In situ transmission electron microscopy observations of 1.8 μm and 180 nm Cu interconnects under thermal stresses

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In situ heating transmission electron microscopy was used to observe the stress relaxation behavior in 1.8 μm and 180 nm wide Cu interconnects in real time. 1.8 μm lines exhibit dislocation nucleation at the grain boundaries, while void nucleation/growth was observed in the 180 nm lines. The difference in the stress relaxation mechanism is due to distinct stress states among the two lines, namely, biaxial for the 1.8 μm lines and quasihydrostatic stress for the 180 nm lines. Quasihydrostatic stresses in the 180 nm lines are likely to lead to an absence of dislocation motion. © 2006 American Institute of Physics. [DOI: 10.1063/1.2360240]

Stress induced void formation in Cu interconnects (CI) is a growing concern in the semiconductor industry as CIs are being downscaled due to the demand for chip speed. Current CIs have linewidths around 90–120 nm and are expected to decrease to 45 nm by 2010. As downsizing continues, a reliability issue of concern is the formation of voids due to thermal stresses that occur during the processing of semiconductors, when subjected to thermal cycling.

The origin of thermal stresses in CIs arise when Cu is deposited on the Si wafer, due to a thermal mismatch between Cu and Si. These stresses are further affected by the dimensions of the CI line. When the width-to-thickness ratio is large (i.e., wide lines), the stresses become biaxial, typical of a Cu thin film. On the other hand, when the width-to-thickness ratio is small (i.e., narrow lines), the stresses become quasihydrostatic (σ_x > σ_y > σ_z and σ_z > 0). This typically leads to stress relaxation by plastic deformation in lines under biaxial stress, and elastic deformation in lines under quasihydrostatic stress, which can affect void formation behavior under thermal cycling.

Previous research has been conducted on Cu thin films and/or in wide Cu lines (~1 μm) to identify the sites of void formation. These papers suggest the importance of texture in void formation, specifically the role of grain boundaries, the preferential formation of voids in triple points and grain boundary/line edge intersections, and the local stress concentration at twin interfaces. However, despite the thorough work performed in these studies, Cu films and/or Cu wide lines have different characteristics from narrow Cu lines, particularly their states of stress.

In this context, this letter addresses how changes in CI linewidth affect the relaxation mechanisms during thermal cycling, using in situ heating transmission electron microscopy (TEM). This technique allows us to (i) monitor void nucleation and propagation in real time, (ii) identify the location for void nucleation, and (iii) observe dislocation dynamics. In situ TEM heating techniques have been formerly used on Cu films, but as previously discussed, Cu films are quite distinct from narrow CI lines.

In this work, two different CI lines with dimensions 1.8 μm and 180 nm in width, 300 nm in thickness, 10 mm in length, and spaced 180 nm apart were provided by Freescale Semiconductors. The Cu lines were fabricated using the damascene method, where the Si wafer deposited with a F-doped SiO2 (F-TEOS) interlayer dielectric, was etched to make trenches. Subsequently, a 10–20 nm thick Ta diffusion barrier (DB) was deposited, on which the Cu was electroplated, annealed at 250 °C for 30 min, chemically mechanically polished, and then passivated with a 50 nm SiN layer. An additional 250 nm of F-TEOS was then deposited on top of the SiN layer.

TEM samples for plan view were prepared by disk cutting, polishing, dimpling, and ion milling conventional techniques. The resulting TEM sample consisted of CI bounded by a DB on each side. The DB on the bottom side of the Cu has been removed, but the SiN layer on top of the Cu has been left intact. Though the samples have been thinned in sample preparation, finite element method calculations show that the stresses were not changed significantly as the electron transparent area was small when compared with the overall dimensions of the sample.

TEM images were taken using a JEOL 200CX and a JEOL 2010FX operated at 200 kV. In situ TEM experiments were conducted in a Gatan heating stage equipped with a thermocouple. A water recirculator was connected to the...
heating stage to minimize thermal drift. Initial in situ TEM heating of the CI samples was conducted from room temperature to 500 °C at a rate of 10 °C/min. Subsequently, three thermal cycles were performed from 300 to 500 °C at a rate of about 15 °C/min. The results shown herein represent the observations recorded in repeated experiments.

