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DISLOCATION CLIMB IN THE ELECTRON WIND

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ABSTRACT

When a current flows in a conductor, the electron wind can cause atoms to diffuse. This paper considers the consequences of such diffusion along dislocation cores. A dislocation climbing in a crystal is viewed as a non-equilibrium thermodynamic system to define the force that drives self-diffusion along the core. Not only is a dislocation a mass-transport pipe, it also climbs and generates more dislocations—all in the electron wind. A prismatic loop climbs like a rigid disk, as atoms electro-migrate along the core from one edge of the loop to the other. Each loop is therefore a mass carrier responding to an electric current. Interstitial and vacancy loops can be simultaneously generated and subsequently climb in the opposite directions. The process can transport mass in single crystal or bamboo-like interconnects at moderate temperatures.

INTRODUCTION

Present-day aluminum interconnects are about 1 μm wide and operate under severe conditions: high stress, intense current, and temperature exceeding one third of the melting point (933 K). Diffusion-mediated degradation is ubiquitous as the *brute forces* act in the *small dimensions*. The intense electric current causes atoms to diffuse. The phenomenon, known as electromigration, has been a persistent problem since the 1960s [1-4]. Voids enlarge at locations where atoms diffuse away, raising electrical resistance and eventually leading to open failure. At temperatures of practical concern, grain boundary transports much more matter than crystalline lattice, so that grain boundaries have been taken to be the dominant diffusion path.

In narrow interconnects having bamboo-like grain structures, grain boundaries no longer connect into a network, and therefore do not by themselves facilitate long-range diffusion. Yet voids still grow in the bamboo-like lines, indicating long-range diffusion. Other fast diffusion paths include conductor-dielectric interfaces, void surfaces and dislocations. This paper will concentrate on dislocations [5-7].

To have a feel for the relative magnitude, compare two conductors: a single crystal having dislocation spacing L, and a polycrystal having grain diameter d. The diffusivities of a grain boundary and a dislocation core are similar in magnitude, $D_b \approx D_d$. Denote δ_b as the grain-boundary thickness and A_d as the core area, which are approximately related by $\delta_b^2 \approx A_d$. The ratio of the mass transfer rates in the two conductors is

$$\frac{D_b \delta_b / d}{D_d A_d / L^2} \approx \frac{L^2}{\delta_b d}.$$

The ratio does not vary with temperature owing to the similar magnitude of the activation energies, but does vary with grain diameter and dislocation spacing. For $\delta_b = 3 \times 10^{-10}$ m, d =

3×10^{-6} m and $L = 10^{-7}$ m, the single crystal conductor transfers mass at a rate about 10% of that in the polycrystal conductor.

It remains unclear whether dislocations aligned in the current direction are as closely spaced as assumed above. Dislocation spacing exceeds 1 μm in annealed bulk metals. In aluminum thin films, dislocations can be generated by thermal stresses during film preparation; spacing ranging $L = 0.1 \sim 0.7$ μm is reported under various anneal conditions [8]. These values are converted from the dislocation density ρ_d by $L = \rho_d^{-1/2}$. Of course not all observed dislocations are aligned in the current direction.

In a previous paper [5], it was noted that (1) dislocations not aligned in the current direction will also participate in transporting mass, and (2) dislocations can be generated in the electron wind. A prismatic loop moves under an electric field as atoms electro-migrate along the core from one edge of the loop to the other. The loop is therefore a mass-carrier responding to the electron wind. Furthermore, prismatic loops can multiply from a dipole, which in turn can be generated in the electron wind. Interstitial and vacancy loops are simultaneously generated and subsequently migrate in the opposite directions; neither lattice nor grain-boundaries need to donate vacancies. The mechanism may dominate in well-annealed metals at low temperatures, where other mechanisms are suppressed.

This paper will present a more general framework applicable to a wider class of problems. A prismatic loop is first viewed as a non-equilibrium thermodynamic system to define the driving force for core diffusion, and then as a dynamical system to study its shape evolution. The dynamical system, containing plain metallurgical ingredients, exhibits extraordinarily rich behaviors.

There is evidence, primarily due to the Soviet scientists, that dislocations can also *glide* under electric current; see a recent review [9]. The phenomenon is referred to as electroplasticity. It is believed that a dislocation scatters the electron wind which, in turn, exerts a force on the dislocation. The dislocation glides if the force exceeds lattice resistance [10]. The effect of electroplasticity on interconnect damage is still an open topic. Dislocation glide will not be pursued in this paper, assuming that glide either takes place instantaneously or does not occur at all for certain loop orientations.

