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Citation: Journal of Applied Physics 85, 2583 (1999); doi: 10.1063/1.369624
View online: http://dx.doi.org/10.1063/1.369624
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Texture analysis of damascene-fabricated Cu lines by x-ray diffraction and electron backscatter diffraction and its impact on electromigration performance

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(Received 4 September 1998; accepted for publication 24 November 1998)

The texture of electroplated Cu lines of 0.375, 0.5 and 1.5 μm widths with Ta and TiN barrier layers was analyzed using x-ray pole figure and electron backscatter diffraction (EBSD) techniques. Both techniques indicate a strong (111) fiber texture relative to the bottom surface of the trench for samples with a Ta barrier layer and a 400 °C, 30 min, postelectroplating anneal. Samples with a TiN barrier and no anneal exhibit a weak (111) texture. For both barrier layers the quality of the texture, as measured by (111) peak intensity, fraction of randomly oriented grains and (111) peak width, degrades with decreasing linewidth. EBSD data also indicate (111) texture relative to the sidewalls of the trench in samples with a Ta barrier and postelectroplating anneal. Electromigration tests at 300 °C of 0.36 μm damascene Cu lines with the same process conditions show that samples with very weak (111) texture have median time to failures that exceed those of the strongly textured Cu lines. These results indicate that diffusion at interfaces, such as the Cu/barrier and Cu/overlayer interfaces, along with diffusion along an electroplating seam play more dominant roles in electromigration failure in damascene-fabricated lines than diffusion along grain boundaries within the interconnect. © 1999 American Institute of Physics. [S0021-8979(99)02905-9]

I. INTRODUCTION

The development of faster and smaller integrated circuits (ICs) has put stringent demands on the IC interconnect system. The smaller feature sizes of the interconnect lines lead to current densities that can be as high as 106 A/cm2. In turn, the interconnect lines are more susceptible to failing by electromigration (EM). This phenomenon is the transport of mass under the influence of the electron “wind” current, with the time to failure of the line decreasing with increasing current density.1 Studies have shown that EM performance, measured by median time to failure (MTTF), can be improved by controlling the texture in conventionally fabricated aluminum-based lines.1-3 The MTTF increases with the strength of the (111) texture.1,4,5 The strong (111) texture minimizes the presence of high-angle grain boundaries along the interconnect line, thus minimizing a fast-diffusion path for EM mass transport.6

The push toward fabricating interconnects with linewidths less than 0.25 μm has spawned the development of the damascene-processing method.7-9 Unlike conventional interconnect fabrication (metal deposition, photolithographic patterning, subtractive line definition through reactive ion etching and dielectric deposition), the damascene process begins with dielectric deposition. Trenches are formed in the dielectric by photolithographic patterning and reactive ion etching. This is followed by metal deposition in the trenches and chemical–mechanical polishing to remove excess metalization. This process has the advantage of being inherently planar, facilitating multilevel metallization schemes.

The high current densities associated with smaller interconnect geometries fueled the recent search for metallurgies to replace Al. The motivation for replacing aluminum-based systems was to develop systems with better EM lifetimes and reduced resistance during signal propagation.10 Researchers have demonstrated that copper, having a lower bulk resistivity and higher melting point compared to Al, has superior EM performance11,12 and can be integrated into a multilevel, damascene process with a low dielectric constant polymer.13,14

Damascene fabrication of Cu interconnects offers additional challenges in controlling the interconnect microstructure. The texture of the interconnect line is complicated by the fact that the nucleation during metal deposition occurs on the sidewalls and the bottom of the interconnect trench, rather than only on the bottom as in conventionally fabricated lines. As the aspect ratio of the trenches exceeds 1:1, the role of the sidewalls is expected to become increasingly important in defining the microstructure of the interconnect line.15,16 In this study, we present the texture analysis of damascene-fabricated Cu lines of varying width and the EM performance of highly textured and very weakly textured Cu lines.

