

Lawrence Berkeley National Laboratory

LBL Publications

Title

Plasticity-amplified diffusivity: dislocation cores as fast diffusion paths in Cu interconnects

Permalink

<https://escholarship.org/uc/item/2cr27402>

Authors

Budiman, A.S.
Hau-Riege, C.S.
Besser, P.R.
et al.

Publication Date

2007-04-19

PLASTICITY-AMPLIFIED DIFFUSIVITY: DISLOCATION CORES AS FAST DIFFUSION PATHS IN CU INTERCONNECTS

A.S. Budiman¹, C.S. Hau-Riege², P.R. Besser², A. Marathe², Y.-C. Joo³, N. Tamura⁴, J.R. Patel^{1,4}, W.D. Nix¹

¹*Dept. Of Materials Science & Engineering., Stanford University, Stanford, CA 94305*

²*Advanced Micro Devices, Inc., Sunnyvale, CA*

³*Dept. Of Materials Science & Engineering, Seoul National University (SNU)*

⁴*Advanced Light Source (ALS), Lawrence Berkeley National Laboratory (LBNL), Berkeley, CA 94720*

ABSTRACT

The mass transport of Cu during electromigration (EM) testing is typically dominated by interface diffusion. If a mechanism other than interface diffusion begins to affect the overall transport process, then the effective diffusivity, D , of the EM process would deviate from that of interface diffusion only. This would have fundamental implications. We have preliminary evidence that this might be the case, and we report its implications for EM lifetime assessment in this manuscript. [*Keywords*: Electromigration, dislocation core, plasticity, reliability assessment.]

INTRODUCTION

As interconnects are aggressively scaled, current density continues to increase, thereby accelerating the EM degradation processes. One way the electromigration current could introduce extra damage is through EM-induced plasticity – a phenomenon that has been observed recently both in Al [1] and Cu [2] interconnects. This plasticity could have very important implications both in the efforts to improve EM reliability as well as in the development of methodologies for EM reliability assessment.

Plastic deformation was observed in metallic interconnect test structures [1,2] during in situ electromigration experiments and before the onset of visible microstructural damage (voids, hillock formation). It has been shown using a synchrotron technique with white beam x-ray microdiffraction [3] that almost as soon as the EM current is turned on, grains start to deform plastically. The extent of this EM-induced plasticity in Cu test structures [2] is dependent on the width of the interconnect lines. In wide lines, plastic deformation manifests itself as grain bending and the formation of subgrain structures, while only grain rotation is observed in the narrower lines. It has also been shown that the grain texture of the Cu line might also play an important role.

Furthermore, the grain bending and the subgrain structure formation were observed not in any random direction, but always in the direction across the width of the lines [2]. Upon further investigation, it was also found that a particular slip system in an FCC crystal is responsible for the kind of plasticity observed in the Cu grains, and that particular slip system always has the $\langle 112 \rangle$ line direction almost coinciding with the direction of the electron flow (within 10°).

This observed plasticity hence leads to a concentration of edge dislocations with cores running along the direction of electron flow such as illustrated in Figure 1 below.

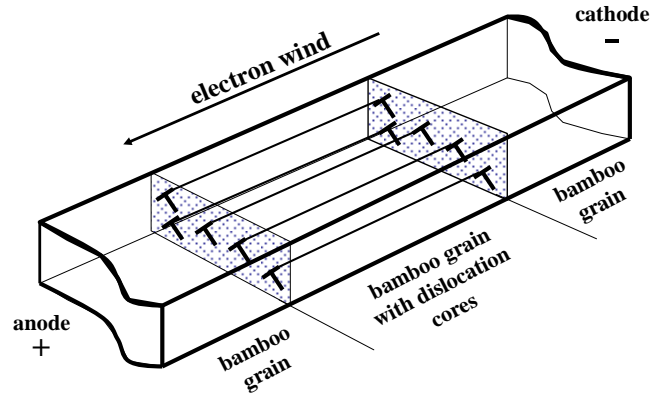


FIGURE 1. SCHEMATIC OF A GRAIN CONTAINING DISLOCATIONS WITH CORES RUNNING ALONG THE DIRECTION OF THE ELECTRON FLOW IN THE INTERCONNECT LINES.

