{\cdot}5\times 10^{-4}$. Hence another source of dissipation must be invoked which produces an internal friction independent of the frequency and nearly independent of the temperature.

Such a source of internal friction has been discussed previously [2] and has been applied to explaining the low frequency component for metals and for the earth's vibrations which also indicate a constant Q independent of the frequency. This source of dissipation is connected with the motion of dislocation kinks over Peierls-type kink barriers. This motion occurs non-linearly and results in the generation of lattice vibrations which abstract energy from the motion. For the present case with a single kink between pinning points the model is shown by Fig. 8. Here the kink occupies one minimum energy position and to move to the next energy minimum it has to cross the kink barrier of height W. According to calculations given by Seeger and Schiller [9], the energy W and the kink stress σ_k required to take the kink across the barrier are given by

$$W = \frac{\sigma_k b^3}{5};\ \sigma_k = \frac{192}{\pi^2\sqrt{3}}\left(\frac{1-\nu}{1+\nu}\right)\frac{b}{w}\left(\frac{\sigma_p}{\mu}\right)^2\mu \tag{3}$$

where ν is the Poisson's ratio, w is the kink width, σ_p the Peierls stress and μ the shearing modulus. For reasonable values of $b/w \doteq 1$ and $(\sigma_p/\mu) \leqslant 10^{-2}$; it appears that σ_k will not be more than 10^8 d/cm^2 and $W \leqslant 5\times 10^{-16}$ ergs $= 4{\cdot}1\times 10^{-4}$ eV. This is a low enough energy so that it will not affect the internal friction down to 4°K.

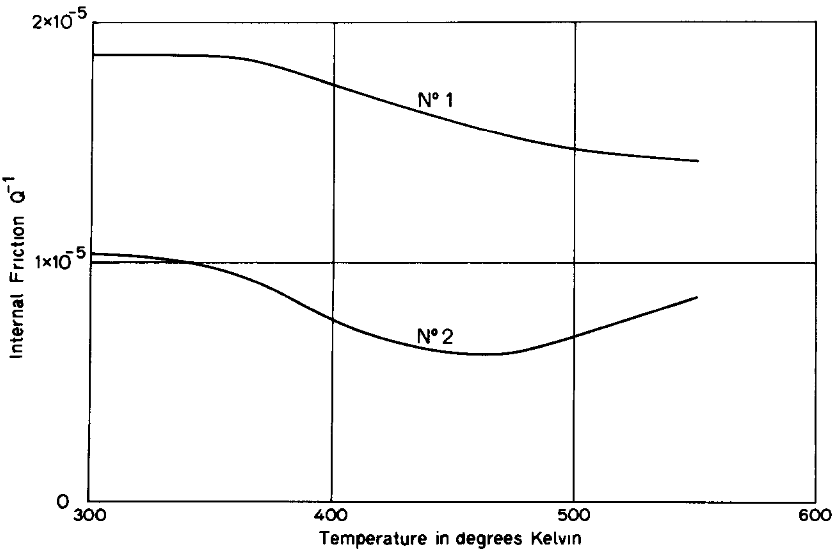

Fig. 7. Internal friction at 17·6 kHz as a function of the temperature.

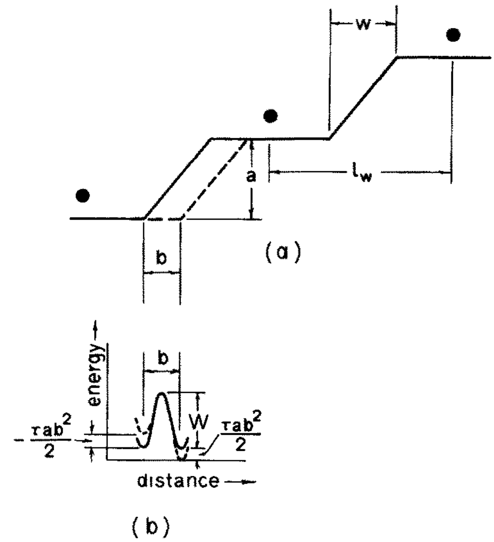

Fig. 8. Dislocation kinks and kink energy barriers. Dashed lines show the effect of an applied shearing stress τ.

For stresses less than the kink stress σ_k, it requires thermal energy to take the kink across the kink barrier. As shown by Fig. 8(b) the effect of a shearing stress τ is to lower one barrier by an amount $\tau ab^2/2$, where a is the height of the kink, and to raise the other barrier by the same amount. If this is the only stress applied, it is well known from reaction rate theory that the net rate of jumps per second is

$$(\alpha_{12}-\alpha_{21}) = \omega_0 e^{-W/kT}\left[e^{\tau ab^2/2kT} - e^{-\tau ab^2/2kT}\right]$$

$$\doteq \omega_0\left(\frac{\tau ab^2}{kT}\right)e^{-W/kT} \qquad (4)$$

since it can be shown that $\tau ab^2/2kT$ is much less than unity even for low temperatures. $\omega_0 = 2\pi$ times the frequency of the kink in the potential well. This is usually taken to be 10^{13} Hz.

