

DISLOCATION PIPE DIFFUSION*

G. R. LOVE†

A model is developed which attributes the observed enhanced diffusivity in dislocation arrays entirely to processes occurring in the dislocation core. If an atom is transferred from the edge of the inserted plane in an edge dislocation into the row of sites into which that plane might grow by negative climb, this creates a "vacancy" in the edge of the inserted plane and an "interstitial" in the line of sites adjacent to that plane. Motion of these defects along the dislocation core would give rise to a net flux parallel to the dislocation line. The effects of correlated motion of the defects (cf. Lothe⁽⁶⁾), are discussed and it is demonstrated that motion of the interstitial is essentially uncorrelated in an edge dislocation. The model is consistent with available data on diffusion in edge dislocations in silver, accounts adequately for the observed difference between edge and screw dislocation diffusivities, and leads to qualitative predictions regarding the diffusivity along extended dislocations and solute diffusion in dislocations.

DIFFUSION PAR LE CANAL DES DISLOCATIONS

L'auteur établit un modèle qui attribue la diffusivité accrue qui survient en pratique lorsqu'il existe des réseaux de dislocations, à un mécanisme qui fait intervenir le centre de la dislocation.

Si un atome est déplacé de l'extrémité du plan supplémentaire d'une dislocation-coin vers la rangée de sites qui correspondrait à la croissance de ce plan par montée négative, il apparaît "une lacune" à l'extrémité du plan supplémentaire et un "interstitiel" dans la ligne de sites adjacents à ce plan.

Le mouvement de ces défauts donnera lieu à un flux parallèle à la ligne de dislocation. Les effets associés au mouvement de ces défauts (cfr Lothe⁽⁶⁾) sont discutés et on montre que le mouvement des interstitiels est essentiellement irrégulier dans une dislocation-coin.

Le modèle proposé est en bon accord avec les résultats expérimentaux de diffusion dans l'argent contenant des dislocations-coin. Il rend compte également de manière correcte de la différence observée entre diffusivités correspondant à des dislocations-coin ou vis et conduit enfin à des prévisions qualitatives concernant la diffusivité le long de dislocations étendues et la diffusion d'atomes solutés le long des dislocations.

VERSETZUNGEN ALS DIFFUSIONSKANÄLE

Das hier entwickelte Modell führt die beobachtete Erhöhung der Diffusion in Versetzungsanordnungen vollständig auf Prozesse zurück, die im Versetzungskern ablaufen. Wird ein Atom von der Kante der eingefügten Ebene einer Stufenversetzung in die Gitterreihe übertragen, in die diese Ebene bei negativem Klettern hineinwachsen würde, so wird eine "Leerstelle" in der Kante der eingefügten Ebene und ein "Zwischengitteratom" in der dieser Ebene benachbarten Gitterreihe erzeugt. Bewegung dieser Fehlstellen entlang dem Versetzungskern ergibt einen Materiestrom parallel zur Versetzung. Der Einfluß korrelierter Bewegung der Fehlstellen (vgl. Lothe⁽⁶⁾) wird diskutiert, es wird gezeigt, daß die Bewegung des Zwischengitteratoms bei einer Stufenversetzung im wesentlichen unkorreliert verläuft. Das Modell steht in Übereinstimmung mit vorhandenen Messungen der Diffusion in Stufenversetzungen in Silber und trägt insbesondere dem beobachteten Unterschied der Diffusion längs Stufen- und Schraubenversetzungen Rechnung. Weiter führt das Modell zu qualitativen Voraussagen über die Diffusion längs aufgespaltenen Versetzungen sowie über die Diffusion von Verunreinigungen in Versetzungen.