NON-EQUILIBRIUM THERMODYNAMICS

A dislocation climbing in a crystal is a non-equilibrium thermodynamic system. Core diffusion is taken to be the only dissipative mechanism. Lattice diffusion can be important over short distances, or in the presence of vacancy supersaturation. To simplify the problem, I will completely ignore these. Core diffusion is slow enough so that both thermal and mechanical equilibrium are established. That is, a uniform temperature, T, and a static elastic field prevail in the crystal. The loop is assumed not to disturb the macroscopic electric potential field, which is governed by the Laplace equation with suitable boundary conditions. For a prismatic loop subjected to a remote electric field, the electric field is uniform everywhere in the crystal.

For simplicity, imagine an interstitial loop lying in a plane normal to the Burgers vector, subjected to an electric field in the same plane. It is a disk of extra atoms in the crystal with its edge defining the loop. Following the classical differential geometry, the position vector $\mathbf{x}$ on the edge is expressed as functions of a parameter u in certain interval, e.g., $\mathbf{x} = \mathbf{x}(u, t)$, $0 < u < 2\pi$, where t is the time. At a fixed time, $\mathbf{x}$ traces the entire loop as u sweeps through the interval. Also defined in the classical differential geometry are the arc element ds, the tangential vector $\mathbf{t}$, the normal vector $\mathbf{n}$ and the curvature κ. [11]. The climb velocity is given by $V_n = \mathbf{n}\cdot\dot{\mathbf{x}}$. The superimposed dot represents the partial derivative with respect to the time. The

normal vector **n** is directed away from the edge of the interstitial disk, so that $V_n > 0$ when atoms are appended to the edge.

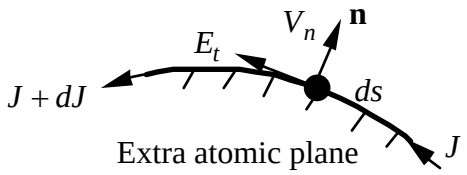

Fig. 1 A segment of the prismatic loop.

Table I Data for Pure Aluminum

Burgers vector	$b = 2.8\times 10^{-10}$ m
Atomic volume	$\Omega = 1.66\times 10^{-29}$ m^3
Resistivity	$\rho = 2.69\times 10^{-8}$ Ωm
Shear modules	$G = 2.54\times 10^{10}$ N/m^2
Core diffusivity [12]	$D_d A_d = 7.0\times 10^{-25}\exp(-0.88\ \text{eV}/kT)$ m^4/s
Effective valence (see text)	$Z^* = 20$

Figure 1 illustrates an element of the loop, d*s*. The Burgers vector *b* is normal to the plane of the paper and the extra atomic plane is below the curve. Denote *J* as the number of atoms per time passing a cross-section of the core. If *J* is uniform along the dislocation line, there are no net atoms flowing into or out of the element, and the element will not climb. If *J* is nonuniform along the dislocation line, there are net atoms flowing either into or out of the element, and the element will climb. Mass conservation requires that the climb velocity be related to the *gradient* of the flux, namely

$$V_n = -\frac{\Omega}{b}\frac{\partial J}{\partial s}, \tag{1}$$

where Ω is the volume per atom. Since atoms only diffuse along the edge, a closed loop will enclose an invariant area. This invariance is implied by (1), as readily verified by integrating over the loop.

Denote *F* as the thermodynamic force per atom that drives self-diffusion along the core. The atomic flux is linearly proportional to the force:

$$J = MF, \tag{2}$$

where *M* is the mobility. This linear law is a good approximation because the potential drop due to *F*, over an atomic spacing, is much smaller than the average thermal energy. That is, $Fb/kT \ll 1$, where *k* is the Boltzmann constant. The mobility is related to the diffusivity by the Einstein relation $M = D_d A_d / \Omega kT$. The mobility, and therefore the diffusivity, are determined experimentally by their macroscopic consequences. For example, the value cited in Table I was determined by measuring the shrinking rate of a void in a foil, as atoms diffuse along the core of

a dislocation connecting the void to the external surface of the foil , under the influence of the capillarity [12].

