On Void Nucleation and Growth in Metal Interconnect Lines under Electromigration Conditions

W.D. NIX and E. ARZT

Electromigration failure in rigidly passivated metal interconnect lines is studied with particular reference to the vacancy supersaturations and hydrostatic stresses that can be developed at blocking grain boundaries under electromigration conditions. It is shown that the high stresses needed for homogeneous void nucleation to occur are probably too high to be developed by electromigration and that failure is more likely to involve the growth of pre-existing voids. We also show that the amount of void growth that can occur at a blocking grain boundary by electromigration of vacancies down the adjoining grain boundaries is small relative to the dimensions of the line unless the adjoining grain boundaries are continuous in a very long section of the line. This suggests that other mechanisms of void growth are responsible for electromigration failure. An analysis of the electromigration of small pre-existing voids shows that above a critical size, large voids migrate faster than smaller ones. This leads to a catastrophic process in which large voids can catch up with and coalesce with smaller ones, growing in size and migrating more rapidly as they do so. We conclude that the migration and coalescence of pre-existing voids is a more likely mechanism of electromigration failure.

DEDICATION

IT is an honor for us to contribute to a volume of work dedicated to the memory of Professor G. Marshall Pound. It was my privilege to know Marsh personally and to work with him at Stanford for some 15 years. Although I never took a course from him, I always considered myself to be one of his students. Indeed, what little I know about thermodynamics and kinetics I learned from him. His enthusiasm for understanding the fundamentals of kinetic processes was absolutely contagious. Those of us who worked and studied with him are still guided by his lessons. There is not sufficient space here to tell even a small fraction of the "Marsh Pound" stories that are known to his friends and colleagues. I will recall just one. I remember going to Marsh sometime in the late 1960's being very confused about the nature of a mole and the chemical entities to which a mole might refer. In the course of straightening me out on this, Marsh told me that we could have a mole of chairs, if we wished! I found that to be a most vivid image and one that I will never forget. It freed me from thinking about statistical mechanics in terms of "chemical" things only. Marsh must have helped thousands of students and colleagues in this way. To my mind, it is one of the main reasons we are dedicating this work to his memory. WDN.

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I. INTRODUCTION

Current densities of the order of 10^{10} A/m^2 or higher can be found in the metal interconnection lines used in integrated circuits, mainly because these interconnects have such small cross-sectional areas. The high current densities produce an "electron wind" that causes the migration of matter to occur, primarily through the drift of vacancies and voids in the line.[1] This form of mass transport can lead to void growth and metal cracking at points of atomic flux divergence and to hillock formation and passivation cracking where the atomic flux converges.[2] Thus, understanding electromigration failures in interconnection lines involves not only understanding the effect of the "electron wind" on the motion of defects in the line but also the various causes of flux divergence. Here we focus our attention primarily on those flux divergences that are caused by microstructural inhomogeneities. Other sources of flux divergence, such as geometrical irregularities or temperature gradients in the line, will not be considered here.

In the present article, we follow the approach of Blech and Herring,[3] who showed that the threshold current density for the migration of thin aluminum stripes[4,5] could be explained by considering the pressure gradients that are set up in the stripe during electromigration. According to this picture, the "electron wind" causes matter to drift from one end of the stripe to the other, and this, in turn, sets up pressure gradients that cause diffusion of matter to occur in the opposite direction. The critical current density for motion of the stripe occurs when the induced stresses reach a critical value. A threshold current density for electromigration failure can also be described in this manner. As discussed by Arzt and Nix,[6] the critical current density for failure can be identified with the maximum stress that can be supported by the line. In the present article, we consider the vacancy supersaturations and corresponding stresses that can be developed at points of flux divergence in an interconnect line. We follow the work of Shatzkes and Lloyd[7] and consider the flux divergences that occur at the ends of grain boundaries that terminate at blocking grains, as shown schematically in Figure 1. Since most electromigration failures occur at relatively low temperatures, we can assume that diffusion occurs only in the grain boundaries and not in the lattice. As a result, the blocking grains are expected to cause both flux discontinuities and failures to occur.