INTRODUCTION

It has been suggested that the observed low-temperature lattice diffusivity is greater than that predicted by extrapolation of high-temperature data because of rapid diffusion along dislocations.⁽¹⁾ Enhanced low-temperature diffusion in the vicinity of grain boundaries has been explained in terms of an assumed dislocation structure of the boundaries and a high intrinsic diffusivity along the dislocations.^(2,3) In low-angle polygonization boundaries, whose dislocation structure is directly observable, high diffusivity along the component dislocations has been observed.⁽⁴⁾ Many discussions of annealing-out of quenched in vacancies, the growth and destruction of Kirkendall voids, and several of the treatments

of dislocation climb have presupposed high mobility of atoms and vacancies in the dislocation vicinity or on the dislocation core.^(5,6) A number of the characteristics of dislocation diffusion have been reported and a number of limiting conditions upon any model of the diffusion process have been presented.^(2,6,7) Nevertheless, despite the ubiquity of diffusion along dislocations in metallurgical phenomena, no serious attempt has been made to present a diffusion model which is both compatible with the theoretical observations on the process and capable of explaining the available experimental results. The present paper presents a semi-quantitative model of diffusion along pure edge dislocations and qualitatively extends this model to a general description of diffusion along dislocations.

BACKGROUND

Due to the small physical dimensions of a dislocation, experimental investigations of diffusion along dislocations are restricted to investigations of

* Received May 24, 1963.

This paper is based on a thesis submitted in partial fulfillment of the requirements for a Ph.D. degree in Metallurgy at the Carnegie Institute of Technology, Pittsburgh 13, Pa.

† Presently in the Metallurgy Department, Oak Ridge National Laboratories, Oak Ridge, Tennessee.

aggregates of dislocations. Although Hart⁽¹⁾ has discussed the effect of an aggregate of dislocations of random orientation and high diffusivity on the apparent low-temperature lattice diffusivity, it is not possible, at present, to determine the dislocation diffusivity from variations in the apparent lattice diffusivity. This approach to determining dislocation diffusivities is also undesirable because, by the nature of a random array, it yields only an average dislocation diffusivity and is insensitive to edge or screw character of the dislocations. It has been demonstrated repeatedly that low-angle boundaries are dislocation arrays,⁽⁸⁾ and it is possible to define the kind and density of dislocations in these arrays accurately. Unfortunately, the techniques for verifying the dislocation nature of small-angle boundaries are not capable of adequate resolution to apply to intermediate-angle boundaries and, further, preferential diffusion down very low angle boundaries is not, in general, observable.^(2,3,9) Fisher⁽¹⁰⁾ has shown that the ratio of boundary diffusivity to lattice diffusivity must be at least 10^4 if observable enhanced penetration is to occur and this ratio is normally obtainable only at relatively high dislocation densities (boundary misorientations greater than about 10°).

Leymonie *et al.*⁽⁴⁾ have actually observed enhanced diffusion along dislocations in low-angle polygonization boundaries in iron. They do not report specific values for the diffusivity along these dislocations, but their observation is taken to be proof that a dislocation can be a diffusion "short circuit". Specific values for the diffusivity along dislocations have been obtained by Hoffman and Turnbull⁽²⁾ by measuring the orientation dependence of the boundary diffusivity in intermediate and high angle tilt boundaries in silver and assuming these boundaries to be dislocation arrays. These authors report that the intrinsic diffusivity along an edge dislocation in silver is described by an Arrhenius equation with a pre-exponent comparable to that for lattice diffusion and an activation energy about half that for lattice diffusion. Recent work by Shewmon and Love⁽³⁾ in the same system indicates significantly higher diffusivities along edge dislocations than along screw dislocations and earlier work by Achter and Smoluchowski⁽¹¹⁾ also qualitatively notes this difference in diffusivities.

There are two possible routes to an explanation of the large effect of dislocations on diffusion at low temperatures. One might consider the effect of interactions of the dislocation with its environment or one might consider the intrinsic properties of the dislocation itself. Qualitative consideration of the first of these alternatives has been the more common. For

