Further it must be kept in mind that the second order solution given is an equilibrium solution, and does not, therefore, contain any time dependence and rate of approach to equilibrium.

Suzuki, T.: My point of discussion may relate to a sort of recovery which happened at low temperatures in your measurement. We have recently measured the electrical resistivity of edge dislocations introduced into copper and copper alloys by bending at room temperature. Watching the change as a function of time after deformation, the resistivity due to the stacking-fault goes through a maximum and the resistivity of the matrix goes through a minimum. Such a phenomenon may suggest the change in the configuration of point defects in the vicinity of the dislocation. Temperature variation of a kind of the atmosphere which is built after deformation or any other means may be possible. The above experiment will be reported in the Kyoto Conference.

Thompson, D. O.: Your experiments sound very interesting, and I will look forward to hearing more of your results.

Lücke, K.: I would like to ask you to comment on the break away behavior observed in your measurments.

Thompson, D. O.: The only point I wish to make at the present time with respect to the amplitude dependence of the decrement is as follows: at small doses one finds that as the amplitude independent decrement is suppressed with increasing temperature, the strain at which apparent breakaway occurs diminishes. This result is at least qualitatively consistent with the second order behavior of the pinning point given earlier. Thus these data may indicate that, as the temperature is raised from 200°K, a few harder pinning points (divacancies) are transformed to a greater number of relatively weaker pinning points.

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Internal Friction due to Diffusion of Dislocation-Kinks Trapped by Point Defects

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It was proved theoretically that the diffusion of dislocation-kinks trapped by point defects give rise to an internal friction peak of relaxation type. This peak is considered to be the one which appears in cold-worked metals above Bordoni peak temperature.

1. Introduction

Internal friction peaks as measured as a function of temperature, which appear after cold-working, were observed in our laboratory many years ago[1)-6)]. Several investigators have also observed the similar peaks[7-13)]. The most important characteristics of these peaks are as follows: (1) The peaks appear at higher temperatures as compared with Bordoni peak. (2) Activation energies obtained from peak shifts are larger than those for Bordoni peaks. They are usually in the range from 0.2 to 0.6 eV in many metals[6)]. (3) The peaks anneal out when the sample is annealed at appropriate temperatures, at which Bordoni peaks do not anneal out. In other words, these peaks are less stable than Bordoni peaks, when annealed.

It was suggested that dislocations associated with point defects may give rise to this kind of peaks[5),6)], and a possible behavior of point-defect-associated dislocations was described elsewhere in some detail[14]. Recently an additional support was found in our laboratory for the view that the peak is caused by a combined effect of dislocations and point defects[15]. It was observed in a low frequency experiment on copper that a once appeared peak almost disappears when the specimen is slightly cold-worked at liquid nitrogen temperature, *e.g.* about 0.35% surface strain in twisting, and that peak appears again after aging the specimen at or a little above room temperature[15]. This is interpreted as follows. The dislocations are pulled off the pinning points by slight cold-working, resulting in free dislocations and free point defects, which do not contribute to the peak. During the aging at or a little above room temperature, free point defects migrate to dislocations to pin down them again, and/or free dislocations move for short distances to adjust themselves and come to point defects to be pinned down again.

There is, however, hitherto no theory which leads from a pinned down dislocation model to the relaxation type internal friction peak, which is expressed by

$$Q^{-1}=\Delta\tau\omega/(1+\tau^2\omega^2) \tag{1}$$

and

$$\tau=\tau_0 \exp(E/kT). \tag{2}$$

The objective of this paper are to derive the equations (1) and (2) from a specific model, and to discuss the expressions for relaxation strength Δ and pre-exponential factor $^1/\tau_0$.

2. The Model

Let us consider a dislocation line between two nodes. If the direction of the dislocation line is, in average, not parallel to the Peierls potential valley, it will consist of a number of dislocation kinks and a number of dislocation segments, which lie along Peierls valleys, as is shown in Fig. 1 (a).

We consider that kinks make diffusion by virtue of thermal fluctuation of local stresses,

Fig. 1 (a). A dislocation configuration with kinks which is considered to contribute to the internal friction peak. The point defects, which are distributed along and around the dislocation, are omitted in this figure.

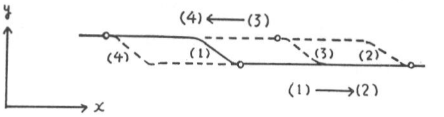

Fig. 1 (b). A part of the dislocation line shown in Fig. 1 (a) with a kink. The successive movements are shown by dotted lines and the numbers 1, 2, 3 and 4. The distribution of point defects along and around the dislocation is idealized.

as was discussed in detail by Lothe and Hirth[16]. The kink diffusion was also discussed by Brailsford[17], who proposed abrupt kinks. We do not, however, consider abrupt kinks in this paper. According to Lothe and Hirth, the kink width may be considered to be about $10a$, where a is the distance between neighboring Peierls valleys. If a kink has a width of this size, crystalline resistance to motion of a kink is negligibly small at temperatures of our interest.

