Electromigration studies in amorphous and polycrystalline alloys

M. F. Chisholm, D. B. Aaron, J. D. Wiley, and J. H. Perepezko

University of Wisconsin-Madison, Department of Electrical and Computer Engineering, Madison, Wisconsin 53706

(Received 28 March 1988; accepted for publication 10 May 1988)

We report the first direct comparisons of electromigration in amorphous and polycrystalline thin-film samples of the same composition. Matched sample pairs of crystalline and amorphous Cu-Ti alloys were subjected to identical test conditions and tested to failure. It was found that the amorphous alloys tested show approximately one order of magnitude improved resistance to electromigration compared to their crystalline counterparts. The improved performance is attributed to the absence of grain boundary migration paths in the amorphous samples. These findings are consistent with earlier work that showed amorphous alloys to be effective diffusion barriers in thin-film metallization systems.

Electromigration is an atomic mass-transport phenomenon in which atoms of a conductor are driven by the "electron wind," or Coulombic force, associated with a large current density.^{1,2} This phenomenon was studied intensively in the 1960's and 1970's when it was found to be one of the major failure mechanisms in the aluminum metallization lines of integrated circuits.³⁻⁵ These early experiments showed that electromigration proceeds most rapidly along grain boundaries, with depletion and accretion of the migrating atoms occurring primarily at grain boundary triple points.⁶ After sufficient migration has occurred, the depleted regions are macroscopically observable as voids that eventually destroy electrical continuity. Atoms that are removed from the void regions are typically driven to the surface where they form "hillocks." Voids and hillocks are the classic signatures of late-stage electromigration.

The realization that electromigration occurs along grain boundaries led Ames *et al.*⁷ to suggest that the electromigration resistance of aluminum metallizations could be increased by alloying the aluminum with copper which would then segregate at the grain boundaries, partially "blocking" them as migration paths. This strategy proved successful, and the use of Al-Cu and Al-Si alloys instead of pure Al has been standard in the IC industry for many years. As linewidths have continued to shrink, however, electromigration has again become a serious problem, even in these alloy metallizations.^{8,9} In previous work, we have shown that the absence of grain boundaries in amorphous alloys enables these alloys to function as extraordinarily effective diffusion barriers.¹⁰ This suggests that amorphous alloys may also show a high resistance to electromigration. In particular, amorphous alloys lack the easy migration paths (grain boundaries) and the nucleation sites that act as sources and sinks for migrating species (grain boundary triple points). Moreover, amorphous metals often exhibit a negative temperature coefficient for electrical resistivity which should serve to counteract the thermal runaway associated with electromigration failure. Thus it may be proposed that electromigration damage should be more difficult to initiate and sustain in amorphous materials than in polycrystalline materials.

To test this proposal we deposited an amorphous Cu-Ti alloy on oxidized silicon substrates. The samples were depos-

ited in a dc magnetron system, then removed and annealed in a furnace with an inert atmosphere. One-half of the samples were preannealed above their crystallization temperature, thereby causing them to become polycrystalline. The rest were annealed below the crystallization temperature, but above the subsequent test temperature in order to decrease the possibility that relaxation effects would interfere with the electromigration results.¹¹ The samples were then patterned into test stripes for the electromigration experiments. Test stripes consisted of lines 5 μm wide, 600 μm long, and 500–700 nm thick, with $3 \times 3 \text{ mm}^2$ bonding pads on each end.

The stripes of both sample types were subjected to identical test conditions (current density and actual sample temperature) in a test furnace while the voltage and current were monitored. The amorphous samples consistently showed better resistance to electromigration than their crystalline counterparts. In samples that were stressed with $2 \times 10^6 \text{ A/cm}^2$ at line temperatures of 320 °C, the crystalline lines failed after only 25 to 40 h, while the amorphous lines lasted an average of 280 h. Thus the amorphous lines showed approximately one order of magnitude improvement in the mean time to failure (Fig. 1). Actually, the observed failure times represent lower bounds for amorphous Cu-Ti since the failure also involved a crystallization event. At this time, it is

FIG. 1. Normalized resistance vs time plot of an amorphous Cu-Ti line, and a crystalline Cu-Ti line of the same composition and stressed under the same conditions: 320 °C line temperature and $2 \times 10^6 \text{ A/cm}^2$ current density.