We examine both the nucleation and growth of voids in rigidly passivated metal interconnect lines subjected to electromigration. We wish to determine if the vacancy supersaturations and hydrostatic stresses created in the line can be large enough to cause homogeneous void nucleation to occur. We also examine the growth of preexisting voids in both stress-free and thermally stressed lines. Here we seek to calculate the amount of void growth that can occur under electromigration conditions and to determine if such void growth is sufficient to account for the large voids that are often found in interconnect lines. Most of this analysis deals with the migration of vacancies to stationary voids that are pinned at blocking grain boundaries. A related failure mechanism involves the migration and coalescence of voids in the line. This problem is briefly considered in the last section of the article.

II. GRAIN BOUNDARY DIFFUSION AND ELECTROMIGRATION UNDER THE CONSTRAINT OF A RIGID PASSIVATION

Consider a void-free interconnect line surrounded by a rigid passivation and subjected to a high current density, as shown in Figure 2. We focus our attention on

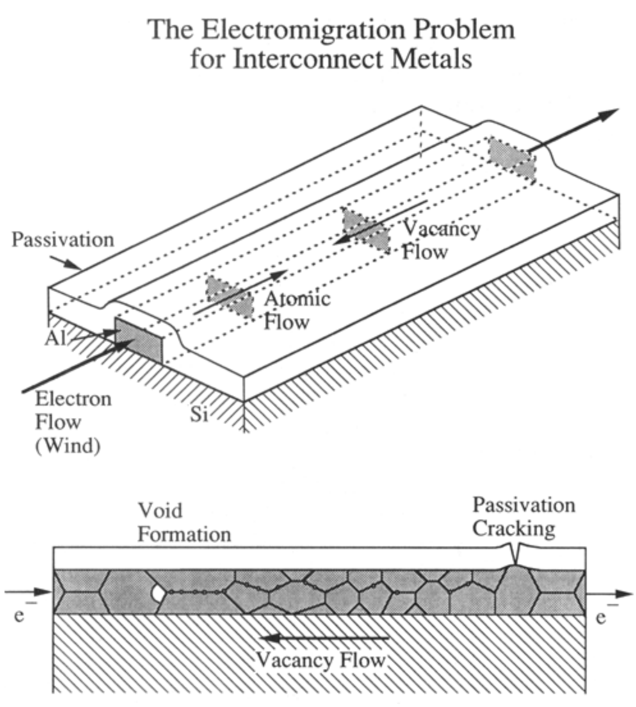

Fig. 1—Illustration of electromigration, void formation, and passivation cracking for a passivated metal interconnection line subjected to a high current density.

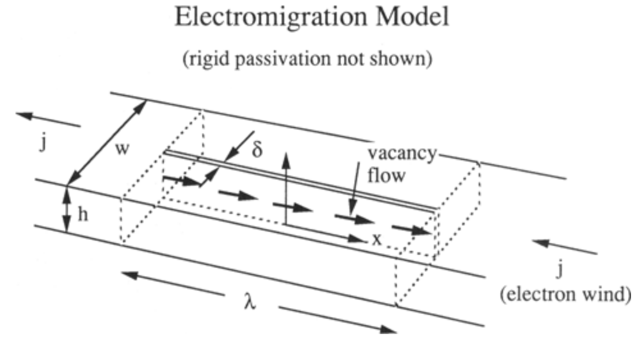

Fig. 2—The electromigration model. A straight grain boundary segment of length λ represents a sequence of grain boundary segments that terminate at blocking grains.

the electromigration of vacancies in a single grain boundary segment of length λ that runs parallel to the line. The boundary segment is considered to end at blocking grain boundaries that extend across the entire cross section of the line. The grain boundary segment in the model is meant to represent a sequence of connecting grain boundary segments that terminate at blocking grains, as illustrated in Figure 1. In a real grain structure, the grain boundaries that carry the vacancy flux are not all perfectly aligned with the interconnect line, with the consequence that some divergence of vacancy flow and void formation could occur at grain boundary triple junctions. However, the potential for void nucleation and the driving force for void growth are much greater at blocking grain boundaries. Thus, we ignore the small divergences along the line and focus our attention on what happens at the ends of the boundary, where void formation is more likely to occur.