II. EXPERIMENTAL DETAILS

Using standard lithography and etch methods, parallel arrays of 0.8 μm deep damascene trenches were patterned into 1.5 μm of plasma-enhanced chemical vapor deposition tetraethoxysilane oxide deposited onto 200 mm single-
crystal Si substrates. The trenches of width 0.375, 0.5, and 1.5 μm had vertical sidewalls, and the trench spacing was a simple multiple (1–2×) the linewidth. A commercially available ultrahigh vacuum, multichamber, sputtering system with a base pressure of \(3 \times 10^{-8}\) Torr was used to deposit 1000 Å Cu onto a barrier of either 300 Å Ta (sputter deposited by physical vapor deposition) or 200 Å TiN (chemical vapor deposition from an organometallic precursor). The trenches were completely filled after deposition of an additional 1.0 μm Cu using a commercially available electroplating tool. The excess Cu and barrier in the field were removed using chemical–mechanical polishing. Prior to deposition of a SiNx passivation on all samples, the Ta barrier samples received a 400°C anneal for 30 min in forming gas.

EM test structures with the same process sequence were fabricated using the structure defined by an 800 μm×0.45 μm single-level line connected to a pad. They were tested at 300°C at a current density of 1.2 MA/cm\(^2\) with a failure criterion of 20% increase in resistance.

The texture of the line arrays was analyzed by both electron backscatter diffraction and x-ray diffraction techniques. X-ray analysis was performed by Advanced Materials Identification and Analysis, Inc. Laboratories (Austin, Texas) using the x-ray pole-figure technique and special optics to achieve a spot size of 10 μms. Orientation Imaging Microscopy\textsuperscript{TM} from TexSEM Laboratories, Inc. (Draper, Utah) was used for collecting and analyzing the electron backscatter diffraction (EBSD) data. EBSD was performed using an Hitachi S800 field-emission scanning electron microscope (SEM) at an acceleration potential of 25 kV. The sample is mounted at a 70° angle to the horizontal position, creating a 20° angle with the electron beam. Details of the EBSD setup can be found elsewhere.\textsuperscript{17} The EBSD data were collected over a fixed area of roughly 80 μm×250 μm with step sizes ranging from 1.5 μm to 2.5 μm and a total number of data points collected exceeding 5000 for each pole figure.

### III. RESULTS AND DISCUSSION

#### A. X-ray versus electron backscatter diffraction techniques

Thin film texture has been characterized by x-ray pole figures\textsuperscript{18,19} and EBSD patterns\textsuperscript{20,21}. A fundamental difference between the techniques is that the x-ray samples the material through the entire thickness of the film with a 10 μm beam size, whereas EBSD receives backscattered electrons from the top 10–50 nm, due to the limited backscattering depth of the electrons.\textsuperscript{22} EBSD samples a surface region whose size is determined by the spot size of the electron beam (approximately 0.06–0.6 μm, depending on the condenser lens setting and the electron source). A complete pole figure is constructed for EBSD by stepping the electron beam through a large area and obtaining the crystallographic orientation of each irradiated region within the area.

The results from both techniques are shown in Table I, where the first number in each column represents the result from x-ray diffraction and the number in brackets was obtained from EBSD. The texture of samples whose pole figures are symmetric about the surface normal may be represented by a plot of the intensity—integrated over the entire (111) pole figure—versus the angle relative to the normal. The texture is quantified by three metrics taken from the plot where the first number in each column represents the result from x-ray diffraction and the number in brackets was obtained from EBSD. The data collected during EBSD fundamentally differs from the x-ray pole plot data. During EBSD, each region irradiated by the electron beam produces a backscattered electron Kikuchi pattern which can then be indexed to determine the crystallographic orientation of the region. The background noise in an EBSD pattern results from regions that are poorly indexed. This