The word "force" uttered in the preceding paragraph is meaningless unless one prescribes a way to compute it. An approximate formula is

$$F = \frac{\partial}{\partial s}\left(Z^* e\phi + \Omega\sigma_A - \alpha\Omega G b\kappa\right), \tag{3}$$

where e is the magnitude of the electron charge, Z^* a material constant called the effective valence, ϕ the electric potential, σ_A the component of the applied stress along the Burgers vector, $\alpha \approx 0.5$, G the shear modules, and κ the curvature of the loop. The force is due to the electron wind, the applied stress, and the line tension. If each effect operates alone, atoms will diffuse toward the location with high electric potential, large stress, or low curvature.

Equations (1)-(3) specify a dynamical system. The following theoretical developments will justify the approximation (3) and also give more general results when this approximation is inadequate. Readers only interested in how loops behave in the electron wind may skip what follows and start reading the section containing Fig. 2.

Core diffusion dissipates energy. Define the force per atom in the core, F, such that the dissipation per time of the system is

$$\dot{D} = \int FJds\,. \tag{4}$$

Unless otherwise stated on one occasion, all integrals in this paper extend over the loop. Energy balance of the system requires that

$$(\text{Energy Rate}) + (\text{Dissipation Rate}) = (\text{Power Supply}).$$

Consequently, F can be evaluated once one knows the energy rate and the power supply.

Electric field E is used in this paper; current density is given by $j = E/\rho$, where ρ is the resistivity listed in Table I. The electron wind, by collision, exerts a force f_E on each atom in the core. The force is proportional to the electric field component tangential to the dislocation line, E_t. Following the conventional notation for grain boundary electromigration, the wind force is written as

$$f_E = -Z^* eE_t\,. \tag{5}$$

Aluminum atoms diffuse in the direction of the electron flow; thus the negative sign. The effective valence Z^* is determined empirically by its macroscopic consequences. An average value reported for grain boundary diffusion is listed in Table I. The electron wind supplies the loop with power

$$\dot{W}_E = \int f_E Jds\,. \tag{6}$$

To embed the interstitial disk into the stressed solid, the stress component normal to the disk does work. The stress in the crystal is a linear superposition of the applied stress σ_A and the self-stress σ_S , both being the component parallel to the Burgers vector. The applied stress is the

same as the one present prior to the introduction of the dislocation, such as the thermal stress. As the loop climbs, σ_A supplies the loop with power

$$\dot{W}_A = \int b\sigma_A V_n ds. \tag{7}$$

The applied stress is independent of *b*, so that the power is linear in *b*.

The self-stress σ_S, which is solved from a three-dimensional elasticity problem, includes the interactions both among the segments of the loop, and between the loop and the traction-free boundary of the crystal. The self-energy, *U*, is computed from

$$U = \frac{1}{2} b \iint \sigma_S dS. \tag{8}$$

The integral extends over the area enclosed by the loop. The self-stress is linear in *b*, so that *U* is quadratic in *b*.

The energy balance requires that the power of the electron wind and the applied stress be either stored in the system as the self-energy or dissipated in the diffusion process. Thus,

$$\dot{U} + \int FJds = \int f_E Jds + \int b\sigma_A V_n ds. \tag{9}$$

This statement holds true for any virtual velocity and virtual flux satisfying mass conservation (1). Everything else having been defined, the equation defines the total force on each atom *F*. For a given geometry, *F* can be computed from (9) at every point on the loop; see the next section.

In summary, non-equilibrium thermodynamics contains three ingredients: the *kinematics* of the motion (1), the *kinetics* connecting the motion and the force (2), and the *energetics* prescribing the force (9). One can readily make changes to the ingredients, such as to include current-induced glide, stacking fault energy, and lattice diffusion.

LINE TENSION APPROXIMATION

Solutions of self-energy are available in the literature for idealized situations; e.g. for a planar prismatic loop in an infinite crystal [13]

$$U = \frac{Gb^2}{8\pi(1-\nu)} \oint n'_i \, ds' \oint \frac{\partial^2 \Delta}{\partial x_i \partial x_j} n_j ds.$$

Here ν is Poisson's ratio, *ds* and *ds*' are two elements of the dislocation line, n_i and n'_i are the normal vectors of the two elements, and Δ is the distance between the two elements. The integral is a geometric factor, and must be evaluated with due regard for the core radius.