TEM planar observations of the 1.8 μm and 180 nm CI lines, prior to in situ heating, showed distinct grain structures. While the 1.8 μm lines showed predominantly an equiaxed grain structure, the 180 nm lines exhibited an apparent bamboo structure [Figs. 1(a) and 1(b)]. However, cross-sectional TEM observations showed that grains in 1.8 μm lines were columnar type, whereas in 180 nm lines a multiple-grain structure existed along the thickness [Figs. 1(c) and 1(d)]. This difference is likely due to the sidewall texture present in narrow Cu lines.12

Upon in situ TEM heating, and in the case of 1.8 μm lines, the nucleation of dislocations at Cu grain boundaries and consequent motion away from the grain boundaries during the first thermal cycle were observed (Fig. 2). This was probably due to the (i) biaxial stress state of wide Cu lines, where shear stresses are predominant, (ii) local thermal activation of dislocations, which is diffusion related, and (iii) higher number of grain boundaries, which are prone to the nucleation of dislocations. During subsequent thermal cycles, due to stress relaxation, dislocation motion was not observed.

In the case of 180 nm lines, dislocation nucleation and motion (despite the presence of dislocations) were not observed during in situ heating. This was due to the predominance of quasihydrostatic stresses (negligible shear stresses) in the lines. As the state of stresses is mainly hydrostatic, the stress is relaxed through the nucleation of voids by stress-assisted diffusion, rather than through a dislocation mechanism.

In this work, most of the voids were observed to nucleate at 250 °C during the first cycle, which is in agreement with the maximum creep rate calculated using the McPherson and Dunn model.13 It is important to bear in mind that these CI lines, before being subjected to in situ heating, are initially in a state of high tensile stresses due to the fabrication process. Thus, as the temperature is increased, the stresses relax but diffusion is enhanced. The temperature of 250 °C seems to be a compromise between sufficiently high hydrostatic stresses and significant diffusion, which allows voids to nucleate.

In all cases observed, voids nucleated at triple point junctions, where the Cu grain boundary terminates at the Ta diffusion barrier (Fig. 3). Void propagation was also monitored by in situ TEM, as shown by the sequence of images taken over a 12 min period (Fig. 4). The kinetics of void
growth match well with a one-dimensional grain boundary diffusion model, for which the flux $J$ was calculated using
\[ J = \frac{P_{gb}}{k T} \left( \frac{\partial \sigma_n}{\partial s} \right), \]
where $P_{gb}$ is the grain boundary diffusion coefficient, $k$ is Boltzmann’s constant, $T$ is the temperature, $\sigma_n$ is the stress normal to the grain boundary, and $s$ is the distance traveled by the void parallel to the grain boundary direction. For the calculations, the stress gradient along the grain boundary was considered linear from the intersection of the void and the grain boundary (point $P_1$ in Fig. 4) to where the grain boundary intersects the Ta DB (point $P_2$ in Fig. 4). As local stress values are not available in the literature, $\sigma(P_1)$ was assumed to be 220 MPa based on previous average stress measurements,$^6$ whereas $\sigma(P_2)$ was considered zero because the flux is blocked by the sidewall at the opposite side of the boundary. For these conditions, the calculated flux was found to be $J = 6.6 \times 10^{19}/m^2s$ and the velocity $v = 0.66 \text{ nm/s}$. We predict these values to be lower since $\sigma(P_2)$ is likely to be less than 220 MPa. Nevertheless, the calculated velocity compares well with the velocity $v = 0.23 \text{ nm/s}$ for void growth observed along the grain boundary direction (Fig. 4).

A careful observation of Fig. 4 shows that grain A was consumed, whereas grain B, although partially taken, seemed to act as a barrier during void growth. As the void propagated, it assumed an angle of 140° along most of the void length until total failure occurs. Most likely, this was due to the presence of low surface energy planes, which opened up during void growth. These low energy planes exhibit high surface diffusion rates and thus, facilitated the propagation of the void. The same void angle was observed in other samples, where, in many cases, the grains on both sides of the void were close to the $\{112\}$ orientation. Void propagation characteristics are likely a balance between (i) stress-assisted diffusion along Cu grain boundaries and Cu surfaces, which is predominant in the initial stages of void growth, and (ii) minimization of surface energy, towards the later stages of growth, when the driving force due to stress gradient is diminished. One additional observation recorded during in situ heating was that nanoparticles appeared within the void [Figs. 4(b)–4(d)]. Based on the image’s mass contrast, these nanoparticles are likely clusters of residual Cu. A closer look at these nanoparticles is needed in the future.

These in situ TEM experiments clearly show a definitive difference in the mode by which stress relaxation likely operates in actual CI lines with distinct width-to-thickness ratios. The knowledge acquired in this work will become increasingly important to future microelectronic devices with dimensions below 100 nm.

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