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Effect of Microstructure on Alternating Current-induced Damage in Cu Lines

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Abstract: The effect of microstructure on alternating current-induced damage in 200 and 300 nm thick polycrystalline sputtered Cu lines on Si substrates has been investigated. Alternating currents were used to generate temperature cycles (with ranges from 100 to 300°C) and thermal strains (with ranges from 0.14 to 0.42%) in the Cu lines at a frequency of 10 kHz. Fatigue loading caused the development of severe surface roughness that was localized within individual grains which depends severely on grain orientations.

Keywords: Fatigue, Alternating current, Metal interconnect, Thin film

1. Introduction

Metal interconnects in modern integrated circuits have been undergoing more and more aggressive environments which led to the serious reliability threat. Cu interconnect/low-k dielectric multilevel metallization systems with the increasing numbers of interconnect layer with smaller metal-feature sizes to meet the demands for the faster speed and the better performance of microelectronics devices accompanied larger temperature budget, higher current density, and more challenging integration issues. The most common reliability issues by the passages of electric current through metal interconnect is the electromigration, which has been extensively explored over the last 30 years[1)]. Interconnect voiding by thermal stress due to the repetitive dielectric deposition processes at high temperature around 400°C, so called stress-induced voiding or stress migration, has been considered as an important reliability concern and has been actively investigated[2)]. By the way, metal interconnects in microelectronic devices are subjected to a wide variety of time-varying mechanical loadings caused by the temperature cycling due to the power changes during normal operation, which can be a serious long-term reliability threat of the electronic systems. It was reported 30 years ago[3)] that alternating current-induced thermal strains could lead to surface damage associated with slip band formation in unpassivated wide (25 μm) Al lines with 1 μm thickness. These thermal strains arose during the application of time-varying currents to metal interconnects because of the thermal mismatch of the metal interconnect with Si substrate during temperature changes caused by Joule heating of interconnects itself, which can eventually lead to fatigue damage evolution and electrical failure of the metal interconnect. It has been recently observed that severe damage was formed in Al and Cu interconnects due to the passage of cyclic alternating currents at a frequency of 100Hz[4-6)]. This damage was not due to electromigration but was caused by thermomechanical fatigue from the cyclic temperatures generated by Joule heating of the metal lines. In this work, details on the experimental methods and the general damage evolution mechanisms by applying AC current at a frequency of 10kHz were shown.

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2. Experimental

Test structure with one-level metallization as shown in Fig. 1 was fabricated using the procedure described below. Patterning by electron beam lithography was performed on a <100> Si wafer which was coated with 50 nm of amorphous silicon oxide (a-SiO_x) and 50 nm of amorphous silicon nitride (a-SiN_x). 10nm thick Ta and 200nm thick Cu were sequentially deposited using a dc magnetron sputtering system without a vacuum break, followed by lift-off to make a metal line pattern like Fig. 1 where the length and width of the line were 800 μm and 8 μm, respectively. Then the samples were annealed for 15 h at 400°C in vacuum ($<10^{-6}$ mbar) to stabilize the Cu microstructure. AC test was conducted *in-situ* at a vacuum ($<10^{-5}$ mbar) in a Scanning Electron Microscope (SEM) in which sample was attached to a probe stage using silver paste and the bonding pads connected by 4 needle probes to the electronics outside. Electrical stressing was performed at room temperature using a pure sinusoidal applied voltage (DC offsets less than several mV) with rms current densities of 10 to 40 MA/cm^2 at frequencies of 100Hz and 10kHz. Temperature of metal line was monitored during AC test by measuring time-resolved 4 point resistance of interconnect. It was assumed that measured temperature is nearly uniform along the whole length of long metal line during each short cycle by neglecting edge effects, which were confirmed by finite element analysis[5]. By the limitation on heat dissipation capability through probe stage, sample holder, and SEM chamber for our high frequency regimes, average temperature of Cu line increased continuously for a few minutes (lightly depends on the magnitude of applied frequency and voltage) and then reached a constant value until it (or resistance) rapidly increased just before failure occurred by open circuit.