Dislocation cores are, in general, already recognized as fast diffusion paths [4], but in this configuration especially, their contribution to the overall migration of atoms from the cathode to the anode end of the line is even more pronounced. When the concentration of these dislocations reaches a high enough level of dislocation density (ρ) in the crystal, their contribution to the overall effective Diffusivity (D_{eff}) can no longer be neglected:

$$D_{eff} = \frac{\delta}{h} D_{int} + \rho \cdot a_{core} \cdot D_{core} \quad (\text{Eq. 1})$$

where a_{core} and D_{core} are the cross-sectional area and diffusivity of a dislocation core, respectively, ρ is the dislocation density, and δ , h and D_{int} are the effective interface diffusion thickness, the height of the line and the diffusivity of the interface, respectively. In this equation, the diffusivities are described in the usual way by $D = D_o \exp(-E_A/kT)$ where E_A is the activation energy, D_o a constant, and k the gas constant.

Diffusion along dislocation cores (“pipe diffusion”) has been commonly included in models of diffusion-controlled deformation in bulk materials [5]. Suo [6] considered the motion and multiplication of dislocations under the influence of an electric current in a conductor line, and suggested that EM-driven dislocation multiplication could itself lead to dislocation densities high enough to affect EM degradation processes. Oates [7], however, did not see

any diffusivity effects that could be attributed to dislocations in his experimental study. Baker et al. [4] through their experimental study of nanoindented Al lines (width = 1 μm , mean grain size = 1.1 μm) showed that the effect of a dislocation density of $10^{16}/\text{m}^2$ is comparable to diffusion through a grain boundary. These studies all essentially suggest that if the dislocation density is sufficiently high, this may affect the overall EM degradation processes in metallic interconnects, and thus could have fundamental implications.

The present paper describes a key piece of experimental evidence that opens up the very possibility that such a high dislocation density is present in the Cu test structures undergoing EM. It then concerns itself with its important implications for EM reliability assessment methodologies.

The synchrotron technique of scanning white beam X-ray microdiffraction has been described in a complete manner elsewhere [3]. It consists of scanning the sample under a submicron size X-ray beam and capturing a Laue diffraction pattern at each step with a CCD detector. Using a small beam allows us to consider each grain of the interconnect sample as a single crystal. The indexing of the Laue pattern gives the orientation of the grain while the shape of the Laue peaks yields information regarding plastic deformation of the individual grain.

EXPERIMENTAL

The interconnect test structure used in this study is an electroplated Cu damascene line fabricated by AMD. The test line is of 200 μm in length and approximately 0.2 μm in thickness, and 0.5 μm in width. The line is passivated with carbon-based CVD oxide (low-k). Both vias at either end of the line connect to a lower metallization level, which in turn connects to unpassivated bond pads which are used for electrical connection.

The white beam x-ray microdiffraction experiment was performed on beamline 7.3.3. at the Advanced Light Source, Berkeley, CA. The electromigration test was conducted at 300°C. Current and voltage were monitored at 10s increments. The sample was scanned in 0.5 μm steps. A complete set of CCD frames takes about 6 to 7 hours to collect. The exposure time was 20 s plus about 10 s of electronic readout time for each frame. In this manner the Laue pattern and information regarding plastic deformation for each grain in the sample was collected for each time step during the experiment. The current was ramped up to 2 mA ($j = 2 \text{ MA}/\text{cm}^2$) and then set at that value for the rest of the test (up to 36 hrs).

Another interconnect test structure was also prepared with dimensions similar to those of the first one, but with a different ILD material to provide comparison and sense of generalities of the extent of plasticity. The ILD material for this set of samples was SiO_2 based. The line length is 200 μm , the thickness is approximately 0.25 μm and the width is 0.7 μm .