We treat the electromigration of vacancies (or mass) in a phenomenological way. Following the approach of Black,[1] the drift velocity of vacancies in the grain boundary can be expressed as

$$v = \frac{D_{gb}^{v}}{\mathrm{k}T} eZ^{*}\rho j \qquad [1]$$

where D_{gb}^{v} is the vacancy diffusion coefficient in the grain boundary, eZ^{*} is the effective charge on the vacancy, ρ is the electrical resistivity, j is the current density, and $\mathrm{k}T$ has the usual meaning. The term $eZ^{*}\rho j$ can be recognized as a force on the vacancy due to the electric field in the interconnect line. The drift of vacancies caused by the "electron wind" causes vacancies to accumulate at one end of the grain boundary. This sets up a vacancy concentration gradient and a diffusional flux that counterbalances the drift due to electromigration.

Usually, a drift term is added to the diffusion equation in the following way:

$$\frac{\partial c}{\partial t} = D_{gb}^{v} \frac{\partial^2 c}{\partial^2 x} - v \frac{\partial c}{\partial x} \qquad [2]$$

where c is the vacancy concentration in the grain boundary "phase," v is the drift velocity of the vacancies, and x is the coordinate position in the boundary. The equation indicates that all of the vacancies arriving at a particular location in the grain boundary (the right-hand side of Eq. [2]) can be used to raise the local concentration (the left-hand side of the equation). However, the grain boundary itself is a source and sink for vacancies, so that some of the vacancies arriving at a particular point in the boundary will be annihilated in an attempt to reach local equilibrium. The process of vacancy annihilation will, in turn, change the local hydrostatic stress in the line because of the constraint of the passivation. In the present treatment, we assume that the grain boundary is a perfect source and sink for vacancies and that the local vacancy concentration is always at the equilibrium value:

$$c = c_0 \exp\left(\frac{\sigma_H \Omega}{\mathrm{k}T}\right) \qquad [3]$$

where c_0 is the equilibrium vacancy concentration in a stress-free grain boundary, σ_H is the corresponding hydrostatic stress associated with the local equilibrium state, and Ω is the atomic volume.

The diffusion equation may be modified to account for vacancy formation and annihilation in a passivated line as follows. We consider that the vacancies arriving at a particular point are used both to raise the local vacancy concentration and to establish the corresponding stress in the passivated line. Thus, we write

$$\left(\frac{\partial c}{\partial t}\right)_{\text{total}} = \left(\frac{\partial c}{\partial t}\right)^{+}_{\text{concentration}} + \left(\frac{\partial c}{\partial t}\right)^{-}_{\text{stress}} \qquad [4]$$

where the "+" sign indicates an increment in the local concentration and the "−" sign signifies vacancy annihilation. We imagine that the stress at a particular point in the line is purely hydrostatic and that the change in stress in an increment of length Δx can be expressed as

$$d\sigma_H = B\frac{dV_v}{V} = B\frac{\Omega dn_v^-}{hw\Delta x}$$

$$= B\frac{\Omega\,\delta h \Delta x dc^-}{hw\Delta x} = \frac{B\Omega\delta}{w}dc^- \qquad [5]$$

where B is the bulk modulus of the line, dV_v/V is the fractional volume change associated with the annihilation of dn_v vacancies, and dc^- is the corresponding equivalent change in vacancy concentration. Using this relation, the second term on the right-hand side of Eq. [4] may be written as

$$\left(\frac{\partial c}{\partial t}\right)^{-}_{\text{stress}} = \frac{w}{\delta B \Omega}\left(\frac{\partial \sigma_H}{\partial t}\right) \qquad [6]$$

But under the assumption of local equilibrium, the local hydrostatic stress in the line must be related to the local vacancy concentration in the boundary according to Eq. [3]. Thus, the loss of vacancies associated with changing the stress may be expected as

$$\left(\frac{\partial c}{\partial t}\right)^{-}_{\text{stress}} = \frac{w}{\delta B \Omega}\frac{\mathrm{k}T}{\Omega}\left(\frac{\partial \ln c}{\partial t}\right) \qquad [7]$$