example, knowing that the strain field of an edge dislocation has a dilatational component, Hoffman and Turnbull postulated that there was marked segregation of vacancies to the edge dislocation vicinity and that this segregation made an empty lattice site (vacancy) available to any atom attaining sufficient energy to make a diffusion jump. The observed activation energy for edge dislocation diffusion would then no longer contain the energy of formation of a vacancy. It has also been suggested, indirectly, that the presence of a strained lattice region of any kind would accelerate diffusion.⁽¹²⁾ These proposals refer to relatively long-range effects of dislocations and this, it appears, is their greatest weakness. Hoffman and Turnbull⁽¹³⁾ have shown that the calculated "pipe" diffusivity or diffusivity per dislocation is quite constant up to boundary misorientations approaching 28° although the dislocation separation in such boundaries is comparable to the lattice parameter. At such dislocation densities, relatively long-range interactions of the dislocation with its environment would appear unlikely. Indeed, Hoffman and Turnbull assume an effective dislocation diameter of only 5 Å.

Alternately, the dislocation core might be assumed to have unique properties leading to rapid diffusion. The term "pipe" diffusion, which has frequently been applied to diffusion along dislocations, might well have been coined to represent the line of vacant sites lying adjacent to the edge of the inserted plane in an edge dislocation. Evidently, this is an oversimplification; nevertheless, it indicates a fruitful line of inquiry into the mechanics of diffusion down dislocations.

THE MODEL

Lothe⁽⁶⁾ has indicated that there must be a minimum of two high-diffusivity paths in the dislocation core in order that diffusion along the core not be inhibited by correlation effects. The "core" of the dislocation, therefore, may be defined as the last line of filled sites in the inserted plane of an edge dislocation together with the line of vacant sites into which that plane would grow by negative climb. This is illustrated for a simple cubic lattice in Fig. 1. An "interstitial" in the dislocation core may be defined as an atom in the row of vacant sites in the core; a "vacancy", as an empty site in the line of atoms in the core (Fig. 2).

If it is assumed that the interchange of atoms or vacancies with the lattice is relatively infrequent (local equilibrium against climb in either direction), interstitials and vacancies are created in pairs and annihilate one another on contact. Each time an

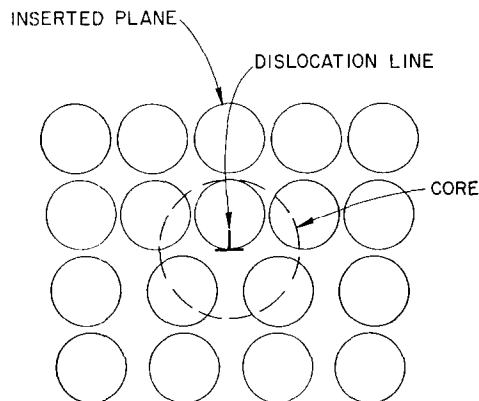


FIG. 1. Pure edge dislocation in simple cubic lattice.

interstitial atom recombines with a vacancy other than the one created simultaneously with it, or each time net motion of a vacancy-interstitial pair occurs between creation and annihilation, diffusion takes place.

Although defects in the dislocation core are assumed to have high mobility along the dislocation line,⁽⁵⁾ not all atomic motion produces net diffusion. The density of defects of either kind is given by

$$\rho a^N \sim \exp(-\Delta G_a/RT)$$

where the free energy of formation of the a^{th} defect is ΔG_a . Since the free energy of formation of an interstitial is at least as high (probably higher) than that of a vacancy, on the average, there will be no more than one interstitial in the dislocation core between any two vacancies. A given interstitial will encounter no other interstitial atom along its diffusion path and the correlation discussed by Lothe⁽⁶⁾ will not limit diffusion if this model adequately describes the process. A second kind of correlated motion can occur. When an interstitial atom recombines with its "parent" vacancy without any net motion of the vacancy-interstitial pair, no net diffusion results. This kind of correlation (the kind normally encountered in lattice diffusion by a vacancy mechanism) can be shown to be relatively insignificant.