FIG. 2. SEM photograph of a failed crystalline Cu-Ti stripe. The composition of the line started as 55 at. % Ti and 45 at. % Cu. After electromigration, region 1 was found to be over 80 at. % Ti. The hillocks (region 2) were composed almost entirely of Cu.

not clear whether crystallization preceded failure or was a consequence of the development of localized heating associated with incipient failure of the amorphous film. Similarly, in samples that were tested at less extreme conditions the amorphous lines showed less electrical resistance change because of electromigration than the crystalline lines.

The failed lines were examined using x-ray microanalysis to determine which species of the alloy migrated more readily than the other. It was found that in the crystalline line (Fig. 2) the copper was deficient around the voids, particularly on the negative electrode side of the voids. The hillocks were found to consist almost entirely of copper. This indicates that the copper was the dominant migrating species.

Note, also, that multiple voids and hillocks were observed in the crystalline samples, indicating that migration damage was occurring in many parts of the sample simultaneously. In the amorphous samples we observed failure only at a single location (Fig. 3). After failure, examination showed high concentrations of titanium on the negative electrode side of the failed region; however, there were also higher concentrations of copper around the failure. The amorphous lines show no clear evidence of hillock formation, minor void, or whisker growth. In some of the failed amorphous samples, clear evidence of crystallization fronts perpendicular to the current direction was detected, leading us to believe that the lines locally crystallized during these experiments. Failures are most likely initiated in these crystalline regions.

FIG. 3. SEM photograph of a failed amorphous stripe of the same composition as the line in Fig. 2. Again region 1 was found to be rich in Ti. There was little hillock formation evident in the amorphous samples; however, region 2 was found to be mostly Cu. Notice the metal scattered in the final phase of electromigration (region 3), presumably due to rapid melting in the final stages of line failure.

We have shown that amorphous alloys can offer significantly improved electromigration resistance over their crystalline counterparts. The order of magnitude improvement in average time to failure for amorphous metals provides another attribute for metallization application. Further investigation is underway to determine both the sequence of events that accompany electromigration failure in amorphous metals and the activation energies in these alloys.

This work was supported in part by Department of Energy under contract No. DE-FG-85ER45096.

¹V. B. Fiks, *Sov. Phys. Solid State* **1**, 14 (1959).

²H. B. Huntington and A. R. Grone, *J. Phys. Chem. Solids* **20**, 76 (1961).

³D. A. Rigney, in *Charge Transfer/Electronic Structure of Alloys*, edited by L. H. Bennett and R. H. Willens (The Metallurgical Society of AIME, New York, 1974), pp. 87-126.

⁴J. N. Pratt and R. G. R. Sellors, *Diffusion and Defect Monograph Series* (Trans Tech, Riehen, Switzerland, 1973).

⁵F. M. D'Heurle and R. Rosenberg, *Phys. Thin Films* **7**, 257 (1973).

⁶R. E. Hummel, in *Electro- and Thermo-transport in Metals and Alloys*, edited by R. E. Hummel and H. B. Huntington (American Institute of Mining, Metallurgical and Petroleum Engineers, Inc., New York, 1977), pp. 93-107.

⁷I. Ames, F. M. D'Heurle, and R. E. Horstmann, *IBM J. Res. Dev.* **14**, 80 (1971).

⁸Y. Pauleau, *Solid State Technol.* **30**, 101 (1987).

⁹H. H. Hoang and J. M. McDavid, *Solid State Technol.* **30**, 121 (1987).

¹⁰R. E. Thomas, K. J. Guo, D. B. Aaron, E. A. Dobisz, J. H. Perepezko, and J. D. Wiley, *Thin Solid Films* **150**, 245 (1987).

¹¹E. A. Dobisz, B. L. Doyle, J. H. Perepezko, J. D. Wiley, and P. S. Peercy, *Mater. Res. Soc. Symp. Proc.* **37**, 479 (1985).