Electromigration-Induced Plastic Deformation in Cu Interconnects: Effects on Current Density Exponent, n , and Implications for EM Reliability Assessment

A.S. BUDIMAN,^{1,3,6} C.S. HAU-RIEGE,² W.C. BAEK,³ C. LOR,³ A. HUANG,³ H.S. KIM,³ G. NEUBAUER,³ J. PAK,³ P.R. BESSER,⁴ and W.D. NIX⁵

1.—Center for Integrated Nanotechnologies (CINT), Los Alamos National Laboratory (LANL), Los Alamos, NM 87544, USA. 2.—Qualcomm, Santa Clara, CA 95051, USA. 3.—Technology Reliability Engineering (TRE), Spansion Inc., Sunnyvale, CA 94088, USA. 4.—Copper Integration & Reliability, Unity Semiconductor, Sunnyvale, CA 94085, USA. 5.—Department of Materials Science & Engineering, Stanford University, Stanford, CA 94305, USA. 6.—e-mail: suriadi@stanfordalumni.org

While Black's equation for electromigration (EM) in interconnects with $n = 1$ is rigorously based on the principles of electrotransport, $n > 1$ is more commonly observed empirically. This deviation is usually attributed to Joule heating. An alternative explanation is suggested by the recent discovery of EM plasticity. To examine this possibility, we have retested samples that had been previously subjected to a predamaging phase of high temperature and current densities to determine whether the loss of median time to failure (MTF) is retained. We find that the predamaged samples exhibit MTFs that are permanently reduced, which is a characteristic of EM plasticity.

Key words: Electromigration, plasticity, current density exponent

INTRODUCTION

The following empirical relationship for median time to failure (MTF) in electromigration (EM) reliability assessment was proposed by Black¹:

$$\text{MTF} = A \left(\frac{1}{j} \right)^n \exp \left(\frac{E_A}{kT} \right), \quad (1)$$

where j is the current density, n is the current density exponent, A is an empirical constant, and E_A , k , and T have their usual meanings in mass transport. While this equation is widely used, the value for the current density exponent and its implications for EM lifetime prediction are still much debated.^{2–8} Under the common atomistic description of electrotransport,⁹ the electromigration flux in a metallic line is proportional to the current density and the product of flux and time (i.e., MTF) and corresponds to the removal of a certain volume of matter per unit length (i.e., the

cross-sectional area) of the interconnect line, which is a necessary condition for failure. Thus MTF is proportional to j^{-1} (i.e., $n = 1$). This description is usually associated with a void-growth-limited failure mode^{4,10} (as opposed to a void-nucleation-limited mode), which has also recently been supported by experimental observation.¹¹

The fact that n is usually found to be larger than 1^{3–7} suggests that this extra dependency on j , especially at high temperatures, could be due to some artificial effects. Joule heating is widely believed to be the source of this deviation, as suggested by Kirchheim and Kaeber,² especially in the case of Al interconnects. However the recent discovery of EM-induced plasticity,^{12–14} wherein plastic deformation leads to new paths for EM transport, could also cause this deviation.⁸

Plastic deformation behavior in metallic interconnect lines during EM experiments has been observed in both Al^{12,14} and Cu interconnects,^{8,13} using a synchrotron technique involving white-beam x-ray microdiffraction that can function as a local probe of plastic deformation in crystals that compose the interconnect.^{15–18} The extent and configuration of

dislocations induced during accelerated EM testing in a set of Cu interconnect lines have been suggested to lead to another, competing EM diffusion path, in addition to interface diffusion in Cu.⁸ An increase in the current density is then suggested to lead to an increase in the dislocation density, ρ , which should lead to an increase in D_{eff} (the overall/effective diffusivity of the EM process). Consequently, there would 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 being $n > 1$.

We are suggesting that the higher n could be traced back to the higher level of plasticity in the lines; the closer n is to unity, the less plasticity must have influenced the EM degradation process. This plasticity effect could thus be correlated with the measured value of the current density exponent and could have important implications for the way device lifetime/reliability is assessed. In the present study, we set out to examine this possibility by investigating the permanent plasticity effects on the current density exponent.

EXPERIMENTAL PROCEDURES

We first investigated the current density exponent as a function of current density, i.e., $n(j)$, through a series of EM test experiments involving various current densities from 0.5 MA/cm² to 9.5 MA/cm². We then investigated the plasticity effects on the current density exponent by comparing $n(j)$ for two sets of otherwise identical Cu interconnect lines:

1. A set of “fresh” samples (control);
2. A set of predamaged samples: fresh samples that had been previously subjected to accelerated EM testing (at $T = 350^\circ\text{C}$ and $j = 3.5 \text{ MA/cm}^2$ for a brief period of 50 h), i.e., the predamaging phase.