Dimensional considerations dictate that the self-energy in general should take the form

$$U = \alpha G b^2 P,$$

where *P* is the total length of the loop and *G* the shear modulus of the crystal. The dimensionless coefficient α depends on the geometry, the Poisson's ratio, and other modulus ratios if the solid

is inhomogeneous. The coefficient also weakly varies with the ratio P/r_0, where r_0 is the core radius. For example, for a circular prismatic loop in an infinite isotropic solid, it is [13]

$$\alpha = \frac{1}{4\pi(1-\nu)}\left(\ln\frac{2P}{\pi r_0} - 1\right).$$

Because of the complexity of the self-energy, it is so far not possible to write an explicit formula for F in general. An approximation is usually made to take α to be a constant, say α = 0.5. It is then sensible to assign self-energy to each unit length of the dislocation,

$$U / P = \alpha G b^2 .$$

This quantity is referred to as line tension. The analogy with surface tension is not rigorously correct. A flat surface only affects the positions of a few layers of atoms, and does not induce long range elastic deformation. A straight dislocation not only affects the positions of the core atoms, but also induces long range elastic deformation. I will adopt the line tension approximation so long as it gives qualitatively sensible results.

Within this approximation, the self-energy is proportional to the loop length. One can confirm that the rate of the self-energy is

$$\dot{U} = \alpha G b^2 \int \kappa V_n ds .$$

Here κ is the curvature of the loop, which may vary from point to point; $\kappa > 0$ for a circular interstitial loop. This equation expresses a global quantity, the rate of the energy stored in the crystal, in terms of an integral of the local quantities on the loop, the curvature and the velocity.

Using (1) one writes the energy rate in terms of the flux:

$$\dot{U} = -\alpha G b^2 \int \kappa \frac{\Omega}{b}\frac{\partial J}{\partial s} ds = \alpha \Omega G b \int \frac{\partial \kappa}{\partial s} J ds .$$

The last step invokes integration by parts, assuming continuous flux along the loop. Similarly the power of the applied stress becomes

$$\dot{W}_A = -\int \sigma_A \Omega \frac{\partial J}{\partial s} ds = \int \Omega \frac{\partial \sigma_A}{\partial s} J ds .$$

Putting the above into (9) and insisting its validity for *arbitrary J*, one finds that

$$F = f_E + \frac{\partial}{\partial s}\left(\Omega\sigma_A - \alpha\Omega G b \kappa\right). \tag{10}$$

The term involving σ_A also appears in grain boundary diffusion [14]. The third term is analogous to the capillary effect and strives to minimize the dislocation length.

If the conductor is subjected to a uniform electric field E in the x-direction, $\phi = -Ex$. In general, the electric potential is governed by the Laplace equation with suitable boundary conditions. The electric field tangential to the dislocation line is given by $E_t = -\partial\phi / \partial s$. If Z^*

is constant along the loop, the wind force can be written as $f_E = \partial\left(Z^* e\phi\right)/\partial s$. This result combined with (10) gives (3).

Within the various approximations leading to (3), F is the gradient of the chemical potential

$$\mu = -Z^* e\phi - \Omega\sigma_A + \alpha\Omega GbK .$$

Indeed, μ is the free energy increase due to adding one atom at the edge of the extra atomic plane.

VARIATIONAL PRINCIPLE

Equations (1), (2) and (9) define a dynamical system. In general the evolution of a prismatic loop would be traced as follows. For a given loop configuration, the force at each point on the loop is computed from (9) or its approximation e.g. (10). The flux is then computed from (2) and velocity from (1). The configuration is then updated accordingly for a small time step. The procedure is repeated for many time steps. The results in the following sections were obtained in an *ad hoc* manner by taking advantage of the simplifying features of the specific situations. This section describes a general numerical solution procedure without making the line tension approximation.

Recall that the energy balance (9) is valid for any virtual velocity and flux satisfying mass conservation (1). For a small change in flux, δJ, the velocity changes according to (1). There will also be an associated change in the energy rate $\delta\dot{U}$. All the variations in the rate quantities are taken on a *fixed* loop configuration. One can rewrite (9) as

$$\delta\left[\dot{U} - \int f_E J ds - \int b\sigma_A V_n ds + \int \frac{J^2}{2M} ds\right] = 0 .$$

Here F has been expressed in terms of J by using (2). This suggests the following variational principle.