Now we shall consider the trapping of a kink by point defects. A kink, which is trapped by a point defect on one side, can diffuse freely to its free side, but cannot diffuse to its trapped side, unless it breaks away from its pinning defect.

The diffusion of trapped kinks can be treated in a similar way as the diffusion of trapped point defects[18], *i.e.* the activation energy of diffusion of trapped kink is expressed in a simple case as the sum of the activation energy of free kink diffusion and the binding energy between a point defect and a dislocation. The activation energy of free kink diffusion can be neglected in our case.

We can consider a number of variations of models, which are the same in principle. We shall present here an idealized example in Fig. 1 (b), which represents a part of Fig. 1 (a). Different variations come mainly from the different distributions of point defects. The distribution of point defects around a dislocation line is considered to be random.

A kink may be situated at a given time at a position (1) in Fig. 1 (b). This kink is trapped on its right-hand side. If a stress is applied in such a way as to push the kink to its right-hand side, and if the thermal energy detrap the kink from the pinning point defect, the kink will move to the position denoted as (2). When the direction of the stress is reversed, the kink will move to the position

(3) without thermal activation. If it is thermally depinned at the position (3), it diffuses to the position (4). These are the examples of diffusion processes of trapped kinks under cyclic stresses.

3. The Theory

The essential part of the mathematical treatments in this theory consists of a modification of Brailsford's treatment of abrupt kink theory, although we do not consider abrupt kinks.

As was mentioned in the preceding section, trapped kinks make diffusion by virtue of thermal fluctuations of local stress. If an external stress is applied to the crystal, kinks make drift according to Einstein relation, *i.e.*

$$v=\sigma ba\mu \tag{3}$$

and

$$\mu=D/kT\ , \tag{4}$$

where v is the velocity of trapped kink along dislocation line, σ is applied external stress, b is the magnitude of Burgers vector, μ is the mobility of trapped kink, D is the diffusion coefficient of trapped kink, and k and T have the usual meanings.

The dislocation shear strain ε_d produced by a dislocation line of length L between two nodes or two permanent pinning points in a crystal of volume V is

$$\varepsilon_d=\frac{b}{V}\int_0^L(y-y_0)dx\ , \tag{5}$$

where the coordinate is taken as shown in Fig. 1 (b). y_0 is the position of the dislocation line, from which the displacements are measured. This is determined by the line energy and the two dislocation nodes or the two permanent pinning points.

Then the rate of displacement of a dislocation line $\partial y/\partial t$ is derived from the diffusion and the drift of trapped kinks as follows:

$$\frac{\partial y}{\partial t}=D\frac{\partial^2 y}{\partial x^2}+\sigma ba^2N\mu, \tag{6}$$

where N is the density of trapped kinks. This equation is essentially the same as that derived by Brailsford in his theory of abrupt kink diffusion, although the physical meanings are different in many respects here. Putting

$$\sigma=\sigma_0\exp(i\omega t), \tag{7}$$

the equation (6) can be easily solved. Here σ_0 and ω are the amplitude and the angular frequency of vibrational external stress, respectively.

Now we shall omit to describe the detail of calculation. Finally we obtain the dislocation strain as follows:

$$\varepsilon_d=\left(\frac{2by_1L}{\pi V}\right)\frac{\exp(i\omega t)}{(i\omega\tau_L+1)}, \tag{8}$$

where

$$y_1=4\sigma_0 ba^2L^2N/kT\pi^3 \tag{9}$$

and

$$\tau_L=L^2/D\pi^2. \tag{10}$$

From imaginary and real parts of dislocation strain, the internal friction and the modulus defects are easily derived, namely

$$Q^{-1}=\varepsilon_d{}''G/\sigma_0=\Delta\omega\tau_L/(1+\omega^2\tau_L{}^2) \tag{11}$$

and

$$\Delta G/G=\varepsilon_d{}'G/\sigma_0=\Delta/(1+\omega^2\tau_L{}^2), \tag{12}$$

where G is the shear modulus.

Now we could obtain precisely the same equation as the equation (1). In other words, our model of diffusion of trapped kinks leads theoretically to an internal friction peak of relaxation type.

The relaxation strength in the equations (11) and (12) is expressed as

$$\Delta=8N\rho Ga^2b^2L^2/\pi^4kT, \tag{13}$$

where ρ is the dislocation density. The pre-exponential factor $1/\tau_0$ in the equation (2) is given by

$$\tau_0=L^2/\pi^2\nu\bar{\Lambda}^2, \tag{14}$$

where $\nu\approx10^{13}$, and $\bar{\Lambda}$ is an average length between traps or an average distance which a kink travels by a single detrapping activation. The activation energy in the equation (2) is the same as the binding energy between a point defect and a dislocation.

4. Discussions

In our theory the activation energy in the equation (2) is the above mentioned binding energy. This is reasonable, because its observed values are in the range from 0.2 to 0.6 eV in many metals as were mentioned in Section 1.