Similarly, the motion of vacancies themselves is

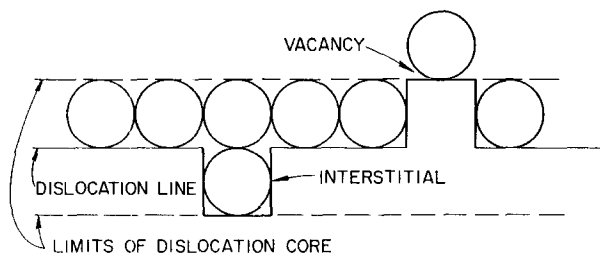


FIG. 2. Illustrating "interstitial" and "vacancy" in pure edge dislocation (simple cubic lattice).

essentially uncorrelated. However, the atomic movements produced by vacancy motion are completely correlated. There is but a single row of sites in which the vacancy has high mobility and the passing of a vacancy down this row of sites has no effect on the sequence of atoms in the row. When a second vacancy travels along the row in the reverse direction, it replaces all atoms in their original position. Under isothermal conditions a net flux of vacancies will not occur; therefore, no diffusion due to vacancy motion will occur. The only method for changing the sequence of atoms along the core involves displacing an atom into an interstitial position and this may be accounted for simply by considering the motion of interstitials. This motion can be described quantitatively by applying the analysis developed by Zener (reported by Shewmon⁽¹⁴⁾) to the particular conditions which have been described for the dislocation core configuration.

Consider a plane constructed perpendicular to the dislocation line and not coincident with a rest position of an interstitial atom in the dislocation. The probability of an atom jumping from left to right across this plane (alternately, the fractional number of jumps per unit time across similar planes on a large number of dislocations) can be written

$$J_{1-2} = aC_1\Gamma_T \quad (1)$$

where J_{1-2} is the flux from left to right across the plane, Γ_T is the jump frequency of an atom in an interstitial position, and aC_1 is the probability that the interstitial site to the left of the reference plane is occupied by a tracer atom (alternately, the number of interstitial tracer atoms in a slab of thickness a to the left of the reference plane if the concentration of interstitial tracer atoms is C_1). The probability of a jump in the reverse direction across the plane is then given by

$$J_{2-1} = a\Gamma_T \left(C_1 + a \frac{\partial C_1}{\partial x} \right) \quad (2)$$

where the distance between rest positions of the interstitial atom is assumed to be a , the lattice parameter, and $\partial C_1/\partial x$ is the local gradient in interstitial tracer atoms. If the free energy required to create an interstitial atom in the dislocation core is ΔG_i , then the probability of a given site in the core containing an interstitial atom is

$$p = \exp(-\Delta G_i/RT) \quad (3)$$

The probability that any occupied site contains a tracer atom is simply the local concentration of tracer atoms, C_0 . The combined probability of an interstitial site containing an atom which is also a tracer

is, then, the product

$$C_1 = C_0 \exp -\Delta G_i/RT \quad (4)$$

The net tracer flux can be obtained from equations (1), (2) and (4) and is given by

$$J_{\text{net}} = -a^2 \Gamma_T \exp(-\Delta G_i/RT) \frac{\partial C_0}{\partial x} \quad (5)$$

Comparison of equation (5) with Fick's first law indicates that

$$D = a^2 \Gamma_T \exp(-\Delta G_i/RT) \quad (6)$$

Further, the jump frequency, Γ_T , is simply the product of the vibration frequency of an interstitial atom on the dislocation core, ν , and the probability that the atom will acquire sufficient energy to cross the barrier between neighboring sites. The vibration frequency ν , is not, in general, equal to the vibration frequency of an atom on a lattice site; however, it should not differ greatly from that frequency. If the barrier between neighboring sites has height ΔG_m , the jump frequency becomes

$$\Gamma_T = \nu \exp(-\Delta G_m/RT) \quad (7)$$

and equation (6) may be written

$$D = a^2 \nu \exp-(\Delta G_i + \Delta G_m)/RT \quad (8)$$

or, in conventional notation

$$D_0 = a^2 \nu \exp \frac{(\Delta S_i + \Delta S_m)}{R}; \quad Q = \Delta H_i + \Delta H_m \quad (9)$$

A major assumption is implicit in the development leading to equation (8). It is assumed that, as soon as an interstitial atom is created and separated from its "parent" vacancy by a single atomic distance, it no longer interacts with that vacancy. Thus, the correlation resulting from recombination of a vacancy-interstitial pair without their net motion is neglected in this development.