Fig. 2 showed the measured (dot point) and fitted (solid line) temperatures of 10nm Ta\200nm Cu line at rms current density of 22.3 MA/cm^2 and frequency of 10kHz. Measured temperature (or, resistance) showed divergence at each half cycle of voltage (or, current) because current and voltage values passed zero at that time instant, which made estimation of minimum temperature difficult. Temperature obtained by measured resistance was best fitted by the fitting line which was obtained by assuming that instantaneous temperature (or, resistance) is proportional to the instantaneous power values[5]. Current (or, voltage) cycle of 10kHz with rms current density of 22.3 MA/cm^2 generated temperature cycle (ΔT) of 130°C with frequency of 20kHz. The cyclic

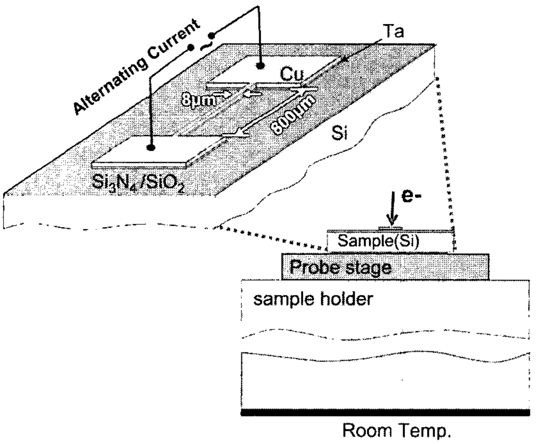


Fig. 1. Schematic of the sample structure used for thermo-mechanical fatigue testing by AC.

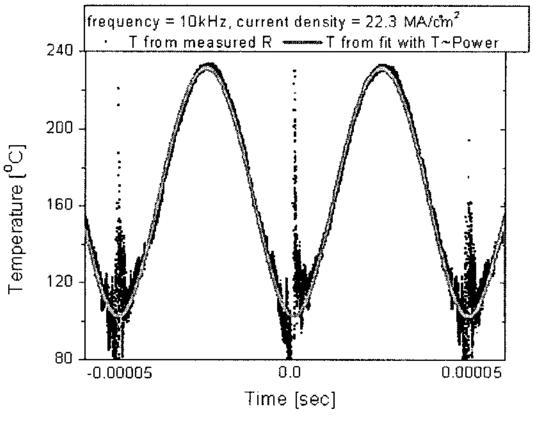


Fig. 2. Measured (dot point) and fitted (solid line) temperatures at rms current density of 22.3 MA/cm^2 and frequency of 10kHz.

change in temperature resulted in cyclic strain in the Cu line due to the thermal expansion mismatches between the Cu line and Si substrate. Average stress of 200 nm Cu film measured by wafer curvature method implied that temperature cycle of 100~270 °C would generate stress cycle of 300~600MPa in Cu film[5]. Grain size and microstructure were characterized by SEM back-scattered electron (BSE) imaging and focused ion beam (FIB) microscopy. Textures of Cu films were investigated by standard X-ray diffraction (XRD) method. Local information on grain orientation was obtained by electron backscatter diffraction (EBSD) and inverse pole figure (IPF).

3. Results

Fig. 3 showed the measured damaged area fraction (DAF) as the number of cycles where DAF was defined as total damaged area measured from high resolution field-emission SEM pictures of whole metal line divided by the whole metal line area. This *in-situ* SEM observation results showed that damages along the line evolved continuously with testing cycles until open failure. Damage growth rate is higher at early stage and then saturated at later stage until open failure. Fig. 4(a) and 4(b) showed typical damage morphologies of (100) out-of-plane oriented grains and (111) out-of-plane oriented grains, respectively, of 10nm Ta\200 nm Cu lines. Left picture showed early stage and right picture showed later stage during cycles. Fig. 5 (a) showed BSE image of the middle location before AC test. Average grain size of 10nm Ta\200nm Cu lines was 1.3±0.5 μm and twin density was high. By XRD and IPF analyses, strong (111) fiber texture and weak (100) component for out-of-plane orientation were observed with randomly oriented in-plane orientations. General aspects of AC-induced damages in common between 100Hz test[6] and 10kHz test performed in this work were mentioned here. Pure AC cycle did not generate electromigration damage due to the very short atomic diffusion length in each cycle[7], but led to thermo-mechanical fatigue damage, as shown in Fig. 4 and 5. Surface damaged grains were distributed along the nearly whole length of the line (cf. Fig 4 or 5) where the damage extents were generally severer at line middle region than line end region. Clearly defined and regularly spaced surface