Finally, the diffusion equation for vacancy diffusion in a grain boundary in a rigidly passivated line in the presence of a drift effect can be written as

$$\frac{\partial c}{\partial t} + \frac{w}{\delta B\Omega}\frac{\mathrm{k}T}{\Omega}\left(\frac{\partial \ln c}{\partial t}\right) = D^v_{gb}\frac{\partial^2 c}{\partial^2 x} - v\frac{\partial c}{\partial x} \qquad [8]$$

We note immediately that for a perfectly rigid line, $B \to \infty$, the second term on the left-hand side of the equation vanishes, as expected. For such a rigid line, an infinitesimal number of vacancies is needed to change the stress in the line.

It is understood that the stress state in the line is not purely hydrostatic, as assumed in this analysis, and that shear stresses must be present. However, the more complete treatment of nonhydrostatic stresses is beyond the scope of this work.

III. VACANCY CONCENTRATIONS AND STRESSES IN A GRAIN BOUNDARY UNDER STEADY-STATE CONDITIONS

As shown in the previous section, Eq. [8] is the appropriate diffusion equation for vacancy diffusion in a rigidly passivated line. In general, the solution of this equation would give the local vacancy concentration $c(x, t)$, from which the local hydrostatic stress $\sigma_H(x, t)$ could be calculated, using Eq. [3]. Such an approach would have to be taken to be able to predict the time at which failure occurs. In the present work, we limit our attention to steady-state problems in which the flux of vacancies associated with the "electron wind" is exactly compensated by a diffusion flux in the opposite direction. In the following sections of the article, we will estimate the rate of homogeneous void nucleation and the rate of void growth under these steady-state conditions.

A. *Steady-State Vacancy Concentrations*

In the steady state, the vacancy concentration at a particular point in the grain boundary does not change with time, so we may set the left-hand side of Eq. [8] equal to zero and obtain

$$D^v_{gb}\frac{\partial^2 c}{\partial^2 x} - v\frac{\partial c}{\partial x} = 0 \qquad [9]$$

This equation may be integrated directly to give the steady-state vacancy concentration in the grain boundary.

$$\ln\left[K_1 + \frac{vc}{D^v_{gb}}\right] = \frac{vx}{D^v_{gb}} + K_2 \qquad [10]$$

where K_1 and K_2 are constants of integration. We also know that the net vacancy flux in the grain boundary is zero at steady state, so we may write

$$J_v = -D^v_{gb}\frac{\partial c}{\partial x} + vc = 0 \qquad [11]$$

By differentiating Eq. [10] with respect to x and combining the result with Eq. [11], it is easy to show that $K_1 = 0$. Thus, at steady state, we may write

$$\ln\left(\frac{vc}{D^v_{gb}}\right) = \frac{vx}{D^v_{gb}} + K_2 \qquad [12]$$

We note that, in general, the constant K_2 may be arbitrarily chosen so that an infinite number of steady-state solutions are known to exist. However, we can use a mass conservation principle to choose among these solutions. Specifically, the value of K_2 can be determined using the condition that all vacancies in the grain boundary are either redistributed in the grain boundary or are used to change the stress in the line. For this analysis, we can write Eq. [12] as

$$c = K_3 \exp\left(\frac{vx}{D^v_{gb}}\right) \quad [13]$$

where

$$\ln K_3 = K_2 + \ln\left(\frac{D^v_{gb}}{v}\right) \quad [14]$$

In the following, we show that the constant K_3 is very close to the initial concentration of vacancies in the boundary prior to establishing the electromigration conditions. Consider the case of an initially stress-free line at equilibrium subjected to electromigration conditions. Using Eq. [5] to relate the number of vacancies annihilated at a particular point to the stress state there, we may express the vacancy (mass) conservation principle as follows:

$$c_0 \delta h \lambda = \int_{-\lambda/2}^{\lambda/2} c\delta h\, dx + \int_{-\lambda/2}^{\lambda/2} \frac{hw}{B\Omega}\sigma_H\, dx \quad [15]$$