### Table I. Texture results from x-ray and EBSD techniques. The values for EBSD are in square brackets.

<table>
<thead>
<tr>
<th>Barrier layer and trench width</th>
<th>(\omega_{50})</th>
<th>(\omega_{90})</th>
<th>(\omega_{95})</th>
<th>(I_{(111)}) at (\phi=0)</th>
<th>% random</th>
<th>Comments</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ta, blanket</td>
<td>0.51</td>
<td>1.44</td>
<td>1.82</td>
<td>7000</td>
<td>&lt;5 [5]</td>
<td>Strong (111) texture, No other textures</td>
</tr>
<tr>
<td>Ta, 1.5 μm</td>
<td>0.53</td>
<td>1.51</td>
<td>1.96</td>
<td>9500</td>
<td>&lt;5 [4]</td>
<td>Strong (111), weak (511)</td>
</tr>
<tr>
<td>Ta, 0.5 μm</td>
<td>0.67</td>
<td>1.94</td>
<td>2.48</td>
<td>1800[210]</td>
<td>&lt;5[24]</td>
<td>Strong (111), weak (511)</td>
</tr>
<tr>
<td>Ta, 0.375 μm</td>
<td>0.66</td>
<td>1.86</td>
<td>2.36</td>
<td>1400[128]</td>
<td>&lt;5[38]</td>
<td>Strong (111), weak (511) + other textures</td>
</tr>
<tr>
<td>TiN, blanket</td>
<td>4.07</td>
<td>12.15</td>
<td>15.40</td>
<td>400 [70]</td>
<td>19[27]</td>
<td>Weak (111) texture [very weak (200), (511) in EBSD data]</td>
</tr>
<tr>
<td>TiN, 1.5 μm</td>
<td>4.55</td>
<td>14.93</td>
<td>18.35</td>
<td>280 [42]</td>
<td>33[34]</td>
<td>Weak (111) texture [very weak (200), (511) in EBSD data]</td>
</tr>
<tr>
<td>TiN, 0.5 μm</td>
<td>17.55</td>
<td>N/A\textsuperscript{a}</td>
<td>N/A\textsuperscript{a}</td>
<td>50 [9.2]</td>
<td>52[50]</td>
<td>Very weak (111) texture</td>
</tr>
<tr>
<td>TiN, 0.375 μm</td>
<td>13.65</td>
<td>N/A\textsuperscript{a}</td>
<td>N/A\textsuperscript{a}</td>
<td>70 [7.5]</td>
<td>48[45]</td>
<td>Very weak (111) texture</td>
</tr>
</tbody>
</table>

\textsuperscript{a}Not available.
happens when two different grains are simultaneously irradiated by the electron beam. The background fraction, identified by the lowest intensity of the EBSD pole plot, is first subtracted from the pole plot. The angular width of the peak at various volume fractions is calculated by the equal area integration method. The volume fraction of randomly oriented grains is determined in a manner equivalent to that of the x-ray pole figure technique.

In terms of the overall texture (see “Comments” in Table I) both techniques give similar results. However, EBSD is able to resolve very weak (200) and (511) textures in the Ta 1.5 μm and Ta blanket specimens, whereas these textures are not evident in the x-ray data. Figure 1 shows the EBSD (200) and (511) pole figures. A (200) fiber texture [200] planes parallel to the plane of the sample surface and bottom of damascene trenches] would generate a high density of poles at the center of the (200) pole figure. As can be seen from Fig. 1, the density of poles at the center of the figure is negligible. The distinct rings at 16° and 54° result from (511) and (111) fiber textures, respectively. The high density of poles in the (511) pole plot indicates the presence of the (511) fiber texture. The other concentric rings in the (511) pole plot result from the (111) and (511) textures [22° from (115), 39° from (111), 56° from (111), 71° from (111)]. Very weak textures are more readily detected by EBSD. To detect a texture in the x-ray technique, there must be enough volume with the given orientation to produce a signal that exceeds the noise level.25 Additionally, textures whose diffraction patterns nearly overlap in reciprocal space may be difficult to resolve. In contrast, the crystal orientation of each region irradiated by the electron beam is determined in EBSD. The quality of the electron backscatter Kikuchi pattern determines the ability to identify individual grain orientations. The difference in the resolving power of these techniques also shows in the results of the % random component for the specimens with a Ta barrier and 0.375 and 0.5 μm lines. In this case, EBSD yields a % random that is roughly 5–7% that of the x-ray technique. Roughly 5% of the increase in % random for EBSD can be attributed to data smoothing (see below), and the rest is probably due to the higher sensitivity of EBSD to detect discrete orientations.