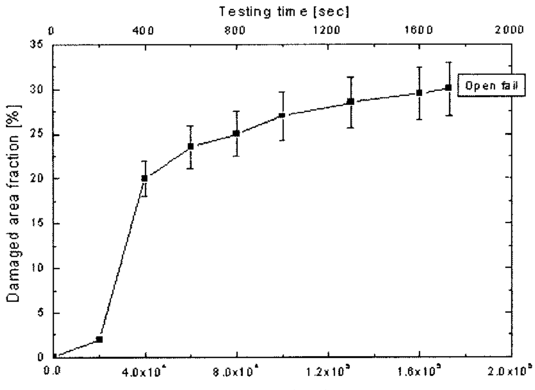
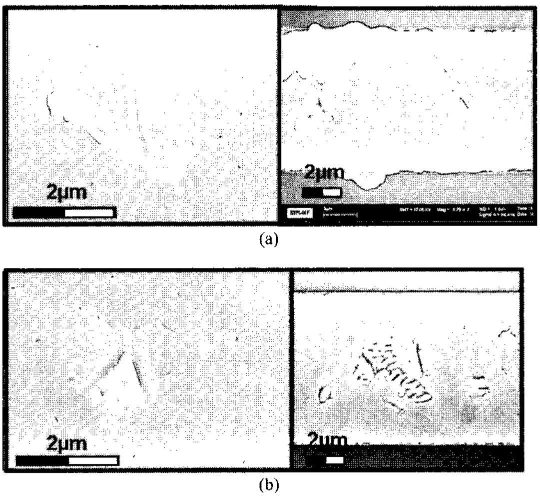


Fig. 3. Measured damaged area fraction (DAF) as the number of cycles where DAF was defined as total damaged area measured from high resolution field-emission SEM pictures of whole metal line divided by the whole metal line area.



(a)

(b)

Fig. 4. Typical damage morphologies of (a) (100) out-of-plane oriented grains and (b) (111) out-of-plane oriented grains, respectively. Left picture showed early stage and right picture showed later stage during cycles.

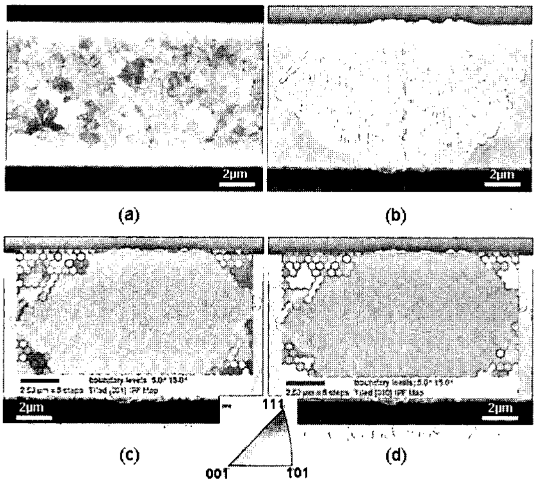


Fig. 5. Damaged (100) grain microstructure; (a) BSE image of the line before AC test, (b) SEM image, (c) EBSD map of out-of-plane orientation, and (d) EBSD map of in-plane transverse (line) orientation for a big damaged (100) grain after AC test at 10kHz frequency, ΔT = 190°C, and 7.5X10^5 cycles. Here, [001] and [010] directions marked in EBSD pictures corresponded to out-of-plane orientation and in-plane line orientation, respectively.