The term on the left-hand side of this equation represents the vacancies initially present in the boundary, while the two terms on the right-hand side represent the vacancies still in the boundary after electromigration has occurred and those that were annihilated in the process of establishing the stresses. Using Eqs. [3] and [13] and performing the integrations indicated in Eq. [15], we obtain

$$c_0 \delta h \lambda = 2\left(\frac{D^v_{gb}}{v}\right)\delta h K_3 \sinh\left(\frac{v\lambda}{2D^v_{gb}}\right) + \frac{hw}{B\Omega}\frac{kT}{\Omega}\lambda \ln\left(\frac{K_3}{c_0}\right) \quad [16]$$

as the equation for K_3. If we let $K_3 = \alpha c_0$ and define

$$A \equiv \frac{w}{B\delta\Omega}\frac{kT}{\Omega c_0} \quad [17]$$

$$s \equiv \frac{vx}{2D^v_{gb}} \quad [18]$$

$$s^* \equiv \frac{v\lambda}{2D^v_{gb}} \quad [19]$$

then Eq. [16] can be written as

$$1 = \frac{\sinh s^*}{s^*}\alpha + A \ln \alpha \quad [20]$$

The constant A is so large that the solution to Eq. [20] is always very close to $\alpha = 1$. Thus, the constant K_3 is simply the initial vacancy concentration in the grain boundary, c_0. For the more general case of a thermally stressed line with an initial vacancy concentration of c_{initial}, the result is $K_3 = c_{\text{initial}}$. Thus, in the steady state, the vacancy concentration everywhere in the grain boundary is simply

$$c = c_{\text{initial}} \exp\left(\frac{vx}{D^v_{gb}}\right) \quad [21]$$

where $c_{\text{initial}} = c_0$ for the case of an initially stress-free line at equilibrium.

B. *Steady-State Hydrostatic Stresses*

Now we may calculate the hydrostatic stress that develops at a blocking grain boundary under steady-state electromigration conditions. We consider first the case in which the line is initially stress-free and has an equilibrium concentration of vacancies. Then, we will consider a thermally stressed line with an initial vacancy concentration $c_{\text{initial}} > c_0$..

When the initial vacancy concentration in the grain boundary is c_0 and the line is initially stress-free, the maximum vacancy concentration in the grain boundary after steady-state electromigration has been established can be found by setting $x = \lambda/2$ in Eq. [21]:

$$c_{\max} = c_0 \exp\left(\frac{v\lambda}{2D^v_{gb}}\right) \quad [22]$$

The corresponding hydrostatic stress is, using Eq. [3],

$$\sigma_H^{\max} = \frac{kT}{\Omega}\ln\left(\frac{c_{\max}}{c_0}\right) = \frac{kT}{\Omega}\frac{v\lambda}{2D^v_{gb}} \quad [23]$$

Next, consider a line subjected to an initial thermal stress $\sigma_H^{\text{initial}}$ and a corresponding vacancy concentration

$$c_{\text{initial}} = c_0 \exp\left(\frac{\sigma_H^{\text{initial}}\Omega}{kT}\right) \quad [24]$$

The thermal stress for a perfectly rigid passivation may be estimated as follows:

$$\sigma_H^{\text{initial}} = 3\Delta\alpha\Delta T B \quad [25]$$

where $\Delta\alpha$ is the difference in thermal expansion coefficients between the line and the passivation, ΔT is the temperature change following the passivation, and B is the bulk modulus of the line. With this initial concentration and stress, the maximum vacancy concentration in the line becomes, using Eq. [22],

$$c_{\max} = c_{\text{initial}} \exp\left(\frac{v\lambda}{2D^v_{gb}}\right) \quad [26]$$

and the corresponding stress is

$$\sigma_H^{\max} = \frac{kT}{\Omega}\frac{v\lambda}{2D^v_{gb}} + \sigma_H^{\text{initial}} \quad [27]$$

By substituting Eq. [1] into Eq. [27], we obtain

$$\sigma_H^{\max} = \frac{eZ^*\rho j}{2\Omega}\lambda + \sigma_H^{\text{initial}} \quad [28]$$

for the maximum stress in the line due to electromigration. Obviously, the maximum stress depends on the length of the continuous grain boundary segment, modeled as a straight grain boundary of length λ in the present analysis. The longer the grain boundary segment, the higher the resulting stress.