The interconnect test structure used in this study is a variation of a back-end-of-line (BEoL) process for a 65-nm complementary metal–oxide–semiconductor (CMOS) technology fabricated in the Submicron Development Center (SDC) facility of Spansion, Inc. In this technology, the dual-damascene Cu fill process includes a standard Ta-based barrier and Cu seed, electroplated Cu fill, postplating anneal, chemical–mechanical polish, a dielectric cap layer, and a standard surface pretreatment. The interlayer dielectric (ILD) was fluorine-doped tetraethyl orthosilicate (FTEOS). In these EM tests, the current was forced from a lower metal layer into a narrow upper metal test line, typically designed to force failure in the upper metal line at its critical dimension.

The test structures consist of 200- μm -long lines, approximately 0.2 μm thick and 0.5 μm wide. The fresh sample set simply consists of 240 statistically identical Cu interconnect lines. The predamaged

samples were 240 otherwise identical interconnect lines, except that they had been subjected to the predamaging phase. Based on our earlier observations on similar Cu interconnect samples^{8,13} such predamaging would result in the microstructures of Cu grains having dislocations lining up along the length of the interconnect lines with a density (ρ) of up to $10^{15}/\text{m}^2$. Evidence of such effects on Cu microstructures has been documented on Cu interconnect samples from various manufacturers^{8,13} using the synchrotron x-ray microdiffraction technique. Reference 8 is most relevant to our present study as it specifically describes the dislocation density and configuration, as shown in Fig. 1a and b, c, respectively, on Cu lines very similar to those used in the present study. The Cu lines in Ref. 8 were fabricated by the Submicron Development Center (SDC) when it was still officially part of AMD in 2006, while the Cu lines of our present study were fabricated by SDC as it is currently part of Spansion, Inc. Both of them were fabricated using the same 65-nm technology node of SDC. The length of the interconnect test structure (i.e., 200 μm) used in the present study was chosen to ensure mortality in the EM experiments even for the smallest current (the jL values in the present experiments are never smaller than 10^4 A/cm) following studies by Hau-Riege et al.¹⁹ using samples fabricated by AMD.

Thus, the main difference between the two sets of samples is in the initial microstructures of the Cu lines. The first set would have typical microstructures of annealed Cu grains, while the second set is expected to have plastically deformed microstructures.

Electromigration (EM) tests were then performed at current densities of $j = 0.5 \text{ MA/cm}^2$, 1 MA/cm^2 , 1.5 MA/cm^2 , 2 MA/cm^2 , 2.5 MA/cm^2 , 3.5 MA/cm^2 , 4.5 MA/cm^2 , 5.5 MA/cm^2 , 6.5 MA/cm^2 , 7.5 MA/cm^2 , 8.5 MA/cm^2 , and 9.5 MA/cm^2 at a high temperature of 350°C . For each current density, the number of samples tested was 20 from each of the two sets of samples for statistical purposes. There are thus 240 samples in total (12 current densities, 20 samples each) in each set of samples involved in this experiment. The failure criterion used was a resistance increase $> 10\%$. No significant resistance increase (not more than 0.2%) was observed during the predamage phase of 50 h. This experiment involved a fairly extended testing time, especially for the tests at lower current densities ($< 1 \text{ MA/cm}^2$).

RESULTS AND DISCUSSION

We begin by describing the EM test results of the fresh set of samples. Figure 2a shows the EM failure time as a function of current density. The median time to failure (MTF) is represented by the blue solid circle, while the error bar represents the range of failure times from the 20 samples at each of the test conditions. All failure times are normalized with respect to the minimum MTF

Fig. 1. Experimental observations of EM-induced plastic deformation in Cu interconnect samples⁸ very similar to those used in the present study: (a) two representative sets of Cu Laue diffraction spots each from before (initial) and after EM loading of $j = 2 \text{ MA/cm}^2$ for 36 h leading to the observed dislocation density of $10^{15}/\text{m}^2$; (b) along the interconnect line, a series of dislocations lining up with the direction of the length of the Cu line has been observed,⁸ as illustrated in three dimensions (3D) in (c). *Grain map estimated based on Laue diffraction observation.

observed in the present study. No actual failure times are given here for proprietary reasons. The current densities are also normalized with respect to the minimum value in the present study, which is 0.5 MA/cm^2 . The data can be broadly divided into two groups: low current densities (with $\ln(j/j_{\min}) < 1.6$, or $j < 2.5 \text{ MA/cm}^2$) and high current densities (with $j > 2.5 \text{ MA/cm}^2$). As seen in Fig. 2a, the data in the low j range are best fit by a straight line with a slope of -1.1 , which indicates $n = 1.1$ in Black's equation

Fig. 2. Electromigration (EM) test data/results from the fresh set of Cu samples showing: (a) $n = 1.1$ in the low j range, whereas $n = 1.7$ in the high j range, or alternatively (b) that n is principally 1 (black solid line) but that MTF tends to be depressed in the high j range (black dotted line) due to extrinsic effects (dashed line/arrow), as suggested by Kirchheim and Kaerber.² Note that the EM failure time data (median and range; blue solid circle and error bar) as a function of j are identical between (a) and (b). (Color figure online.)