extrusions were often found particularly in (100) out-of-plane oriented grains, which is very similar to that which develops due to irreversible slip processes at the surfaces of mechanically-fatigued Cu8). These damaged (100) grains consumed neighboring grains (both (111) and (100) grains) and were often extended over the entire line width (cf. Fig. 4 (a) and 5). Fig. 5 (b), (c), and (d), respectively, showed the SEM image, EBSD map of out-of-plane orientation, and EBSD map of in-plane line (transverse) orientation for a big damage of 10nm Ta\200nm Cu line which was obtained after AC testing at 10kHz frequency, T= 190°C, and 7.5X10^5 cycles. Here, [001] and [010] directions corresponded to out-of-plane orientation and in-plane line orientation, respectively. EBSD results showed that the big damage in Fig 5(b) is a single grain with close (001) out-of-plane orientations and close (010) in-plane line orientation while the microstructure of undamaged regions in Fig. 5(b) were not changed during AC test, that is, still had preferred (111) out-of-plane orientation as same as the untested Cu line (cf. Fig. 5(a)), as can be seen in Fig. 5(c). More cycles caused these long wrinkles break up into several sub-domains and rotate each other to the different orientations6). While these shallow and uniform wrinkles were formed inside (100) grains, clear extrusions were located along grain boundaries and even outside the line edges to form facetted shape, as can be seen in Fig. 4(a) and 5. By the way, Fig 4(b) showed that different types of damage were observed mostly in (111) out-of-plane oriented grains where grain boundary grooving and deep thinning near grain and twin boundaries were dominantly observed. Compared to the damage morphologies in (100) grains like Fig. 4(a) and 5, wrinkles in (111) grains had rougher surface, wider spacing, relatively random orientations, and deeper thinned region, and seemed to transferred to the neighboring (111) grains. As the testing cycle numbers increased, surface roughness grew in amplitude, density, and size like Fig. 4(b). Later, crack propagation to the line width-direction, that is, normal to the line length-direction preferentially occurred through deep thinned paths inside heavily damaged regions, which eventually led to electrical open fail due to the localized resistance increases and so melting at the very end of the test.

4. Discussions

Severe surface damages of Cu line were developed at high temperature cycling conditions due to the relief of stresses induced by the thermal expansion mismatch between metal interconnect and Si substrate during temperature changes. This surface damage evolution can cause serious reliability concerns because roughened surface can increase sheet resistance, and the very thin regions which develop between each wrinkle can cause higher current density which promote electromigration or even local heating and electrical open failure of metal interconnect. Also, extrusions at boundary or inside of the damaged region (cf. Fig. 4 or 5) can punch through the soft low-k inter-level dielectrics in multilevel metallization structure and presumably lead to inter-layer electrical shorts similarly with hillocks in electromigration phenomena[1]. Hard inter-level dielectrics like chemical vapor deposition silicon dioxide presumably hinder this kind of surface damage evolution while soft low-k dielectrics extensively adopted recently may not retard it[9].

There are three interesting points which are noteworthy to comment about damaged (100) grains shown in Fig. 4(a) and 5. At first, damaged (100) grain growth preferentially generated an orthogonal or facetted configuration of grain boundaries with the segments aligned in close 45° tilted from in-plane line direction. Secondly, boundaries of damaged (100) grains were always parallel to close (110) orientation which is also parallel to the slip trace direction of (111) plane in Cu (100) grain. Thirdly, the long, parallel, partially crosshatched, and evenly distributed wrinkles inside damaged (100) grains were preferentially arrayed in close (100) orientation which is not parallel to the slip trace orientation. The wrinkles inside grain boundaries which are tilted close 45° to the line length direction are close normal to the line length direction like Fig. 4(a) and 5 while the wrinkles inside grain boundaries which are tilted away from 45° to the line length direction are also declined around 15° from the line width direction, which consistently make wrinkles stay in (100) orientation like Fig. 5. Our preliminary transmission electron microscope (TEM) results[10] showed that there were no typical dislocation structures similar as in bulk materials but individual dislocations, and segments of individual or tangled dislocations were arrayed along the same position with the wrinkles inside damaged (100) grains. The surface wrinkles in (100) grains seemed to be resulted from the dislocation interactions and possible dislocation dipoles annihilation[10]. This meant that extensive dislocation activity occurred inside (100) grains by the repetitive fatigue load during AC test and led to extruded wrinkle arrays at film surface which is similar with the mechanical fatigue damage of bulk metal[8].