RESULTS AND DISCUSSIONS

We first describe the in situ electromigration observations on both of the Cu damascene test structures. Figures 2(a) – (c) show the typical evolution of the Laue diffraction spots during the in situ EM test for 36 hours. We find that as the EM test progresses, most grains in both Cu lines exhibit plasticity, which manifests itself either in the form of diffraction spot broadening (streaking) such as shown in Figures 2(b) and (c), or in the form of diffraction spot splitting into clearly two or even more different spots. The broadening of the diffraction spots represents crystal bending of the Cu grains in the

line, whereas the split diffraction spots indicate the formation of low-angle boundaries or subgrain structures. From the amount of broadening we can calculate the bending of the Cu crystal, and from the amount of splitting, the angle of misorientation.

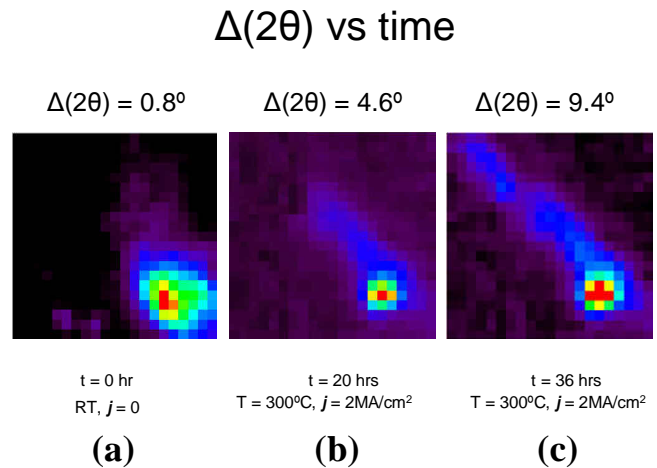


FIGURE 2. THE TYPICAL EVOLUTION OF LAUE DIFFRACTION SPOTS FROM THE CU CRYSTALS DURING IN SITU OF EM TESTS.

The extent of plasticity such as indicated by the Laue spot streaking shown in Figure 2 is fairly large, and represents a marked departure from our previous observations on a similar set of Cu damascene lines [2]. Figure 3(a) illustrates the extent of the large amount of plasticity observed experimentally in Cu grains in the test structure. In this Laue photograph of the Cu interconnect line with the Carbon-based CVD oxide dielectric (after 36 hours of EM at 300°C and 2MA/cm²) most diffraction spots are shown to be broadened or split to various extents. The one set of diffraction spots that are relatively sharp and rounded belong to the Silicon substrate crystal. The most extreme broadening is shown by a set of Cu diffraction spots marked by red boxes, which represents a Cu grain undergoing intense EM-induced plasticity.

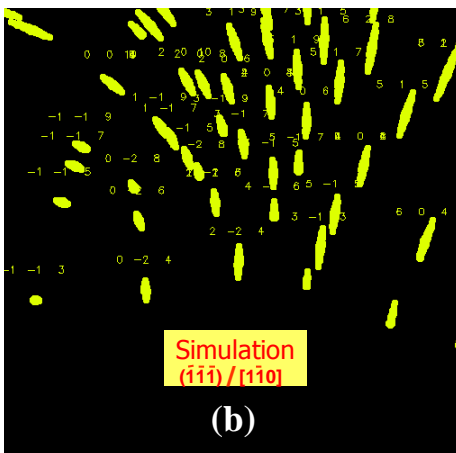
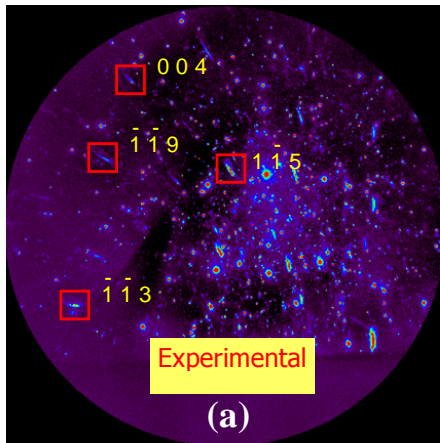


FIGURE 3(A) THE GENERAL PLASTICITY OBSERVED EXPERIMENTALLY IN CU INTERCONNECTS; (B) THE SIMULATED STREAKING LAUE PATTERN OF A PARTICULAR CU GRAIN REPRESENTED BY A SET OF LAUE DIFFRACTION SPOTS WITH RED BOXES IN THE FIGURE ABOVE (FIG. 3(A)).