Of all virtual velocities and fluxes that conserve mass in the sense of (1), the actual velocity and flux minimizes functional

$$\Pi = \dot{U} - \int f_E J ds - \int b\sigma_A V_n ds + \int \frac{J^2}{2M} ds . \qquad (11)$$

For a given loop configuration, Π is a number depends on the virtual flux J which, in turn, is a function defined on the loop. One can readily confirm that this is indeed a *minimal* principle.

In numerical analysis, let β_1, β_2, ... be a set of parameters approximately characterizing the loop geometry. They evolve with time. On the right-hand side of (11), the first three terms will be linear in $\dot{\beta}_1$, $\dot{\beta}_2$,..., and the last term will be quadratic. Thus,

$$\Pi = -\sum_i g_i \dot{\beta}_i + \frac{1}{2}\sum_{i,j} A_{ij}\dot{\beta}_i\dot{\beta}_j .$$

The coefficients g_i and A_{ij} depend on βs but not on their rates; they are evaluated by numerical integration. One may interpret g_i as the generalized forces associated with β_i, and A_{ij} the generalized viscosities.

Minimizing Π as a function of $\dot{\beta}$s, i.e. setting $\partial\Pi / \partial\dot{\beta}_k = 0$, one finds that the rates are determined by a set of linear algebraic equations

$$\sum_j A_{ij}\dot{\beta}_j = g_i .$$

Once $\dot{\beta}$s are solved, βs will be updated for a small time step. The process is repeated for many steps to trace the evolution over a long time.

The virtual work principle (9) and the variational principle (11) have counterparts in surface diffusion problems. They can be used in several ways to obtain physically important results. Using these principles, we have estimated the condition for a θ-phase Al_2Cu precipitate particle in an aluminum matrix to break away from a grain boundary in the electron wind [15]. We have also simulated the process of a void collapsing to a transgranular slit under the thermal stress [16] or in the electron wind [17].

SHAPE-PRESERVING CLIMB

In the electron wind, a prismatic loop migrates as atoms diffuse in the core, from one edge of the loop to the other. An interstitial disk migrates in the direction of the electron flow, and a vacancy disk migrates in the opposite direction, i.e. in the applied field direction.

First consider a circular interstitial disk of radius R in an infinite lattice, subjected to electric field E. The circular shape is assumed to be in equilibrium in the absence of electric field. I will now show that a circular disk translating in the electron wind is a solution. In Fig. 2, A is the symmetry point where the flux vanishes, and B is a point on the edge at height y. Let the disk translate at a uniform velocity V. In unit time, interstitials of volume Vby are removed from the segment AB. They flow out of the element from point B. Mass conservation requires that

$$J = -Vby / \Omega . \tag{12}$$

The electric field component tangential to the dislocation line at point B is $E_t = -Ey / R$. The electron wind is the only force that drives diffusion, so that $J = -MeZ^* E_t$. Thus, the velocity is

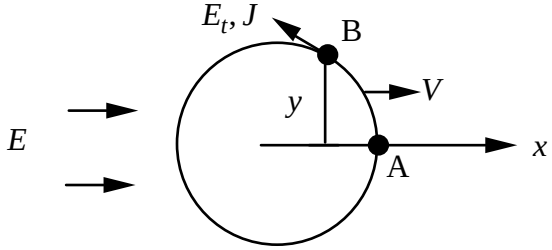

Fig. 2 An interstitial disk drifting in the electron wind.

$$V = -\frac{M\Omega}{Rb} eZ^* E . \tag{13}$$

The disk migrates at a velocity proportional to the applied electric field, and inversely proportional to the loop radius. The negative sign indicates that the interstitial disk migrates in the direction of the electron flow, as anticipated. Under temperature T = 500 K and electric field E = 1000 V/m, a disk of radius R = 0.1 μm translates at velocity $V = 1.8 \times 10^{-11}$ m/s.

Next consider the effect of stress on the migration. We have assumed that the self-stress is in equilibrium and does not distort the circular loop. If the applied stress σ_A is uniform in the plane of the loop, it will not cause the loop to move. If σ_A is independent of y, but has a uniform gradient in the x direction, similar analysis shows that the circular interstitial loop will translate at velocity

$$V = \frac{M\Omega^2}{Rb} \frac{\partial \sigma_A}{\partial x} . \tag{14}$$

The interstitial loop translates towards locations with higher tensile stress.