DISCUSSION

There is marked similarity between equations (8) and (9) and the Zener expression for lattice diffusion by a vacancy mechanism. Much of this resemblance is superficial; although it is possible to compare the pre-exponential term in equation (8) with the constant in the expression obtained by Zener, the physical interpretation of the terms in the activation energy is both quite different and rather less amenable to independent measurement.

The pre-exponential term is identical in form to that obtained by Zener but for a constant which, in his

development, related the number of sites in a given plane into which an atom from a particular site in a neighboring plane can jump. Since diffusion down the edge dislocation core is true uniaxial diffusion, the constant is identically one in equation (8). Further, deliberate omissions in this development, the use of the lattice parameter a rather than other estimates of the jump distance, the use of a to represent the thickness of the source of diffusing atoms, the failure to define the vibration frequency accurately, are all of the same kind and magnitude as the corresponding approximations in the Zener development of the lattice diffusion expression. Since the commonly observed values of the pre-exponent in lattice diffusion in fcc lattices is of the order unity, one might expect the pre-exponential term in equation (8) or (9) to be of the order unity.

The terms in the activation energy of equation (8) are not derived from the corresponding lattice diffusion process. In particular, they are not equal, other than coincidentally, to the activation energy for lattice diffusion less the activation energy for vacancy formation as has been suggested by Hoffman and Turnbull⁽²⁾. Since there are numerous indications that the mobility of an atom on the dislocation line is quite high, it may be assumed that U_m is quite low. The dominant term in the activation energy for the process is then the energy required to form an interstitial atom on the dislocation core. Unfortunately, independent estimates or measurements of this energy are not available. One might reasonably estimate extreme values for it: it should be larger than the energy of a single climb jog in the dislocation line (see Friedel⁽⁷⁾) by up to a factor of two, and it should be smaller than the energy of an interstitial solvent atom in the lattice by roughly the (rather large) interaction energy between such an interstitial and the dislocation. The activation energy for silver self-diffusion along edge dislocation lines determined by Hoffman and Turnbull⁽²⁾ is certainly within these limits. Further, the pre-exponential term reported in the same investigation is within the range of values commonly reported for self-diffusion in f.c.c. materials (10^{-2} to 10) and is therefore quite consistent with this model of the diffusion system.

APPLICATIONS

The primary usefulness of the present model is in its application to qualitative prediction of the variation to be observed in dislocation diffusion under varying experimental conditions. The following paragraphs demonstrate this utility by predicting the variation in dislocation diffusivity to be expected when an

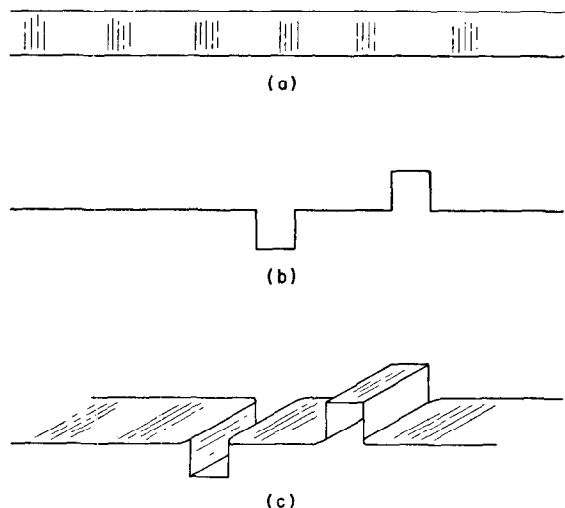


FIG. 3. Schematic: the formation of "vacancy" and "interstitial" defects on an extended dislocation (changes in configuration of the stacking fault plane).

edge dislocation dissociates, when the dislocation is pure screw in character, when a dislocation is randomly oriented in the lattice rather than part of a regular array, and when the diffusing atom is a solute rather than a solvent atom.