IV. HOMOGENEOUS VOID NUCLEATION UNDER STEADY-STATE ELECTROMIGRATION

We now wish to determine if the stresses produced by electromigration are sufficiently large to cause homogeneous void nucleation to occur in the line. We first determine the stress needed to cause homogeneous nucleation to occur. The theory of homogeneous void nucleation indicates that the free energy formation of a critical sized void nucleus is approximately

$$\Delta F^* = \frac{16\pi\gamma^3}{3\sigma_H^2} \qquad [29]$$

and that the corresponding rate of void nucleation is

$$J^*_{\text{nuc}} \sim \exp - \left(\frac{\Delta F^*}{kT}\right) \qquad [30]$$

Taking the surface energy of the void to be about $1\ \text{J/m}^2$, it is easy to show that a hydrostatic tension stress of the order of 5 GPa would be needed to cause void nucleation to occur at a significant rate. We now determine the electromigration conditions under which such a high stress might be developed. We assume for these calculations that the current density in the interconnect line is $10^{10}\ \text{A/m}^2$ and that the resistivity of the metal is about $28 \times 10^{-9}\ \Omega \cdot \text{m}$. These values are expected to be typical of aluminum interconnect lines subjected to high current densities. We also take the constant Z^* in Eq. [1] to be 20 for these calculations. Then, for the case of an initially stress-free line, a grain boundary segment length of 185 μm would be needed to reach a stress of 5 GPa. Since boundary segments of this length do not occur in integrated circuits, it seems that electromigration-generated stresses alone cannot account for void nucleation. Even if the line were subjected to an initial thermal stress of 1 GPa, a typical value given by Eq. [25], the grain boundary segment would have to be 148 μm in length to produce a large enough stress to nucleate voids. We conclude that homogeneous void nucleation is a difficult process and that the vacancy supersaturations caused by blocking grains and grain boundaries are not likely to lead to void nucleation. It seems more likely that voids are present in the line just after the passivation is deposited and that the electromigration serves to grow these existing voids. In the next section of the article, we focus our attention on the amount of void growth that can be expected at blocking grain boundaries under steady-state electromigration conditions.

V. GROWTH OF A PRE-EXISTING VOID UNDER STEADY-STATE ELECTROMIGRATION CONDITIONS

Consider a small void to be present in an initially stress-free line at $x = \lambda/2$, where the grain boundary joins up with the blocking grain. We assume that the void is small relative to the dimensions of the line but is sufficiently large to allow the local concentration of vacancies to be fixed at c_0 and the corresponding hydrostatic stress to be zero. During electromigration, vacancies will be created in the grain boundary and transported to the void, eventually resulting in a stationary state in which the net vacancy flux is zero everywhere along the grain boundary. Under this steady-state condition, the vacancy concentration in the line can be found by using Eq. [13] and the condition that $c = c_0$ at $x = \lambda/2$. The result is

$$c = c_0 \exp\left(\frac{v}{D^v_{gb}}\left(x - \frac{\lambda}{2}\right)\right) \qquad [31]$$

and the corresponding hydrostatic stress in the line is

$$\sigma_H = \frac{kT}{\Omega}\ln\left(\frac{c}{c_0}\right) = \frac{kT}{\Omega}\frac{v}{D^v_{gb}}\left(x - \frac{\lambda}{2}\right) \qquad [32]$$

It is an easy matter to determine the total number of vacancies that are removed from the line and deposited in the void in the process of establishing the stationary state. By inserting Eq. [32] into the second term on the right-hand side of Eq. [15] and performing the indicated integration, we can determine the number of vacancies removed from the line to be

$$n_v^{\text{stress}} = \frac{hw}{B\Omega}\frac{kT}{\Omega}\frac{v}{D^v_{gb}}\frac{\lambda^2}{4} \qquad [33]$$

and the corresponding void radius to be

$$\begin{aligned} r_{\text{void}} &= \frac{3}{4\pi}(n_v^{\text{stress}}\Omega)^{1/3} \\ &= \frac{3}{4\pi}\left(\frac{hw}{B\Omega}kT\frac{v}{D^v_{gb}}\frac{\lambda^2}{4}\right)^{1/3} \\ &= \frac{3}{4\pi}\left(\frac{hw}{B\Omega}eZ^*\rho j\frac{\lambda^2}{4}\right)^{1/3} \end{aligned} \qquad [34]$$