The nature of plastic deformation exhibited by this particular grain can be further investigated by comparing the initial crystal orientation of the grain (through indexation of the diffraction spots), and a set of possible simulated streaking directions that represent the 12 possible plastic bending events in the FCC crystal. If one of the simulated streaking directions match with the actual streaking directions observed in Figure 3(a), we could infer the active slip system of the plastic deformation. The out-of-plane orientation of this particular Cu grain is close to $\langle 111 \rangle$, which is consistent with the typical texture of Cu lines fabricated by AMD [8]. In this approximately $\langle 111 \rangle$ oriented grain, the simulation (Figure 3(b)) suggests that the $(\bar{1}\bar{1}\bar{1})/[\bar{1}\bar{1}0]$ slip system is active, causing the crystal to deform plastically and to lead to the pattern of the diffraction spot streaking. The specific $\langle 112 \rangle$ line direction of this plastic deformation, which essentially becomes the rotation axis of the crystal bending, is observed to be $[-1-12]$ and that it is very close to the direction of the electron flow (off by 7.9°) in the interconnect line. This observation holds true (within 10°) throughout the many grains along the Cu line, and is consistent with our previous observation [2]. This further confirms that the observed plasticity leads to the concentration of edge dislocations in Cu grains with cores running along the direction of electron flow throughout the full length of the interconnect lines.

Grain orientation mapping of these Cu lines unfortunately could not be obtained in the present study. The X-ray spot size ($0.8 \times 0.8 \mu\text{m}$) currently used in ALS beamline 7.3.3 was relatively large for the dimensions of these state-of-the-art interconnect lines. That makes diffraction spot indexation often very difficult and thus mapping of grain orientations and other further quantitative analyses unreliable. The one Cu grain that we discussed and analyzed above was one of the few grains in the two Cu lines for which indexation of the diffraction spots happens to be sufficiently clear and unambiguous for this analysis. In general, the larger the Cu grains, and the more bamboo-like they are, the more they diffract sharply and give numerous diffraction spots, thus giving higher confidence on the reliability of these results. That being said, however, it is fortunate that we could still always compare qualitatively the evolution of Cu diffraction spots before and after some period of EM testing, as has been demonstrated in Figure 2.

Low-k vs Hybrid Dielectrics

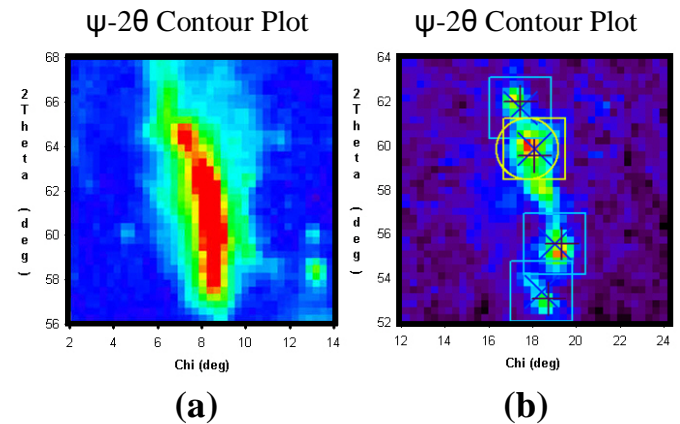


FIGURE 4. THE LAUE PEAK STREAKING/SPLITTING OBSERVED FROM CU INTERCONNECT TEST STRUCTURES WITH (A) LOW-K, AND (B) HYBRID DIELECTRICS; IN ψ - 2θ SPACE/CONTOUR INTENSITY PLOT.