A linear regression of the $I_{(111)}$ for the trench regions normalized to the $I_{(111)}$ for the blanket regions, $I_{(111)}/I_{(111)-blanket}$ versus linewidth for each technique shows that the values obtained through x-ray characterization lie within the 95 confidence interval of those obtained through EBSD (Fig. 2). This indicates that there is no statistically significant difference between the normalized intensity of central (111) fiber peak for the trench regions.

The angular half widths of the (111) fiber peaks are in fair agreement, with EBSD consistently resulting in larger values. Again, this can be traced back to the fact that EBSD is able to detect the orientation of individual regions within the SEM electron beam. The half widths at % random volume fractions greater than 50% were unable to be resolved by the x-ray technique. These were only resolvable in EBSD after smoothing of the raw data so that the (111) pole plot displayed a function that monotonically decreased from the surface normal ($\phi=0°$). This had the effect of retaining the approximate shape from the raw data and increasing the random fraction by ~5%.

In essence, both x-ray and EBSD characterization give similar results for the normalized (111) peak intensity and random components of the fiber-textured Cu. Except in the case of very weak fiber textures, the EBSD $\omega_{50}$–$\omega_{95}$ values are in fair agreement. The EBSD and x-ray $\omega_{50}$ values diverge at smaller linewidths (0.5 and 0.375 μm), with EBSD resulting in $\omega_{50}$ that are up to 2.5× that of x-ray values. For the remainder of the article, we will focus on the figures of merit obtained through EBSD.

B. Texture of Cu in blanket regions

The complete stereographic projections of the (111) pole figures (from EBSD) of the Cu blanket areas and their (111) pole plots are shown in Figs. 3 and 4. The strongest (111)
texture of the electroplated Cu in both TiN/no-anneal and Ta/anneal specimens is found on the blanket regions. The high density of poles located at the center of the projection and in a ring at ~70° about the center (Figs. 3 and 4) indicates that (111) texture is developed in both samples. These poles correspond to the poles that are parallel to surface normal vector and other poles of the (111) family that are 70.5° away from the central (111) pole, as expected by the face-centered cubic crystal symmetry of the Cu.

The Ta/anneal treatment results in a much tighter (111) texture compared to the TiN/no-anneal treatment with \( \omega_{90, \text{TiN/no-anneal}} = 0.80^\circ \) vs \( \omega_{90, \text{Ta/anneal}} = 3.72^\circ \) (blanket regions, Table I). There is also a lower volume fraction of randomly oriented (111) poles resulting from the Ta/anneal treatment (5%) compared to the TiN/no-anneal treatment (27%). These results are consistent with those of another study that showed a strong (111) fiber texture on a Ta barrier layer and a weak texture on a chemical vapor deposited (CVD) TiN barrier layer.\(^{24}\) We suspect that the high degree of texture on the Ta/anneal samples can be attributed to a combination of (or interaction between) the Ta barrier layer and the post-electroplating anneal. Cu deposited by EP occurs epitaxially,\(^{25}\) suggesting that the Cu seed may be strongly textured in the Ta/anneal sample. Previous studies have shown that strong texture can be inherited from barrier layers\(^{19,26}\) and developed during post-deposition annealing.\(^{19,27,28}\) The textures of the Cu films with Ta and TiN barrier layers in our experiments are both (111). After secondary grain growth, either (111) or (200) textures have been reported to be dominant.\(^{29,30}\) The driving mechanism for the (111) texture is attributed to the surface or interfacial energy minimization; that of the (200) texture is a minimization in the strain energy density. These two mechanisms compete with one another and as the film thickness increases, the strain energy minimization becomes more dominant. Films change from (111) to (200) texture as the thickness increases. Therefore, there is a transition thickness \( h_t \) below which (111) texture is dominant. The absence of other fiber textures suggests that the thickness of the Cu film is below the transition thickness in the regime where the surface and interfacial energy minimization dominate the evolution of texture during postdeposition processing.\(^{30}\)

