ELECTROMIGRATION RESISTANCE AND MECHANICAL STRENGTH: NEW PERSPECTIVES FOR INTERCONNECT MATERIALS?

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ABSTRACT

A brief review is given of models which propose a correlation between electromigration resistance and the mechanical strength of thin film interconnects. In an attempt to achieve metallurgical strengthening and improved electromigration resistance, aluminum films were implanted with oxygen ions. Preliminary electromigration tests on line arrays patterned from these films resulted in lifetimes comparable to the standard Al films. The lack of improvement is attributed to enhanced hillock/whisker growth during electromigration in the implanted interconnects. This behavior is coincident with a lower compressive strength in similarly treated continuous films at elevated temperatures as measured by the substrate curvature technique.

INTRODUCTION

Electromigration is the major failure mechanism for conductor lines in integrated circuits and, with further miniaturization, its importance is likely to increase in the future. Strategies to inhibit electromigration behavior of pure aluminum lines have involved the addition of alloying elements, such as copper, silicon, magnesium, and nickel, among others. However, due to the lack of a generally accepted explanation for alloying effects, progress is mainly made by trial and error. The second strategy, i.e. grain coarsening, is somewhat more clear-cut because grain boundaries are known to provide the dominant diffusion paths at the low homologous temperatures that are relevant for chip operation. It is curious that the measures of alloying and of grain structure manipulation seem to parallel the steps taken to achieve high mechanical strength in advanced aluminum alloys for structural applications. Therefore it might be speculated that strength is somehow related to electromigration resistance. In order to provide a more rational basis for the development of conductor line alloys, a better understanding of this possible connection is hence desirable.

In this paper we attempt to carry this hypothesis a step further. We first review the classical Blech model and more recent theories, all of which seem to propose a connection between electromigration resistance and yield or creep strength. Then a first attempt to test these predictions is described: aluminum thin films, which were ion implanted in an attempt to raise the mechanical strength, were tested for strength (using continuous films) and for electromigration resistance (using patterned interconnects). The results, while still preliminary, help to shed new light on the electromigration process.

MODELS FOR THE EFFECT OF MECHANICAL STRENGTH ON ELECTROMIGRATION

Mechanical considerations have influenced the understanding of E-M ever since the discovery that the directed diffusion of metal atoms, due to the momentum transfer from the electrons, can lead to stresses in the conductor line, e.g. [1-3]. The probable mechanism is schematically depicted in Fig. 1 [4], where the E-M drift is assumed to be concentrated along a grain boundary running parallel to the line. If the flux continuity is locally disturbed, in the most extreme case by blocking boundaries (as shown in the figure), then the accumulation of atoms (removal of vacancies) at the anode end will set up a hydrostatic compressive stress; likewise, a hydrostatic tensile stress will develop at the cathode end because of the removal of atoms (accumulation of vacancies) there. Such stress gradients have in fact been measured experimentally [3,5,6] and

magnitudes of several hundred Megapascals (up to a Gigapascal) must be considered as typical. These "electromigration stresses" superimpose on any thermal stresses that are usually present in the line.

Gradients in the hydrostatic stress along the line are thought to influence the E-M kinetics in the following way [1]: they cause gradients in the chemical potential for atoms (or vacancies), which in turn drive a diffusional flux that opposes the drift due to E-M. When this effect is incorporated in Black's [7] phenomenological equation, the following expression for the drift velocity results [1]:

$$v = \frac{D}{kT} eZ^* \rho (j - j_c) \qquad \text{where} \qquad j_c = \frac{\sigma^* \Omega}{eZ^* \rho \ell} \tag{1}$$

Here D is the appropriate diffusivity (for the case considered in Fig. 1, D is identical to the grain boundary diffusivity), eZ^* the effective charge of the atom (vacancy), ρ the electrical resistivity, j the current density, Ω is the atomic volume, ℓ the length of the element considered, and σ^* the maximum hydrostatic stress the line can sustain. The quantitiy j_c signifies the critical current density up to which the backdiffusion mechanism manages to fully counterbalance the E-M drift such that no damage occurs. With this reasoning, the line appears "immune" to E-M damage as long as

$$j_c \cdot \ell = \ell_c \cdot j \leq \frac{\sigma^* \Omega}{eZ^* \rho} \tag{2}$$

In other words, E-M occurs only when (at a given element length) the current density exceeds j_c, or when (at a given current density) the element is longer than ℓ_c. Many conceptual problems remain with this simple approach, but equation 2 clearly suggests a correlation between E-M resistance and the maximum hydrostatic stress, σ^*. While it must be remembered that hydrostatic stresses by themselves cannot produce plastic deformation, they can open pores, and thus σ^* can be related to the flow strength of the conductor line material.