The particular geometry of the "interstitial" atom in the dislocation core when the dislocation dissociates can most easily be visualized with the help of the thought experiment illustrated in Fig. 3. Consider a straight planar extended dislocation (Fig. 3(a)); compress it along its glide plane to a total dislocation; create a vacancy-interstitial pair in the total dislocation (Fig. 3(b)); and then allow the total to dissociate into its component partials (Fig. 3(c)). Following Friedel's concept of an extended jog,⁽⁷⁾ it is probable that the interstitial and the vacancy would also at least partially "extend", i.e. their exact atomic position would be undetermined within the stacking fault width. Since the area of stacking fault is increased by the presence of the defects and since, if they do not "extend" as completely as the dislocation line itself, defects introduce curvature into the component partial dislocations in low-stacking-fault energy materials (relative to a total dislocation in a similar material). On the other hand, neither the defect nor the dislocation itself would dissociate with an increase in the total energy, so the change in the defect energy due to dissociation should be small. This model, then, predicts that diffusion along dislocations in low stacking fault materials should show activation energies of the order $1/2$ the activation energy observed for lattice diffusion (similar to those observed for silver self-diffusion along dislocations). On the other hand, the formation or the motion of one of these extended

defects would require the cooperative motion of a number of atoms (roughly, the stacking fault width at the defect divided by the atomic diameter) and such cooperative motion indicates an appreciable negative activation entropy for the process. Pound *et al.*,⁽¹⁵⁾ among others, have shown that requiring cooperative atomic motion between a number of atoms in a diffusion process greatly reduces the pre-exponential (D_0) term in the Arrhenius expression. The model also predicts, then, pre-exponential terms which are significantly smaller in low stacking fault energy materials than those observed in relatively high stacking fault energy materials or in lattice diffusion. Quite recently, Gertsriken and Revo⁽¹⁶⁾ have investigated grain-boundary diffusion of Sn and Sb in Cu and in alloys of Cu + 0.1% Sn and Cu + 0.1% Sb. They report that 0.1% Sn in solution in Cu lowers the apparent D_0 for Sn diffusion in the grain boundary by 50% but introduces only very small (0.13%) changes in the activation energy; 0.1% Sb lowers the apparent D_0 by 13% with no change in the apparent activation energy. Since both these solute elements lower the stacking fault energy of Cu, this may be taken to be limited confirmation of this prediction of the model. If, in fact, larger additions of these solutes continued to reduce the apparent D_0 , this prediction might be considered verified.

Even qualitative discussion of diffusion down pure screw dislocations is made more difficult by the necessity for visualizing the "three-dimensional" nature of the defect in the dislocation core. In Fig. 2, defects in an edge dislocation were represented schematically by a "double-jog" of atomic dimensions in the dislocation line. Following the same convention, point defects in the screw dislocation core may be represented as single-turn helices of atomic dimensions; the "pitch" of the helix would be positive (relative to the dislocation Burgers vector) for a vacancy and negative for an interstitial, as shown in Fig. 4(a). It should be apparent that these helical turns obey the same general rules of behavior outlined for point defects on an edge dislocation line; they are mobile along the dislocation line and may be created and destroyed by pairs. If this were an adequate description of the defect on the dislocation line, one might expect the expressions for screw dislocation diffusivity to be essentially identical to the expressions for edge dislocation diffusivity derived above. This is not the case. A more careful examination of the defects in a pure screw dislocation reveals why.

Seeger and Schiller⁽¹⁷⁾ have shown that coplanar glissile jogs in a dislocation line repel one another, and a number of authors have reported that when

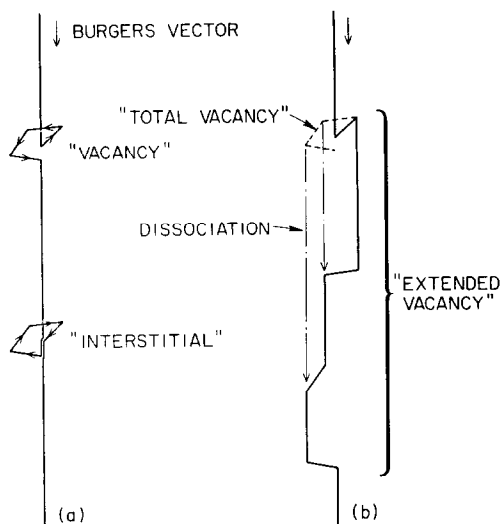


FIG. 4. Schematic: formation of defects on total pure screw dislocation (a) and dissociation to form "extended" defects (b).