The results from the TiN/no-anneal are consistent with other research that has shown that TiN barrier layers can lead to Cu films with (111) textures with \( \omega_{50} \) (measured by x-ray diffraction) ranging from 3.2° to 11.2°.\(^{9}\) Smaller \( \omega_{50} \) are known to correlate with greater degrees of crystallinity of the TiN barrier layer.\(^{19}\) Also note that the TiN/no-anneal develops a weak (200) fiber texture component, whereas the Ta/anneal contains only the (111) fiber component. Other researchers have observed (200) texture on a Ta barrier layer but not on a Ta barrier layer.\(^{24}\)

In comparing the texture of the blanket regions of the TiN/no-anneal to the Ta/anneal specimens, it is clear that the
Ta/anneal specimens have a far superior texture, with smaller central (111) peak widths and a lower % random component. Additionally, Ta/anneal specimens developed only (111) fiber texture, whereas the TiN/no-anneal specimens contained a very weak (200) fiber texture. It is suspected that both barrier layer crystallinity and anneal treatment play a significant role in tightening the texture for the Ta/anneal specimens.

C. Texture in trench line arrays

In the trenches, the sidewalls can serve as nucleation sites for (111) grain growth\textsuperscript{15,16,24} and result in a (111) fiber texture initiating from the sidewalls as well as the bottom of the trench. Figures 5 and 6 show the change in the (111) pole figures as the linewidth decreases. The pole figures for the blanket regions are shown for reference. With decreasing linewidths, both the Ta/anneal and TiN/no-anneal specimens show a degradation in (111) texture (see Fig. 2). The poorer texture is indicated by a drop in the intensity of the central (111) peak, an increased % random and an increase in the angular spread of the central (111) peak (see Table I). This trend has been observed in Al\textsuperscript{15,21,31,32} and Cu (Ref. 27) lines fabricated by the damascene process. As expected, the (111) texture in Al measured relative to the substrate normal direction by x-ray diffraction (XRD) and EBSD becomes poorer with decreasing linewidth, or equivalently, with increasing proportion of sidewall nucleation and growth.\textsuperscript{15,21,31,32} Ueno et al.\textsuperscript{33} found similar results for CVD Cu in a damascene structure. They used peak area intensity of the x-ray diffraction peaks of the preferred growth direction [\((200)\) for CVD Cu] as a measure of texture. They found that the higher aspect ratios (line depth/linewidth) resulted in poorer (200) texture relative to the direction perpendicular to the bottom of the trench. They concluded that this was a result of the volume fraction within the trench being dominated by sidewall nucleation and growth. Our results are consistent with previous work.

Incidentally, a study by Hsu et al.\textsuperscript{24} shows the opposite trend: larger linewidths lead to a lower $I_{\text{Cu}(111)}/I_{\text{Cu}(200)}$ x-ray peak intensity. In that study, x-ray diffraction with the Bragg–Brentano $\theta$–2$\theta$ configuration was used to characterize PVD Cu on Ta and TiN barrier layers. The incongruence of their results and ours may be explained by the difference in the characterization techniques. In a study comparing the various x-ray techniques for texture characterization, Rodbell et al. have shown that only complete pole figures give unambiguous results.\textsuperscript{5} Our results support the conclusion of Rodbell et al. The asymmetric distribution of the (111) poles (discussed below) can only be detected from a complete (111) pole figure.