This result has recently been extended by Arzt and Nix [8] in an attempt to model the concurrent effects of mechanical strength and line width on the time-to-failure distribution.

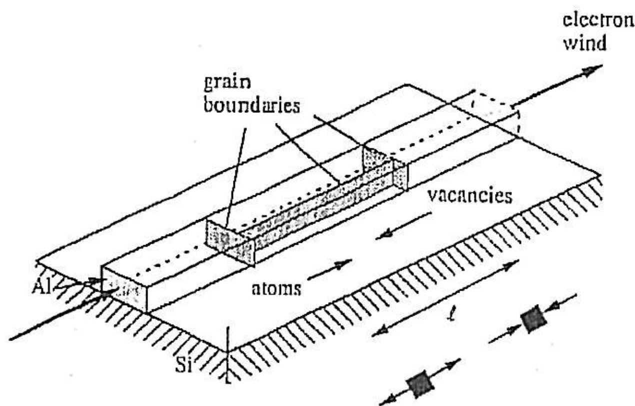

Fig. 1: Schematic illustration of the development of stress gradients along a conductor line as a result of divergences in the E-M flux. Such an element is generally referred to as a "Blech-Schreiber element".

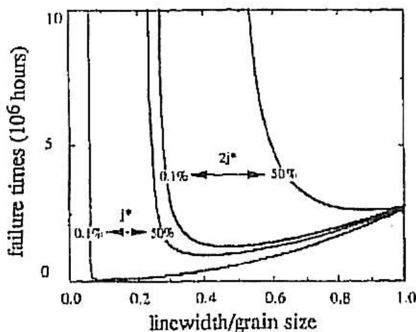

Fig. 2: The hypothetical effect of doubling the mechanical strength of the conductor line (curve marked j* vs. 2j*) on the line-width dependence of the failure times. Calculations following the model by Arzt and Nix [8].

By applying grain structure statistics to the Blech length effect, they arrive at an equation for the failure distribution which contains both the maximum stress and the line width-to-grain size ratio. A typical prediction from this model, shown in Fig. 2, emphasizes the strong effect of mechanical strength on failure times, especially for the case of "near-bamboo" grain structures. With regard to alloy development, the authors also stress the importance of distinguishing between measures (such as alloying) that i) lower the diffusivity D or ii) increase the strength σ^*. The analysis of E-M data showed that both effects may be present in metallization alloys.

In more recent models, an additional element has been added: in view of the elevated temperatures, creep-type deformation is allowed as a stress relaxation mechanism and the assumption of a sharp deformation barrier σ^* is dropped. Glickman et al. [9,10] consider hillock growth by power-law creep, rather than the E-M drift, as the rate-controlling mechanism. Furthermore, a recent model by Arzt [11] treats the coupling between the E-M flux and whisker/hillock growth by climbing grain boundary dislocations. It can be shown that under realistic conditions the kinetics of E-M seem to be governed by the dislocation mobility, rather than by the drift process. In both approaches, the σ^*-concept is replaced by a "creep strength" concept, which allows recent displacement measurements to be better rationalized.

In conclusion, both the early Blech-type models and the refined versions of the theory lead to a seemingly inescapable speculation: if E-M resistance is indeed correlated with mechanical strength or dislocation mobility, then the classical concepts of metallurgical strengthening should be applicable in order to improve E-M behavior. The remainder of this paper describes a first, preliminary attempt to systematically test this hypothesis.