Figure 4(a) and (b) show still different additional diffraction spots observed during this experiment (after EM testing of 36 hours, at 300°C and $2\text{MA}/\text{cm}^2$ current loading) coming from the Cu lines with the low-k and the hybrid dielectrics, respectively. The diffraction spots have been converted to ψ - 2θ space, with ψ running along the direction of the length of the line, and 2θ across the direction of the width of the line. We can then use the broadening and the spot splitting observed to obtain information about the dislocation structure induced into the grain by electromigration. For instance, from the streak length of Figure 4(a), as measured in the CCD camera, and the sample to detector distance we obtain the curvature angle of the grain of 9.8° . Assuming a near bamboo structure, the grain width is the same as the width of the line ($0.5 \mu\text{m}$), from which we get the radius of curvature of the grain, $R = 2.34 \mu\text{m}$. The geometrically necessary dislocation density to account for the curvature observed can be calculated from the Cahn-Nye relationship $\rho = 1/Rb$ where b is the Burgers vector. The geometrically necessary dislocation density is then $\rho = 1.68 \times 10^{15}/\text{m}^2$. The total number of dislocations in the area of the cross-section of the Cu line/grain is approximately 142.

To obtain quantitative information on polygonization walls (subgrain boundaries) from the spot split in Figure 4(b) we observe that the Laue spot splitting, $\Delta_{\text{total}} = 9.1^\circ$. From this misorientation

and Burgers' model of a small angle grain boundary $\theta = b/L$, where L = dislocation spacing, we find $L = 16 \text{ \AA}$ which amounts to a total of 110 dislocations in the subgrain boundaries in the cross-section of the Cu line/grain. The extent of the plasticity such as described here is typical of that observed in both the Cu lines. The significance of the difference, in terms of the extent of plasticity between the two Cu lines (with low- k and hybrid dielectrics), requires added confirmation. Nevertheless they provide a sense of the generality of the extent of plasticity in these Cu lines.

We now discuss the implications of these observations. We have just established that dislocations with cores running along the electron flow direction and densities in the order of $10^{15}/\text{m}^2$ are present in Cu lines undergoing EM (accelerated test conditions) for 36 hours. This level of ρ (the red line in Figure 5) is just above the threshold of dislocation density necessary for the dislocation core diffusion to be on par with the interface diffusion (the yellow line) at the test conditions ($T = 300^\circ\text{C}$ or in other words, $1000/T = 1.75/\text{K}$). In other words, at this dislocation density we can expect the contribution of dislocation cores to the overall/effective diffusivity in the Cu line during accelerated EM to be at least the same order of magnitude as interface diffusion. This is illustrated in Figure 5 using diffusion coefficient values relevant for the Cu lines used in the present study, available from the literature [5,9]. It is to be noted, however, that at temperatures below 100°C , it takes ρ in the order of $10^{17}/\text{m}^2$ (the green line) for the effect of dislocation cores to be significant. These lower temperatures correlate with the typical use conditions of the device/interconnects. The typical initial dislocation density in Cu/metallic lines was taken to be $10^{12}/\text{m}^2$ [4] and the corresponding diffusivity is as shown by the blue line.