The (111) pole figures for the Ta/anneal specimen also show evidence of a strong (111) texture from the sidewalls [Figs. 5(b) and 5(c)]. For clarity, the pole figures for 0.5 \(\mu m\) linewidths are omitted, since they are similar to those of the 0.375 \(\mu m\) lines. For sidewalls that are 90° from the base of the trench as shown schematically in Fig. 7, (111) texture on the sidewalls, assuming no preferred orientation in the plane of the sidewall, would be expected to produce a diffraction peak perpendicular to the sidewalls and a cone of diffraction with a 70.5° half angle about the perpendicular peak. The resulting pole figure would show the diffraction

![FIG. 5. (111) pole figures for the line arrays and blanket areas of specimen with Ta barrier/anneal. A strong (111) texture is seen in the blanket region, the 1.5 \(\mu m\) lines, and 0.375 \(\mu m\) lines (from left to right). The texture in the 0.5 \(\mu m\) lines (not shown) is in between that of 1.5 and 0.375 \(\mu m\) lines. The faint rings centered about the pole figure at 39° and 56° are due to weak (111) texture. (The origin of the nonconcentric rings in the 0.375 \(\mu m\) pole figure is discussed in Sec. III C).](image1)

![FIG. 6. (111) pole figures for line arrays and blanket areas of specimen with TiN barrier/no anneal. A weak (111) texture is seen in the blanket region, and the 1.5 \(\mu m\) lines (left and center, respectively). The (111) texture in the 0.375 \(\mu m\) lines (right) is very weak.](image2)

![FIG. 7. Schematic of (111) texture from trench surfaces and resulting pole figure. (111) grain growth from the trench surfaces (top) with no in-plane orientation would result in a (111) pole figure (bottom) with superimposed diffraction traces from the (111) fiber textures of the sidewalls and the trench bottom. The central (111) peak and associated ring at 70.5° result from (111) texture from the bottom of the trench. The sidewalls generate two additional pole figures, each rotated 90° from the central pole figure.](image3)
Traces of three superimposed (111) fiber textures, with the two from the sidewalls rotated 90° from the central (111) pole figure (Fig. 7). The traces from the sidewalls show faintly in the 1.5 µm lines and more strongly in the 0.375 µm lines, which have a greater proportion of sidewall nucleation. The (111) pole figure for the sidewalls is represented (Fig. 8), where the data from Fig. 5 are mathematically rotated 90° about the axis along the line shows the (111) texture on the sidewalls (bottom). A central (111) peak and associated ring are evident in the (111) pole figure when viewed perpendicular to the sidewalls.

FIG. 8. (111) pole figures from Ta/anneal 0.375 µm trenches. The (111) pole figure from the top of the trench shows the trace of the (111) texture (top). Mathematically rotating the data through 90° about the axis along the line shows the (111) texture on the sidewalls (bottom). A central (111) peak and associated ring are evident in the (111) pole figure when viewed perpendicular to the sidewalls.

D. Impact of texture on EM performance

The EM performance was evaluated by testing the 0.45 µm lines at 300 °C and 1.2 MA/cm² until failure. A 20% increase in initial resistance was used as a failure criterion. SEM examination of the failed lines revealed voids that spanned the entire width of the lines (Fig. 9) and whose boundaries tended to be near perpendicular to the line axis. The failure times of the TiN/no-anneal exceed those of the Ta/anneal (Fig. 10). Given that (111) texture in the Ta/anneal samples is superior to that of the TiN/no-anneal samples by all measures, this result is somewhat unexpected.