EXPERIMENTAL PROCEDURE AND STRESS CHARACTERIZATION

Al films 0.5 µm thick were magnetron sputtered onto unheated thermally oxidized (100) orientation 100 mm diameter silicon substrates. The deposition rate was 18 Å/sec. at an Ar pressure of 3.0 10^{-3} torr. Sputter system base pressure was 4.0 10^{-7} torr. The films were then annealed at 400 °C for 45 min. in forming gas. Several films were ion implanted with O_{16}, in order to metallurgically harden the films, at a total dose of 1.0 10^{17} ions/cm^2. The implantation was carried out in five partial implants, ranging in energy and dosage from 10 kev to 190 kev and 5.0 10^{15} ions/cm^2 to 4.5 10^{16} ions/cm^2, respectively, in order to achieve a relatively uniform O concentration through the film thickness. Final O concentration is approximately 3.0 at.%. Preliminary TEM investigation of the O_{16} implanted film microstructure showed contrast from a high density of small features ≈10 nm in diameter, presumably as a result of the implantation. Future more detailed work will fully characterize these features.

Al films from each condition were patterned into parallel line arrays (PLA) of 20 lines 1.0 mm long using standard lithographic and dry etch processes. Such PLA patterns are suitable for

simultaneous E-M testing of a large number of interconnects, without refractory metal underlayers, at constant applied voltage. PLA structures preserve the failure sites by preventing arcing as the line produces an open circuit and are ideal for characterizing the nature and microstructural dependence of the induced damage and failure morphologies as a function of testing conditions. E-M testing, at 228 °C and at a current density of 2.0 10^6 A/cm^2, was performed at wafer level on unpassivated lines ≈ 1.8 µm wide using diced wafer chips on a probe station with a hot chuck. Both the standard Al and O_{16} implanted Al films received a final 400°C-45 min. anneal after processing prior to testing. Circuit resistance was constant as the current was increased to the test condition, indicating that joule heating did not increase the line temperature measureably above the hot chuck temperature.

Al film stress behavior during annealing was measured for each film using the wafer curvature technique. Several stress-temperature cycles with final temperatures up to 560°C were carried out in order to monitor the variation of film stress behavior under repeated annealings. A comparison of the second heating and cooling cycle stress curves is shown in Fig. 3. Note that the implanted film is weaker in compression than the standard "control" Al film, with a lower compressive yield stress (98 MPa) at a lower temperature (300 °C) compared to 117 MPA at 340 °C for the control Al film. A more complete characterization of film microstructure along with a correlation to film stress behavior and tendency for hillock formation will be presented elsewhere.

ELECTROMIGRATION RESULTS AND FAILURE ANALYSIS

Preliminary results of E-M testing and analysis of induced damage for several film conditions are reported here. Film resistivities at room temperature as measured on the circuit were 3.32 µΩ-cm for the control Al and 5.38 µΩ-cm for the O_{16} implanted films. The E-M Median Times to Failure (MTF) for the tested film conditions are 27 hours for the control Al and 30.5 hours for the implanted Al interconnects. However, these differences are not conclusive as the O_{16} implanted Al interconnects showed significant resistance increases prior to open circuit failure. The accuracy of testing of PLA structures under constant voltage relies on small resistance changes in each interconnect prior to open circuit, since a gradual change in resistance will change the current supplied to the circuit. Individual O_{16} implanted interconnects showed resistance increases of as much as 50% prior to failure, significantly reducing the current density in each surviving line. Therefore the MTF for the O_{16} implanted structures is significantly less than the measured 30.5 hours.

SEM inspection was performed over the complete line length for each line in the PLA structure for each film. The SEM observations of the tested structures revealed clear differences in the nature and extent of E-M damage between each film condition. It has been shown that the extent of damage, the spacing between corresponding voids and hillocks, and the average void

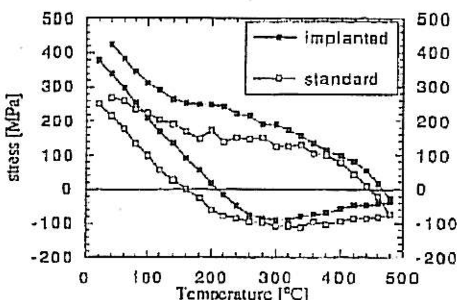

Fig.3: A comparison of the stress vs. temperature curves for the implanted and standard Al films during the second annealing cycle. Note that the implanted film is weaker in compression, but stronger in tension.

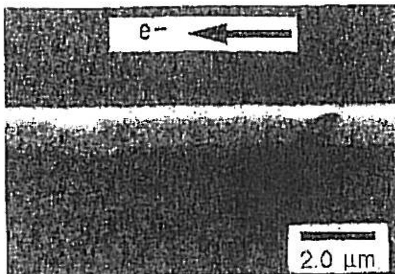

Fig. 4: SEM of E-M void and corresponding hillock in the standard Al film. Void at right is a complete open circuit failure.