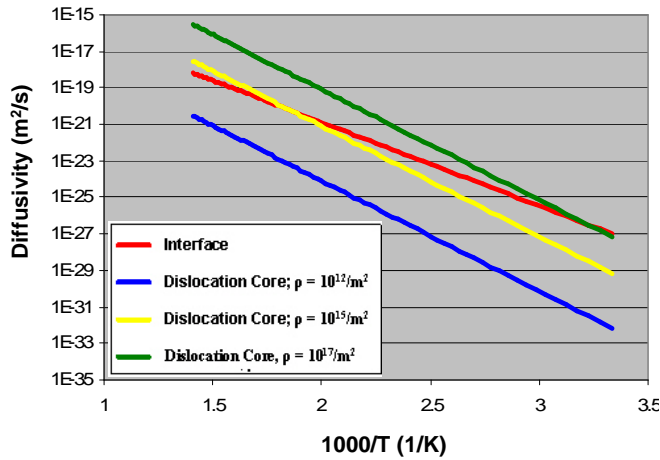


FIGURE 5. COMPARISON OF DIFFUSIVITIES AS A FUNCTION OF TEMPERATURE BETWEEN THE INTERFACE DIFFUSION PATH AND THOSE OF DISLOCATION CORES OF DIFFERENT DENSITIES IN CU INTERCONNECT LINES (FROM $10^{12}/\text{M}^2$, TO $10^{15}/\text{M}^2$ TO $10^{17}/\text{M}^2$) WHEN EACH DIFFUSION MECHANISM IS ASSUMED TO ACT ALONE. DIFFUSIVITIES WERE CALCULATED USING VALUES IN [5,9]

Furthermore, if ρ should increase with j , then we will find that D_{eff} should also increase with j . Consequently, there will be an extra EM flux, and thus an extra reduction in the time to failure of the device with increasing j . This is an extra dependency on j , which would manifest itself in the value of the current density exponent, n (in Black's equation), being > 1 .

$$t_f = A \left(\frac{1}{j} \right)^n \exp \left(\frac{E_A}{kT} \right) \quad (\text{Eq. 2})$$

The fact that n is usually found in real cases to be > 1 (as opposed to $n = 1$ for the prevailing model of void growth limited failure) suggests that this extra dependency on j , especially at high temperatures of the test conditions, could be due to dislocation core diffusion. In other words, the higher n could be traced back to the higher level of plasticity in the crystal, and the closer n is to unity, the less plasticity must have influenced the electromigration degradation process.

Kirchheim and Kaerber [10] experimentally observed the *MTF* dependency on current density, j , in an Al conductor line, for a wide range of j , such as shown in Figure 6 (the black-colored dots with error bars were the original data points). It clearly shows that at low current densities, the *MTF* data is best fit by $n = 1$ (blue dotted line), while at higher current densities, the *MTF* data is better fit by $n > 1$ (red dotted line). Kirchheim and Kaerber however suggested in their paper that these deviations occurring at higher current densities might have been caused by Joule heating[10].

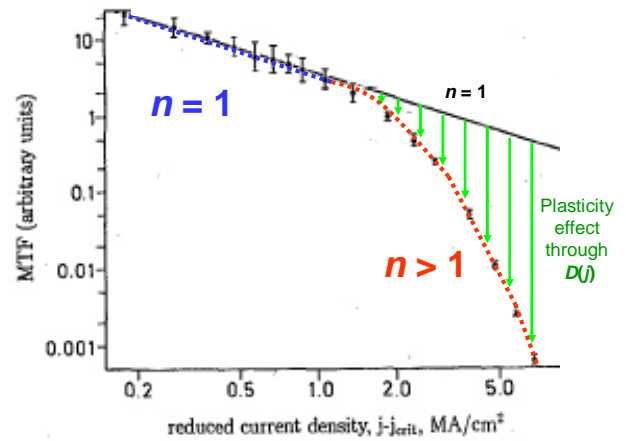


FIGURE 6. KIRCHHEIM AND KAEBER'S EXPERIMENTAL *MTF* DATA AS A FUNCTION OF REDUCED CURRENT DENSITY, $J-J_{\text{CRIT}}$ (ALL THE BLACK-COLORED FEATURES; COURTESY OF [10]). THE BLUE, RED AND GREEN COLORED FEATURES ARE ADDED TO LEAD TO OUR ARGUMENT.

Plasticity, especially in the form described here, could just as likely be the source of such deviations of *MTF* dependency on j at high current densities. As j increases, plasticity also increases leading to increasingly higher EM fluxes (the green line in Figure 6), and thus increasingly lower *MTF*, and therefore eventually a current density exponent, $n > 1$ has to be used to fit the failure time distribution.