In an interconnect a stress gradient sets up as electromigration proceeds [14]. The electromigration-induced stress gradient exerts on atoms a force opposite to the electron wind, and is therefore beneficial. The stress gradient can be increased with a passivation coating which constrains plastic flow, or closely spaced diffusion-barriers which build up stresses over a short distance.

High stress gradient can also be induced by defects. A well-known microscopy observation involves a small prismatic loop, migrating with a constant size, under the stress gradient set by a long dislocation [18]. Figure 3 illustrates a straight edge dislocation attracting a small interstitial loop; the extra atomic planes are shaded. The gradient of the stress due to the long dislocation is

$$\frac{Gb}{2\pi(1-\nu)H^2} .$$

Thus, the small loop migrates at velocity

$$V \approx \frac{M\Omega^2 G}{2\pi(1-\nu)H^2 R} .$$

This analysis is approximate in that the edge dislocation will also bend under the influence of the small loop, and the stress gradient due to the long dislocation is not uniform. Nonetheless, the formula can be used to estimate the order of magnitude of the core diffusivity from the experimentally determined loop velocity.

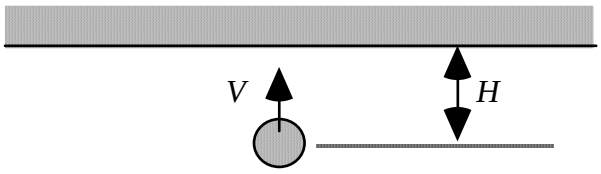

Fig. 3 Migration of a small loop under the influence of a straight dislocation.

Owing to crystal anisotropy, image force or other imperfections, a loop can assume a non-circular equilibrium shape in the absence of the electron wind and the applied stress gradient. After the electric field is turned on, the loop may adjust its shape initially, and then reaches a non-circular steady-state when every point on the loop translates at the same velocity, denoted as V. Suppose the steady-state shape is still symmetric so that the flux vanishes at $y = 0$. Consequently, the flux at an arbitrary point is still given by (12). The electron wind force due to a uniform electric field E in the x-direction is

$$f_E = Z^* eE \frac{\partial x}{\partial s}.$$

The self-energy is invariant at the steady-state, i.e. $\dot{U} = 0$, and the applied stress σ_A is taken to be uniform in the plane of the loop. Consequently, the energy balance (9) becomes

$$\int FJds = \int f_E Jds.$$

In words, the power supplied by the electron wind is completely dissipated in diffusion. However, this does not imply that $F = f_E$ at every point along the loop, for the above equation is valid only for steady-state motion, and not for arbitrary virtual motion. The driving force F will include the contribution of the self-energy.

Replacing F by (2) and J by (12) in the above energy balance equation, one finds that the velocity is given by

$$V = -\frac{M\Omega Z^* eES}{b\int y^2 ds}, \tag{15}$$

where S is the area enclosed by the loop. One can confirm that (15) reduces to (13) when the loop is circular. This equation is useful when the loop shape is known.

SHAPE INSTABILITY

Figure 4a illustrates a straight edge dislocation. The two ends of the dislocation are pinned and immobile, but can be either atomic sinks or sources. The electric field is applied normal to the dislocation line. No force drives diffusion along the straight dislocation, so that the straight line is in equilibrium. Yet this can be an unstable configuration. If the dislocation bends slightly, atoms diffuse in the electron flow direction, i.e. in the direction opposing E_t. Figure 4b shows that E_t drives atoms to diffuse toward the two ends, so that the extra atomic plane recedes. Figure 4c shows the opposite situation: E_t drives atoms to diffuse away from the two ends, so that the extra atomic plane extends. In both cases, the electron wind causes the dislocation to bend further. On the other hand, the line tension strives to straighten the dislocation. The dislocation buckles if the electron wind prevails over the line tension. The critical electric field is given by [5]

Fig. 4 An edge segment buckles in the electron wind.

$$\frac{Z^* eE_c}{\alpha Gb\Omega} L^2 = \pi^2. \tag{16}$$

The dimensionless ratio measures the relative significance of the electron wind and the line tension. For a segment of length L = 1 μm, the critical field is E_c = 180 V/m^2. The buckled dislocation will evolve into either a vacancy dipole (Fig. 4b) or an interstitial dipole (Fig. 4c).