Using the following material properties (for aluminum), interconnect dimensions, and electromigration parameters:

$B = 55$ GPa;
$\Omega = 1.66 \times 10^{-29}\ \text{m}^3$;
$h = 1\ \mu\text{m}$;
$w = 3\ \mu\text{m}$;
$e = 1.6 \times 10^{-19}$ coulombs;
$Z^* = 20$;
$\rho = 28 \times 10^{-9}\ \Omega \cdot \text{m}$; and
$j = 10^{10}\ \text{A/m}^2$,

the steady-state void size is $r_{\text{void}} = 0.16\ \mu\text{m}$ for a grain boundary of length $\lambda = 20\ \mu\text{m}$ and $r_{\text{void}} = 0.25\ \mu\text{m}$ for $\lambda = 40\ \mu\text{m}$. While voids of this size are expected to be significant, they are smaller than the dimensions of the line and are much smaller than the electromigration voids typically observed. Other factors must be considered to account for the large voids that are observed. One important factor might be the thermal stress in the line. If the line were covered with a rigid passivation at a few hundred degrees Celsius, an initial hydrostatic tension stress of up to 1 GPa could exist in the line. This would increase the capacity of the line to form vacancies. The resulting void radius after electromigration would then be

$$r_{\text{void}} = \frac{3}{4\pi}\left(\frac{hw}{B\Omega} eZ^*\rho j \frac{\lambda^2}{4} + \frac{\sigma_H^{\text{initial}}}{B} \lambda wh\right)^{1/3} \qquad [35]$$

which for $\sigma_H^{\text{initial}} = 1$ GPa gives $r_{\text{void}} = 0.26\ \mu$m for $\lambda = 20\ \mu$m and $r_{\text{void}} = 0.36\ \mu$m for $\lambda = 40\ \mu$m. Again, the void sizes predicted by this analysis appear to be smaller than those observed. We conclude that other mechanisms are likely to contribute significantly to the growth of voids under electromigration conditions. In the next section of the article, we consider briefly the migration and coalescence of voids as a mechanism of electromigration failure.

VI. MIGRATION AND COALESCENCE OF VOIDS UNDER ELECTROMIGRATION CONDITIONS

It has been known for some time that voids in interconnect lines migrate as a result of the "electron wind" in the line.[8] Here, we estimate the migration velocity of such voids and consider briefly their coalescence. For simplicity, imagine a small void with radius r in the shape of a right circular cylinder with its axis aligned parallel to the interconnect line. The "electron wind" causes atoms to be pushed from one end of the void to the other. The rate at which volume is transferred from one end to the other is approximately $J_s 2\pi r \delta \Omega$, where J_s is the atomic flux in the surface "phase," δ is the thickness of the surface, and Ω is the atomic volume. Following Black,[1] we assume that the atomic drift velocity in the surface is

$$v = \frac{D_s}{kT} F = \frac{D_s}{kT} eZ^*\rho j \qquad [36]$$

as before, so the atomic flux in the surface is approximately

$$J_s = \frac{v}{\Omega} = \frac{D_s}{\Omega kT} eZ^*\rho j \qquad [37]$$

Finally, the drift velocity of the void, v_{void}, is found by setting

$$\pi r^2 v_{\text{void}} = J_s 2\pi r \delta \Omega \qquad [38]$$

which, when combined with Eq. [37], leads to

$$v_{\text{void}} = 2\frac{\delta}{r}\frac{D_s}{kT} eZ^*\rho j \qquad [39]$$

We note that the void drift velocity depends strongly on void size, increasing with decreasing void size. We might expect from this that small voids would catch up with larger ones and coalesce with them in the course of electromigration. This could be an important mechanism of electromigration failure. We note also that voids can travel through the centers of grains as easily as they travel along grain boundaries, because their motion is controlled by surface diffusion, rather than by diffusion in the lattice or grain boundary. However, such transgranular void motion can occur only if the voids are pulled away from the grain boundaries to which they are pinned. The problem of detachment of voids from grain boundaries needs further study.