Improved (111) texture is believed to increase the EM lifetime by minimizing high-angle grain boundaries in the interconnect line that can serve as fast diffusion paths. Stronger textures have been shown to increase EM lifetimes in conventionally fabricated Al lines. Abe, Harada and Onada, in a study of the texture of Cu on TiN barrier layers, found that improved texture in Cu damascene lines resulted in a higher MTTF, directly opposite our results. However, they also observed that Cu did not agglomerate on the TiN barrier layer that was used in the samples with the best EM performance. This leads to the conclusion that the samples in their study that happened to have the strongest (111) texture also had the most coherent Cu/TiN interface, eliminating the Cu/TiN interface as a fast-diffusion path.

There are two main theories of EM in Cu damascene structures. One theory is that the microstructure, such as

FIG. 9. SEM micrograph of an EM void. The void has boundaries that are near perpendicular to the line axis and extend the width of the line, indicating that an entire bamboo grain was removed.

FIG. 10. Probability plot for failure times of Ta/anneal and TiN/no-anneal treatments. The failure times of the TiN/no-anneal exceed those of the Ta/anneal.
grain size and texture, controls the EM behavior. Ryu et al.\textsuperscript{35} have shown that the time-to-failure for EP Cu in a damascene structure depends on the grain size and grain size distribution and texture. The other prevailing theory is that EM is primarily controlled not by microstructure but by transport along the barrier/Cu and/or cap/Cu interfaces created in the damascene structure.\textsuperscript{36}

Our results, superior EM performance corresponding to inferior (111) fiber texture, seem somewhat contradictory to the body of research that has correlated good texture with long EM lifetimes. However, the data clearly indicate that for the set of processing conditions used for the TiN/no-anneal and Ta/anneal sample sets, texture is not the first-order controlling factor in EM performance. Further evidence of this is that the grain size in the Ta/anneal samples was several microns and 10× larger than the grain size of the Cu on the TiN/no-anneal sample. Grain size was approximated with an EBSD map of the crystallographic orientation of data points relative to the direction normal to the sample surface. If grain-boundary diffusion were the dominant diffusion path, the TiN/no-anneal samples would be expected to have a lower MTTF. The poor EM performance of the Ta/anneal samples can be explained by considering the other active diffusion paths in EM failure.

The kinetics of electromigration failure has been modeled for Al systems\textsuperscript{37–39} with varying linewidths.\textsuperscript{40,41} Although these studies were focused on conventionally fabricated lines, the principles can be extended to Cu damascene structures. EM is fundamentally a problem of mass transport under the influence of an electron current. The net drift velocity of the conductor atoms is given by the relation\textsuperscript{42}

\[
v_d = v_e - v_b = \left( \frac{D_{\text{eff}}}{kT} \right) \left( Z^* e \rho j - \frac{\partial \sigma}{\partial x} \Omega \right), \tag{1}
\]

where \(v_e\) is the electromigration drift of the conductor atoms, \(v_b\) is the back-flow velocity due to the stress gradient in the interconnect grain, \(D_{\text{eff}}\) is the effective diffusivity, \(k\) is the Boltzmann constant, \(T\) is the absolute temperature, \(Z^*\) is the effective charge number of the diffusing atoms, \(e\) is the absolute value of the electronic charge, \(\rho\) is the resistivity of the conductor, \(j\) is the current density, \(\partial \sigma / \partial x\) is the EM-induced stress gradient along the interconnect line and \(\Omega\) is the atomic volume of the conductor. Increasing \(v_d\) leads to shorter MTTF. Under a given set of external conditions (fixed \(j\) and \(T\)), EM failure is dominated by the diffusion mechanisms that define the \(D_{\text{eff}}\) term [see Eq. (1)].