point defects (normally vacancies) precipitate on a screw dislocation line, the line becomes a continuous helix. Both these observations indicate that it is quite probable that a point defect dissociates *along* the screw dislocation line even when the screw dislocation itself remains a total dislocation (see Fig. 4(b)). There are two consequences of such dissociation which are particularly discouraging to diffusion down screw dislocations.

First, motion of an extended defect (which may be identified with the extended helix) involves minor but cooperative motion of a large number of atoms. As in the case of diffusion along an extended edge dislocation, this implies a small change in the activation energy but a significant negative entropy change accompanied by a significant decrease in the pre-exponent, D_0 , of the diffusion expression. Second, and more important, although the presence of a single helical turn may be identified with the presence of a single point defect, it is not possible to identify that particular point defect which produces the extended helix. In particular, motion of the helix along the dislocation core may not correspond to motion of a particular atom along a series of interstitial positions but may represent a succession of atoms being displaced, one at a time, into the dislocation core by an "interstitialcy" mechanism (an interstitial atom moving into an occupied atomic site by displacing that atom into the following interstitial site⁽¹⁴⁾). If this interstitialcy mechanism is operating, neither the positive helix (vacancy) nor the negative helix (interstitial), regardless of their mobility, would have any effect on the sequence of atoms along the dislocation core, all atomic motion within an

individual screw dislocation would be completely correlated, and no net flux of tracer atoms along that dislocation would occur.

Nevertheless, appreciable net flux of tracer atoms has been observed along screw dislocations.⁽³⁾ This apparent inconsistency is resolved if one recalls the necessary experimental limitations on dislocation diffusion investigations. In order to obtain sufficiently high screw dislocation densities, it was necessary to measure diffusion in twist boundaries, boundaries which consist of an intersecting grid of screw dislocations. These grids, if atomic mobility along the dislocation is high, constitute a multiply-connected region of high diffusivity. Therefore, although migration on a single dislocation line may be completely correlated, separate dislocation lines in these grids provide the "two or more high-diffusivity paths"⁽⁶⁾ which are necessary if dislocation diffusion is to be rapid. It was observed that D_0 was markedly reduced and the apparent activation energy was little effected for screw dislocation diffusion. This is consistent with the assumption that the defects dissociate along the screw dislocation line, introducing an appreciable activation entropy to the diffusion process, but that the effects of correlation within the dislocation core are compensated by the multiple connectivity of the dislocation array.

Some of the observations presented above make it readily evident that diffusion along a random dislocation in the lattice may be greatly different from diffusion along a dislocation which is part of an array. An isolated screw dislocation apparently would not be a diffusion "short circuit" at all, although such a dislocation could easily climb into a configuration having an appreciable edge character. It can also be shown that an isolated edge dislocation is significantly different from an edge dislocation in a tilt boundary. Since a tilt boundary consists of an array of parallel dislocations, each dislocation in the boundary is constrained by its neighbors to remain, for the most part, straight. If it climbs in either direction, it is resisted by the neighboring dislocation it approaches and attracted by the dislocation it leaves.⁽¹⁸⁾ A portion of the observed activation energy for defect formation in the dislocation core might derive from these neighbor interactions; in particular, the slight increase in the apparent activation energy for diffusion in dislocations at large dislocation densities⁽²⁾ may result from these interactions. A random dislocation is not exposed to these constraints and might therefore be expected to have a relatively high concentration of climb jogs.⁽⁷⁾ These jogs could act as permanent sources and sinks for either vacancies or interstitial atoms and, since the

energy of formation of either defect would be smaller (approximately one-half as large) at a jog, they might become the dominant contribution to diffusion along dislocations. Neglecting correlation effects for the moment, it follows that diffusion down an isolated dislocation does not necessarily resemble diffusion in a boundary array of dislocations.