It was reported over 30 years ago that the migration of grain boundaries in polycrystalline bulk metals fatigued at high homologous temperatures resulted in a diamond or orthogonal configuration of boundaries with respect to the uniaxial stress axis with the boundary segments aligned in the maximum shear stress directions, that is, 45° tilted from the uniaxial stress axis[11-15], which seemed to be a general phenomenon for metal fatigued at high temperatures. If the uniaxial stress axis direction for the mechanical fatigue of bulk metal at high temperatures is assumed to be same as the metal line length direction for our case, that is, for high temperature thermo-mechanical fatigue of thin metal line, it could be seen that damaged (100) grain growth (cf. Fig. 4(a) and 5) showed the same configuration of grain boundary with the reported case[11-15]. Although the static residual stress state in 8 μm-wide Cu line seemed to be close to biaxial rather than uniaxial, it's difficult to estimate the dynamic stress state in the Cu line fatigued continuously during so fast temperature cycles between 100Hz and 10kHz. AC-tested Cu lines frequently showed that crack propagation inside extensive damage preferentially occurred to the line width direction through the whole line width [6,9], which might be an evidence that stress state of metal line during AC cycle seemed to have partial uniaxial component to line length direction rather than isotropic biaxial state. So, the partially uniaxial thermo-mechanical fatigue of Cu line at high temperature might result in damaged (100) grain boundary migration with the boundary segment aligned in the maximum shear stress directions and also aligned in (110) orientation which is parallel to the slip trace direction of (111) plane. Therefore, the repetitive dislocation glide on (111) plane inside (100) grain by the maximum resolved shear stress under the partially uniaxial fatigue loading to the line length direction were thought to result in the characteristic boundary morphology of the damaged (100) grain like Fig. 4(a) and 5. It's not completely understood yet why the wrinkles are arrayed in (100) orientations rather than (110) orientations, which was discussed extensively in ref. 6.

By the way, (111) out-of-plane oriented grains showed the different types of damage like Fig. 4(b). They seemed to have diffusion-controlled processes because such as grooving and thinning near grain and twin boundaries and spherical void formation along twin boundaries were observed by our preliminary TEM results[10,16]. As same as damaged (100) grains, no typical dislocation structures similar as in bulk materials but individual dislocations were found. Characteristic parallel glide dislocations[17] and their emission from grain boundaries were observed, which seemed to be closely related to the grain boundary grooving. The observed voids along twin boundaries in damaged (111) grains by TEM[10,16] were suggested to be resulted from the interaction of the dislocations with twin boundaries. Results on extensive TEM studies to investigate the relationship between dislocation structures and the deformation mechanisms of the damaged (100) and (111) grains will be shown later.

5. Conclusions

Severe fatigue damages were formed in thin Cu lines during temperature cycling by alternating current, and the larger temperature amplitude led to the shorter failure cycles which implied that alternating current could be used to perform thermo-mechanical fatigue test of metal interconnect. (100) grains generated the extensive facetted grain growth by consuming neighboring twin and grain boundaries and had the long, parallel, and evenly distributed wrinkles which were preferentially arrayed in close (100) orientation where boundary segments were aligned in close 45° tilted from metal line length direction and also aligned in (110) orientation which is parallel to the slip trace direction of (111) plane. The repetitive dislocation glide on (111) plane inside (100) grain by the maximum resolved shear stress under the partially uniaxial stress state to the line length direction during so fast cyclic loading were thought to result in this characteristic boundary configuration which seemed to be a general characteristic of grain boundary migration of metal fatigued at high temperatures. (111) grains generated different types of damage such as grooving and thinning near grain and twin boundaries and spherical void formation along twin boundaries which seemed to be related to the diffusion-controlled processes.

Acknowledgements

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