If we assume $L\approx10\bar{\Lambda}$, then

$$f_0=1/2\pi\tau_0\approx10^{11}$$

which is in good agreement with some observed values[6]. However, observed values of f_0 sometimes scatter widely as the experimental condition differs. This may be understood as due to the wide variation of the ratio $L/\bar{\Lambda}$.

The calculated value of relaxation strength is about ten times larger than observed values. This is not surprising, because it is considered that only a fraction, say one tenth, of total length of dislocation lines in the crystal contributes to give rise to this kind of peaks. In other words, only favorable configurations of point-defect-associated dislocations may contribute to cause the peak. For example, a dislocation line without intrinsic kinks, which lies along a Peierls potential valley, may be ruled out of the favorable configurations. It is interesting, in this connection, to point out that the Bordoni peak grows when our peaks anneal out (See Fig. 4 in reference (6)). This is interpreted as follows. An oblique dislocation line with a number of kinks, which contributes to give rise to our peak, anneals to become a dislocation line parallel to Peierls valley, which contributes to cause the Bordoni peak.

Finally it should be added here, that this theory might also explain the Köster peak, if the model is a little modified.

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DISCUSSION

Thompson, D. O.: 1) What is the fraction of the dislocation population which contributes to your mechanism? 2) Are these dislocation related to those which contribute to the Bordoni mechanism?

Hasiguti, R. R.: 1) Under an appropriate assumption it was found that about one tenth of total dislocation contribute to our mechanism. I think this is not unreasonable. 2) I think the dislocations are rather different from those which contribute to Bordoni mechanism, although it is possible that ours change into Bordoni's and *vice versa*. This is supported experimentally by our result shown in Fig. 4 in our paper published in Acta Met. 10 (1962) 442. This figure shows that Bordoni peak (large peak in the figure) grows, when our two small peaks anneal out. This is interpreted that dislocation configuration changes on annealing. This kind of experiments were done more in some detail by Okuda, one of our colleagues, while he was at the Rice University, U.S.A. I should like also to add that the dislocation model, which I adopted in my theory, is definitely different from that which contribute to Bordoni peak, if we adopt Seeger's theory as the Bordoni mechanism. In other words, our dislocation lies in average with a finite angle to the Peierls valley, while Seeger's dislocation lies parallel to Peierls valley.

Suzuki, T.: I think we should be more careful in using the Brailsford's analysis, because he neglected the fact that some correlated motion of kinks is necessary under the assumption that the total length of dislocation must be conserved.

Hasiguti, R. R.: I know that our treatment is simplified. But I think this is still useful as the first step of this kind of analysis.

Nowick, A.: I would like to suggest that an estimate of the fraction of dislocations which contribute to your peak can be obtained by comparing the modulus defect due to dislocation at room temperature (which can be attributed to all the dislocations) to the relaxation strength of the peak itself. I suspect that this method would show that considerably less than 0.1 of the dislocations contribute to the peak.

Hasiguti, R. R.: Thank you for your suggestion. We have some data on modulus defect, but at present our data are not sufficient nor suitable for this kind of analysis, which you suggested.

Thompson, D. O. (Comment to A. S. Nowick): Measurements of the modulus defect in a copper crystal yields, typically, a value of about 1 in 50 at room temperature, and about 1 in 500 at 10°K. This indicates that a sizable fraction of the dislocations experience a thermally activated motion. However, it is not clear that one can compare numbers directly to obtain a ratio, for it is probably true that the average dislocation free length is different at the two temperature.

Fujita, F. E.: If the number of point defects, which trap the kinks, increases, or the temperature increases, or the stress amplitude is increased, the number of detrapping will be not only one but many. Is this point already taken into account in your theory? On the other hand, if the number of point defects is small, there will be many kinks between two trapping point defects. How did you treat this point in the present theory?

Hasiguti, R. R.: I do not say that detrapping occurs only at one point. The first part of your comments is already included in my theory, although we have to exclude the case where the pinning point defects are too dense to apply our theory without making any modification. As to the second part of your comments, it is easily shown that a free kink, *i.e.* a kink which is not trapped by a point defect, do not contribute to our peak. Although it was in a little different manner, it was already shown in my talk and in the preprint.

Seeger, A.: Would Prof. Hasiguti please indicate how his theory could be extended to cover the Köster or cold-work peak in b. c. c. metals?

Hasiguti, R. R.: I have written in the preprint that the theory may be extended to cover the Köster or cold-work peak in b. c. c. metals. However, it seems difficult to do so in the present form. I believe that the major modification, which must be done to extend the theory, is in the distribution of impurity atoms which interact with dislocations. For example, Professor Wert presented his work this morning as paper No. III-9,* which shows that impurity atoms form clusters at dislocation lines in niobium. These kinds of experimental results are being accumulated now. At this moment I think I should wait for a little more experiments before we extend our theory.

* *Proc. Int. Conf. Cryst. Latt. Def.* (*1962*): J. Phys. Soc. Japan **18** Suppl. I (1963) 141.