The candidates for fast-diffusion paths in a damascene structure are the grain boundaries along the length of the interconnect line, the Cu/barrier layer interface, the Cu/overlayer interface or gross voids (or "seams") that may run down the length of the interconnect and form during the electroplating process. Each term contributes to \(D_{\text{eff}}\), weighted by the area fraction along which the diffusion takes place. This is expressed in the equation

\[
D_{\text{eff}} = D_{gb} \left( \frac{\delta_{gb}}{d} \right) + D_{Cu/Ta} \left( \frac{2}{w} + \frac{1}{h} \right) + D_{Cu/o} \left( \frac{\delta_{Cu/o}}{h} \right) + D_{s} \left( \frac{\delta_{seam/s}\ h}{w} \right), \tag{2}
\]

where \(D\) designates the diffusion coefficient along the various diffusion paths that are indicated by the subscripts, \(\delta\) represents the width of the specified diffusion path, \(d\) is the grain size, \(l\) is the length of the seam, and \(w\) and \(h\) are the width and height of the interconnect line. The term representing the contribution of the seam assumes a thin, rectangular seam running down the center of the trench. The proposed EM failure mechanism must consider the relative size of the contribution from each diffusion path in Eq. (2).

In comparing the performance of the Ta/anneal to the TiN/no-anneal samples, we must note that—texture aside—the samples have two different Cu/barrier layer interfaces. Because our data fail to show a correlation between good EM performance and good texture, the EM failure is dominated by either transport along the interfaces or along a possible seam. In considering the role played by the interfaces, it is important to consider the quality of the interface. One study showed that there is no evidence of a reaction or intermixing at the Cu/Ta interface up to 630 °C.\textsuperscript{33} Wong et al.,\textsuperscript{44} however, detected a 4 nm amorphous layer at the Ta/Cu interface after a 400 °C, 1 h anneal. An interface without some degree of mixing could act as a fast diffusion path for EM transport [large \(D_{Cu/Ta}\) term in (2)]. On the other hand, Ti is known to react with Cu at ~350 °C.\textsuperscript{35} It is possible that the Ti within the CVD TiN reacted with Cu to form a more stable Cu/TiN interface which would restrict diffusion at the interface. Other fast diffusion paths may be the Cu/overlayer interface and the seam. At this point, further studies are required to isolate the exact source of EM failure. Additionally, because interfaces are so sensitive to processing conditions, it is likely that the dominant EM diffusion path can change under a different set of processing conditions.

IV. CONCLUSIONS

The x-ray pole figure technique and electron backscatter diffraction yield similar results for thin films with strong fiber texture. EBSD measures of texture, such as \(\omega_{50}\) and % random, tend to be higher than those from the x-ray technique due to greater sensitivity to detecting the texture of small volume fractions. Cu interconnects fabricated with the damascene process on a Ta barrier layer and subsequently annealed have a strong (111) fiber texture relative to the bottom and sidewalls of the interconnect trench. The quality of the (111) fiber texture (relative to the sample surface) as measured by the (111) peak intensity and the angular spread, degrades with decreasing linewidth (i.e., increasing proportion of sidewall nucleation and growth). The same damascene fabrication sequence on a CVD TiN barrier layer with no subsequent anneal results in a weak (111) fiber texture. Because there is no correlation between the quality of the fiber texture and the EM performance for the set of processing conditions used in this study, diffusion at either the Cu/
barrier layer interface, the Cu/overlayer interface and/or an electroplating seam along the interconnect line plays a more dominant role in EM failure. The relative contribution of each term depends on the processing conditions.

ACKNOWLEDGMENTS

The authors would like to thank Derrick Carpenter of Lehigh University for helpful discussions and his foundational work on the algorithms to compare EBSD data directly to x-ray pole figure data. They also thank Susan Chen, Dirk Brown, Takeshi Nogami, Christy Woo, Jacques Bertrand, Fei Wang, Jerry Cheng, Bhanwar Singh, and Minh Tran for processing support and John Sanchez, Jr. for helpful discussions. They would also like to thank Dave Kyser, Ming Ren Lin and Mahboob Khan for AMD-NSF program support. This work was supported in part by a grant from the National Science Foundation (Grant No. ECS-9709447).


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