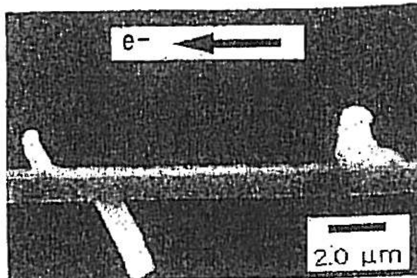

Fig. 5: SEM of E-M void and hillocking in the O_{16} implanted Al film. The failure void is at the base of the whisker on the right.

volume are functions of film microstructure and applied testing conditions. Fig. 4 shows a typical open circuit void and corresponding hillock in the control Al film. Fig. 5 shows the voiding and corresponding whisker formation in the O_{16} implanted interconnects. The tendency for whisker formation and extent of voiding is much greater in the O_{16} implanted film. Whiskers as long as 30 µm are present. Each line in this PLA showed such extensive damage at many sites along its length. It is evident that the large resistance increases during testing for this film are due to this extensive void formation. Correspondingly, the small resistance change prior to open circuit failure in the control Al films is evident in the relatively small amount of voiding and hillock formation seen after testing.

The amount of E-M induced damage in the O_{16} implanted film prior to open circuit failure is remarkable. Extensive voiding and whisker formation has been typically observed only in multiple layer interconnects such as Al alloys deposited over TiN or TiW barrier layers. Barrier layers serve as redundant conduction paths which allow for continued E-M induced mass flux and damage in the Al film even after a complete open circuit in the Al. In the present case, the effect of O_{16} implantation has been to dramatically increase the rate of mass flux, flux divergence and damage in the interconnects without the aid of redundant conduction layers.

The extensive E-M damage in the O_{16} interconnects corresponds to the increased size and density of annealing hillocks, and lower compressive flow stress, in the similar continuous films especially when compared to the corresponding behavior of the standard films. The extensive E-M whisker formations may also be characterized as sites of positive mass flux divergences and of mass accumulation. These mass accumulations may be considered as regions of increased compressive stress, and the whiskers the result of a compressive stress relaxation process at these sites. These results suggest that the effect of the implantation is to produce a weaker Al film in compression, and that the formation of hillocks in continuous films during annealing and whiskers in interconnects during E-M are a result of decreased compressive strength or increased plastic deformation.

DISCUSSION

The aim of the implantation treatment of the Al films was to introduce dislocation obstacles which would raise the yield strength of the conductor line material. The stress measurements by the wafer curvature (Fig. 3) clearly show that a strengthening effect was only achieved in the tensile stress region, while the compressive strength was slightly weaker compared to the "control" film. It is therefore surprising that the "control" film showed much less hillock growth (and no whiskers) both of which are usually as the result of high compressive stresses. This

effect of the implantation on the tendency to enhanced hillock/whisker growth is at present not understood.

The failure of the implanted film to produce longer E-M lifetimes can be explained on the basis of the hillock/whisker observations. The kinetics of E-M damage involves at least three sequential processes: I) the drift due to the electron wind (which is usually considered to be rate-controlling), II) the depletion of atoms at the cathode-near flux divergence (which leads to void nucleation and growth), and III) the depletion of vacancies at the anode-near flux divergence (resulting in whisker or hillock growth). Some of the recent models as discussed above assume process III to be rate-controlling. This is in qualitative agreement with our observed correlation between hillock/whisker growth and damage formation.

Our preliminary experiment suggests that the strategy of strengthening the conductor line material can be successful only if it improves the compressive strength. This conclusion is supported by early measurements of stress gradients [3,5] which detected compressive, but no tensile stresses. Also it can be expected that a grain coarsening treatment would reduce the whisker/hillock growth. Such experiments are currently underway.

CONCLUSIONS

1. Early and recent models of electromigration suggest a correlation between mechanical strength and electromigration resistance.
2. First experiments in which the conductor line material was deliberately strengthened by ion implantation resulted in higher tensile strength, but in slightly lower compressive strength. E-M tests failed to result in longer lifetimes in the implanted films, presumably due to enhanced hillock/whisker growth.
3. It appears that an increase in compressive strength is necessary to improve E-M behavior. The reason for the failure of the implantation process to achieve that goal is at present not understood.

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