In the electron wind, a circular loop may undergo a shape-preserving migration, as shown in a previous section. However, the circular shape becomes unstable when the electric field E or the radius R exceeds a critical value given by [7]

$$\frac{Z^* eE}{\alpha Gb\Omega} R^2 = 10.65. \tag{17}$$

FINGERING DIPOLES

Figure 5a illustrates a vacancy and an interstitial dipole extending in the opposite directions from a grain boundary. The extra-atomic-plane for each dipole is shaded. With this arrangement, the grain boundary does not donate or accept vacancies, but facilitates vacancies to diffuse from the interstitial dipole to the vacancy dipole. In a real crystal, a dislocation wall may serve the same purpose of the grain boundary. Interstitials and vacancies are transported in the opposite directions, so that the mechanism can lead to both voids and hillocks.

Grain boundary diffusion is fast, so that the flux in the straight portion of the dipoles are driven by the electron wind alone. Each dipole is making a shape-preserving extension at velocity V. Mass conservation requires that

$$VbL = 2M\Omega eZ^* E, \tag{18}$$

where L is the width of the dipole. For shape-preserving extension, this width cannot be arbitrary, and is selected as a result of the competition between the electron wind and the line tension [6]:

$$\frac{Z^* eE}{\alpha Gb\Omega} L^2 = 23.04 . \quad (19)$$

For a given electric field, the width and the velocity are given by the above two equations.

Fig. 5 (a) An interstitial and a vacancy dipole migrate in the opposite directions.
(b) A long dipole blows out loops.

DIPOLE INSTABILITY

A long dipole loses stability again: it blows out a prismatic loop (Fig. 5b). The process repeats itself as the loop drifts away. The instability can be understood by considering two straight edge dislocations placed in the vicinity of each other, which is a segment of a vacancy dipole (Fig. 6). One dislocation exerts on the other a tensile stress normal to the plane. When both dislocations are perfectly straight, there is no driving force for diffusion. When they bend slightly, the tensile stress is greater where the two dislocations are closer. The stress gradient along each dislocation will cause atoms to diffuse along the core such that the two dislocations will bend further. Once again, this positive feedback may be stabilized by the line tension which strives to straighten each dislocation. The dipole becomes unstable when the ratio of the perturbation wavelength λ to the dipole spacing L exceeds a critical value [5].

It is of theoretical interest to mention that the line tension approximation will break down if one pursues problems like these more deeply. In the above situation, both the driving force and the restoring force for the shape instability have the same origin: the self-stress of the dislocation loop. We distinguish them on an *ad hoc* basis. The distinction becomes ambiguous when one analyzes the actual process of a loop breaking away from a dipole. The phenomenon should be analyzed by the more general methods described in this paper.

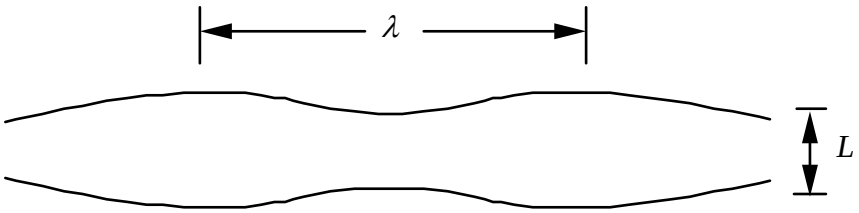

Fig. 6 Instability of a dipole.

CONCLUDING REMARKS

Dislocation climb by means of rapid mass transport along the core of the dislocation itself is considered. In the literature such motion has been called conservative climb because a prismatic loop does not gain or lose atoms as it migrates. Core diffusion tends to be important at low temperatures where the relatively low activation energy causes it to transport more mass than crystalline lattice. In interconnects with bamboo-like grain structures, grain boundaries themselves do not facilitate long-range mass transport. Dislocations, as well as various metal-metal and metal-dielectric interfaces, may become dominant diffusion paths. Conservative climb has been observed in situations where a large stress gradient is set up by a nearby defect, such as another dislocation; see [19] for a literature survey. At present time, experimental observation of such climb under an electric current is unavailable. Yet the theory in this paper is based on well established dislocation mechanics. A prismatic disk is a mass-carrier responding to the electric current. When the electric field is sufficiently large, both interstitial and vacancy disks can be generated and will subsequently migrate in the opposite directions. Considered so far have been some of the elementary features in the dislocation dynamics in the electron wind. Further experimental and theoretical work is needed to assess the effect of the dislocation network on mass transport.

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