Solute atom-vacancy interactions during lattice diffusion are rather complicated.⁽¹⁴⁾ To properly account for the effect of such interactions on solute diffusion, it is necessary to consider the relative population of vacancies in the vicinity of the solute atom and the relative rate of interchange of the vacancies with the solute atom as well as its nearest neighbor, and non-nearest neighbor, solvent atoms. Furthermore, variations in each of these terms are reflected, primarily, in variations in D_0 of the diffusion expression and D_0 is difficult to measure accurately. For diffusion along the dislocation core, it appears that this vacancy-solute atom interaction can be defined with less difficulty and measured with greater accuracy. The free energy required to create a solute interstitial atom in the dislocation core will differ from the free energy required to create a solvent interstitial by ΔG_k , the free energy of the interaction between solute and vacancy. Since this interaction energy effects the probability of formation of the defect, it appears directly in the activation energy for grain boundary diffusion. Thus the diffusivity of tracer quantities of dissimilar solutes in a given solute should depend quite strongly (in their apparent activation energy) on the relative interaction energies between the solutes and vacancies in the solvent. A test of this prediction of the model is planned using tracer quantities of zinc and silver in tilt boundaries in copper.

SUMMARY

The present model for the atomic processes occurring during diffusion along the dislocation core takes adequate account of the existing experimental and

theoretical observations on diffusion in dislocations and makes a number of qualitative predictions regarding diffusion in different dislocation geometries and under different experimental conditions. Some of these predictions appear to be supported by available experimental results; others indicate experiments which test the model further. In the absence of a thorough, quantitative, mechanistic picture of the dislocation process, this model does present a convenient qualitative grasp of the processes accompanying diffusion in the dislocation core.

ACKNOWLEDGMENTS

Particular thanks to Dr. Jens Lothe and Dr. Paul Shewmon for continued interest and assistance in the development of these ideas. This work was made possible through the support of the U.S. Atomic Energy Commission under Contract AT(30-1)-2314 and a Westinghouse Electric Corporation fellowship.

REFERENCES

1. E. HART, *Acta Met.* **5**, 597 (1957).
2. R. HOFFMAN and D. TURNBULL, *Acta Met.* **2**, 419 (1954).
3. P. SHEWMON and G. LOVE, *Acta Met.* **11**, 899 (1963).
4. P. COULOMB, C. LEYMONIE and P. LACOMBE, *C.R. Acad. Sci. Paris* **246**, 1209 (1958).
5. R. BALLUFFI and R. THOMPSON, *J. Appl. Phys.* **33**, 817 (1962); **33**, 803 (1962).
6. J. LOTHE, *J. Appl. Phys.* **31**, 1077 (1960).
7. J. FRIEDEL, *Les Dislocations*, Gauthier-Villars, Paris (1956).
8. S. AMELINCKX and W. DEKEYSER, *S.S. Phys.* **8**, 325 (1959).
9. R. SMOLUCHOWSKI, *Phys. Rev.* **87**, 482 (1952).
10. J. FISHER, *J. Appl. Phys.* **22**, 64 (1951).
11. M. ACHTER and R. SMOLUCHOWSKI, *Phys. Rev.* **83**, 163 (1951).
12. F. SEITZ, *Advanc. Phys.* **1**, 43 (1952).
13. R. HOFFMAN and D. TURNBULL, *J. Appl. Phys.* **22**, 634 (1951).
14. P. SHEWMON, *Diffusion in Solids*, Ch. 2. McGraw-Hill, New York (1963).
15. G. POUND, W. BITLER and H. PAXTON, *Phil. Mag.* **6**, 473 (1961).
16. C. GERTSRIKEN and A. REVO, *Ukrain. Fiz. Zhur.* **6**, 398 (1961).
17. A. SEEGER and P. SCHILLER, *Acta Met.* **10**, 348 (1962).
18. A. COTTRELL, *Dislocations and Plastic Flow in Crystals*. Clarendon Press, Oxford (1953).