However, under use conditions where the temperature is much lower (eg. 100°C), the level of ρ associated with such elevated diffusivity is almost impossible to reach, so that this plasticity-amplified diffusivity is associated only with the high temperature and high current density of the accelerated EM test. In other words, there is not likely to be much plasticity under use conditions, and thus the diffusivity is dominated only by interface diffusion, and ultimately the failure time distribution for various use conditions should be as

expected from Black's equation using $n=1$. This is consistent with the observations of Kirchheim and Kaerber [10] discussed above.

This interpretation of the Kirchheim and Kaerber data is consistent with the physical model (void growth limited failure) which has also been observed through *in situ* EM studies on similar material [11] by Zschech *et al.*

It can be further stated that plasticity-amplified diffusivity is simply an extra mode of deformation under test conditions (which is not typically present under use conditions), and that its effect is wholly captured in the n value being greater than unity. This plasticity-inflated n could thus lead to inaccurate extrapolations of lifetimes under use conditions.

To improve the accuracy of the reliability assessment of devices under use conditions, we therefore propose that the effect of plasticity has to be removed first from the EM lifetime equation. This can be done simply by insisting on $n=1$ in our lifetime assessment (i.e., j_{max} calculation) which in most typical EM test conditions will result in a more conservative prediction of device lifetime.

CONCLUSIONS

We use the synchrotron-based white beam Laue X-ray microdiffraction technique to investigate electromigration-induced plasticity in Cu interconnect structures undergoing electromigration testing. We discovered that the extent and configuration of dislocations in the Cu grains induced during this accelerated EM testing could lead to another competing EM diffusion mechanism in addition to interface diffusion. We have suggested that this plasticity effect can be correlated to the measured value of current density exponent, n , in Black's equation. We have observed that this correlation would then lead to an important implication for the way device lifetime/reliability is assessed.

ACKNOWLEDGMENTS

The authors would like to thank Advanced Micro Devices (AMD) for generous support and sample assistance. One of the authors (ASB) more specifically would like to thank John M. Ennals, AMD/SRC Program Manager, Technology Research Group, AMD, for the opportunity of a Summer Internship Program in 2006. Both ASB and WDN gratefully acknowledge support by the U.S. Department of Energy, Office of Basic Energy Sciences through Grant No. (DE-FG02-04ER46163). The Advanced Light Source (ALS) is supported by the Director, Office of Science, Office of Basic Energy Sciences, of the U.S. Department of Energy under Contract No. DE-AC02-05CH11231 at the Ernest Orlando Lawrence Berkeley National Laboratory (LBNL). ASB would also like to acknowledge Jun He of Intel Corporation for valuable discussions during the 2006 Gordon Research Conference (GRC) on Thin Film and Small Scale Mechanical Behavior.

REFERENCES

- [1] B. C. Valek *et al.*, *Appl. Phys. Lett.*, **81**, 2002. pp. 4168.
- [2] A. S. Budiman *et al.*, *Appl. Phys. Lett.*, **88**, 2006, 233515.
- [3] N. Tamura *et al.*, *J. Synchrotron Rad.*, **10**, 2003. pp. 137.

- [4] Baker, S.P., Joo, Y.-C., Knaub, M.P., Artz, E., *Acta mater.*, 2000, **48**, 2199
- [5] H.J. Frost and M.F. Ashby, in *Deformation-Mechanism Maps: The Plasticity and Creep of Metals and Ceramics*. Pergamon Press, Oxford, 1982, p.21
- [6] Suo, Z. *Acta metall. mater.*, 1994, **42**, 3581
- [7] Oates, A.S., *J. Appl. Phys.*, 1996, **79**, 163
- [8] Besser *et al.*, *J. Elec. Matls.* **30**, No. 4, p. 320 (2001)
- [9] Gan, D., Ho, Paul S., Pang, Y., Huang, Rui, Leu, J, Maiz, J., Scherban, T., *J. Mater. Res.*, 2006, **21**, No. 6, p.1512
- [10] Kirchheim & Kaerber, *J. Appl. Phys.*, **70**, 1991, pp. 172.
- [11] Zschech *et al.*, *Mat. Res. Soc. Proc.* **812**, 2004. F.7.5.1.