Equation [39] is expected to be valid in the limit of small voids which occupy only a small fraction of the cross section of the line in which they are migrating. For larger voids, the current density on the sides of the void will increase with increasing void size because of the reduced cross section, which will cause larger voids to move faster than smaller ones. For an interconnect line with a cross-sectional area of wh, this leads to a void velocity law of the following form:

$$v_{\text{void}} = 2\frac{\delta}{r}\frac{D_s}{kT} eZ^*\rho j \left(\frac{1}{1 - \dfrac{\pi r^2}{wh}}\right) \qquad [40]$$

This relation indicates that the void velocity decreases with increasing void size up to a critical size and then increases with increasing void size beyond that point. This function is shown in Figure 3. The void size corresponding to the lowest drift velocity is

$$r^* = \sqrt{\frac{wh}{3\pi}} \qquad [41]$$

which for $w = 3\ \mu$m and $h = 1\ \mu$m is $r^* = 0.56\ \mu$m. One might expect smaller voids to catch up with larger ones if they are smaller than this critical size, but once they grow beyond this size, the larger voids are expected to move faster than some of the smaller ones. The consequence could be that critical sized voids would begin to catch up with and coalesce with the smaller ones, moving more rapidly as they do so and resulting in a catastrophic mechanism of void growth and failure. We think this is a likely mechanism of failure that appears

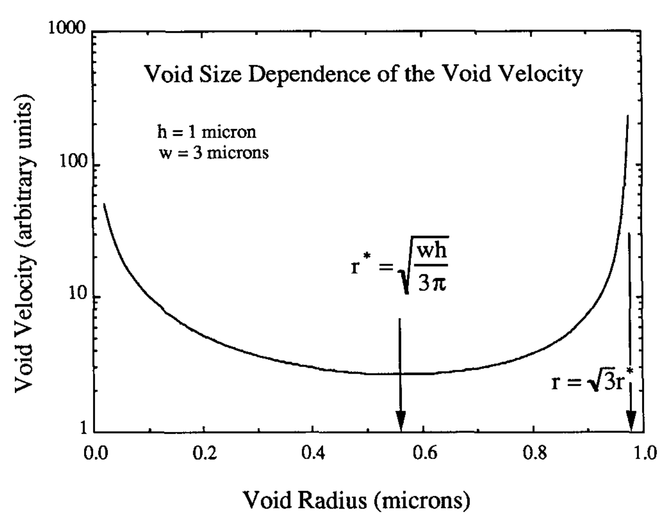

Fig. 3 — Void drift velocity as a function of void size (see Eq. [40]). The void velocity decreases with void size below r^* and increases with void size above that critical value. The void velocity exhibits singular behavior at $r = 1.732r^*$, as the current density on the sides of the void tends to infinity.

to be consistent with direct observations of void growth and motion during electromigration.[8]

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REFERENCES

1. J.R. Black: *IEEE Trans. Electr. Dev.*, 1969, vol. 16, pp. 338-47.
2. J. Sanchez, J.W. Morris, and J.R. Lloyd: *J. Met.*, 1990, vol. 42, pp. 41-45.
3. I.A. Blech and C. Herring: *Appl. Phys. Lett.*, 1976, vol. 29, pp. 131-33.
4. I.A. Blech: *J. Appl. Phys.*, 1976, vol. 47, pp. 1203-08.
5. E. Kinsbron, I.A. Blech, and Y. Komem: *Thin Solid Films*, 1977, p. 446, 139.
6. E. Arzt and W.D. Nix: *J. Mater. Res.*, 1991, vol. 6, pp. 731-36.
7. M. Shatzkes and J.R. Lloyd: *J. Appl. Phys.*, 1986, vol. 59 (11), pp. 3890-92.
8. P.R. Besser, P.A. Flinn, and M. Madden: Stanford University, Stanford, CA, and Intel Corporation, Santa Clara, CA, unpublished research, 1990.