When the kink has crossed the barrier it acquires an energy $\sigma_k ab^2/2$ from the negative slope of the potential barrier. If the kink had no interaction with adjacent kinks or pinning points and did not generate any lattice vibrations, this energy would be sufficient to carry it over the next barrier, etc. However Weiner [3] and Atkinson and Cabrera [4] have shown, using a Frenkel–Kontorowa model, that it takes a dynamic Peierls stress τ_{dp} of from 0·01 to 0·1 of the Peierls stress to keep the dislocation moving after it first crosses the barrier. This stress is necessary to replace the energy lost to mechanical lattice vibrations generated when the dislocation moves across the barrier. While the calculation is for a straight dislocation, it has generally been considered [9] that it applies to a kink also.

When the dislocation is pinned and interacts with adjacent kinks, the energy acquired from the negative slope of the barrier is stored in stretching the dislocation in a string model or in pushing the kinks closer together in the kink model. In either case the energy stored is

$$E = n\sigma_k ab^2/2 \qquad (5)$$

while the energy dissipated is

$$\Delta W = n\tau_{dp} ab^2/2 = n\beta\sigma_k ab^2/2 \qquad (6)$$

where β is the ratio of τ_{dp} to σ_k and n is the number of kink displacements. Hence for the dislocation system alone the internal friction Q^{-1}, which is by definition the ratio of the energy dissipated to the energy stored, is

$$Q^{-1} = \beta. \qquad (7)$$

For the complete system we have to consider the energy stored in the elastic part of the crystal and for this case

$$Q^{-1} = \beta\left(\frac{\Delta\mu}{\mu-\Delta\mu}\right) \doteq \frac{\beta\Delta\mu}{\mu} = \beta(\Delta S/S) \qquad (8)$$

where ΔS is the stiffness change for any mode such as the longitudinal mode used. From the measurements of Table 1, it is found that $\beta \doteq$ 0·03 which is intermediate to the two theoretical values. Since the number n of kink jumps is proportional to the applied stress and is equal to the total number n_0 when $\tau = \sigma_k$, then we have the relation

$$n = \tau n_0/\sigma_k. \quad (9)$$

Some idea of the number of kinks moving can be obtained from the following calculation. The total shearing strain in the material is the sum of the elastic strain plus the plastic strain or

$$S = S^E + S^P = S^E + nsb = S^E + nab^2 \quad (10)$$

where n is the number of kinks displaced and s their area equal to ab. But from Equation (9) this can be written

$$S = S^E + \frac{n_0\tau}{\sigma_k}ab^2. \quad (11)$$

Dividing through by the shearing stress τ and collecting terms

$$\frac{\Delta\mu}{\mu} = \frac{\mu^E - \mu}{\mu^E} = \frac{n_0\mu^E ab^2}{\sigma_k}. \quad (12)$$

We found that σ_k may be in the order of 10^8 d/cm^2 and $\mu^E = 4\times 10^{11}$ d/cm^2; $b = 3\times 10^{-8}$ and a is in the same order. For the turned sample $\Delta S/S = 2\times 10^{-2}$. Hence we have

$$2\times 10^{-2} = n_0 \times 4000 \times 27 \times 10^{-24}$$

or

$$n_0 = 1{\cdot}85 \times 10^{17}. \quad (13)$$

If there is on the average of one kink between each pinning point and the separation between points is 10^{-7} cm then the number of dislocations for the strained sample is about $1{\cdot}5\times 10^{10}$. For the annealed sample this might drop to $1{\cdot}5\times 10^9$, which appears to be a reasonable value.

The frequency and temperature variations of the internal friction at low strain amplitudes are also in agreement with this model. Since the energy loss is proportional to the number of kink displacements and for a given strain the number of kink displacements is a constant, the energy loss is proportional to the frequency and the internal friction Q^{-1} is independent of the frequency. The energy loss is proportional to the height of the kink barrier and this should be constant up to higher temperatures when the height may drop somewhat. This is in agreement with the curves of Figs. 6 and 7.

REFERENCES

1. MASON W. P. *Plasticity* (Edited by C. J. McMahon), pp. 287–363. Interscience, New York (1968).
2. MASON W. P., *A Source of Dissipation That Produces An Internal Friction Independent of the Frequency.* Tech. Report No. 61, Institute for the Study of Fatigue and Reliability; also published in *Int. Conf. Fundamentals of Dislocation Theory*, Natl. Bureau of Standards (1969).
3. WEINER J. H., *Phys. Rev.* **136**, A863–868 (1964).
4. ATKINSON W. and CABRERA N., *Phys. Rev.* **138**, A763–766 (1965).
5. COTTRELL A. H., *Dislocations and Plastic Flow In Crystals*, p. 139. Oxford University Press, Oxford (1953).
6. MASON W. P. and WOOD W. A., *J. appl. Phys.* **39**, 5581–5584 (1968).
7. ALEFELD G., *Phil. Mag.* **11**, 809–826 (1965).
8. OEN O. S., HOLMES D. K. and ROBINSON M. T., *U.S. At. Energy Comm. Rep.* No. ORNI-3017, 3.
9. SEEGER A. and SCHILLER P., *Physical Acoustics* (Edited by W. P. Mason), Vol. 3A, Chap. 8, Academic Press, New York (1966).