(1), whereas in the high j range, $n = 1.7$. These data are consistent with the trend in Cu EM reported recently by several researchers,^{5–7} most notably by Hu et al.,⁷ who reported $n = 1.1$ and 1.8 for low and high current densities, respectively, under similar EM test conditions.

Alternatively we may also suggest that n is principally 1 (especially true in the lower j range, as indicated by the solid line in Fig. 2b), but that the MTF tends to be depressed in the high j range, which, in turn, causes the n value to deviate from 1. Kirchheim and Kaerber² attributed this extra depression of MTF in the high j range to a Joule heating effect. Our earlier studies^{12–14} have also suggested electromigration-induced plasticity that can lead to new paths for EM transport which could also be responsible for this deviation.⁸ It is certainly within the range of possibilities that either one of

these effects dominates or that both effects contribute to various different extents to the MTF deviation from the $n = 1$ line at the high j range.

In these particular samples, the plasticity effect argument is certainly feasible given the level of plastic damage that has been reported earlier in similar samples⁸ (all the geometries and materials set are nominally the same) manufactured by the same wafer fabrication facility, as is equally the Joule heating effect argument. However other Cu interconnect lines (with perhaps different materials set, or different geometries, or different fabrication methodologies, etc.) may have different levels of plastic damage upon similar extent of electromigration loading and thus potentially different extents of contribution to the MTF deviation from the $n = 1$ line at the high j range.

For instance, our own previous study of Cu interconnect lines made by Intel Corporation¹³ reported a much lower level of plastic damage ($\rho \approx 10^{13}/\text{m}^2$); our quick calculation following a methodology similar to that described in Ref. 8 suggests that such a low level of dislocation density would only generate diffusivity two orders of magnitude lower than that of the Cu/dielectric interface diffusion of the samples at the elevated temperature. This thus indicates that the plasticity effect here in these Intel Cu interconnect samples¹³ could not possibly be very substantial, and therefore practically all the MTF deviation from the $n = 1$ line at the high j range may have been caused by the Joule heating effect alone.

However, as far as the present set of Cu interconnect samples are concerned, both arguments are feasible, and it could just simply be that both contribute to different extents, perhaps one more than the other. However, if this deviation were caused by Joule heating alone then the effects would not be permanent; if the same sample were subsequently subjected to EM testing with low j , the degradation in failure times should not be retained. This hypothesis motivated us to study the second set of samples, i.e., the predamaged samples. If Joule heating is wholly responsible for the deviation from $n = 1$, then the effects of the predamaging phase would not be permanent and the failure times for this second set of samples should be about the same as for the fresh samples. On the other hand if the predamaging phase causes a permanent change in the diffusional pathways, then the effects would be permanent and the degraded failure times would be retained for EM testing with low j (or, in other words, the failure times for the second set of samples should be significantly lower than those of the fresh samples). This is the main focus of the present study.

EM test results for the two sets of samples (fresh versus predamaged) are shown in Fig. 3a. The median times to failure (MTFs) here are represented by the solid features (blue circles for fresh samples and red squares for predamaged ones),

Fig. 3. Electromigration (EM) test data/results showing (a) comparison between fresh versus predamaged samples, and (b) that n is principally 1 in both cases (black solid line for the fresh samples, red dashed line for the predamaged samples), but MTF tends to be depressed at high j range due to extrinsic effects and thus deviates (black dotted line for the fresh samples, red dash-dotted line for the predamaged samples) from the $n = 1$ trajectories; The green "hypothetical" data points (i.e., the "+" and "-" signs) indicate the shortened lifetime of the predamaged samples if the predamaging phase degrades the MTF more than its nominal 50 h due to very aggressive void growth. (Color figure online.)

while the error bars represent the ranges of the failure times from the 20 samples from each group at each of the test conditions. All failure times are again normalized with respect to the minimum MTF observed in the present study. The current densities are also normalized with respect to the minimum value, i.e., 0.5 MA/cm². It is clear from these data sets that there is a significant difference in time to failure between samples from the two different groups, especially in the low j range ($\ln(j/j_{min}) < 1.95$ or $j < 3.5$ MA/cm²).

In Fig. 3b, we consider just the MTFs of two sets of samples (the solid data points, without the error bars for clarity). The significant difference in the MTFs between the two sets of samples in the low

j range suggests that Joule heating alone cannot be responsible for the deviation of MTF from the $n = 1$ line at the high j range such as shown in Fig. 2b. However, since the predamaging phase was done at accelerated conditions of high temperature and high current density (at $T = 350^\circ\text{C}$ and $j = 3.5 \text{ MA/cm}^2$), albeit for a very brief period of time (50 h), it could perhaps cause very aggressive void growth such that the failure times for the second set of samples become somewhat lower (beyond the nominal 50 h of time) than those of the fresh samples. To account for this equivalent lifetime used by the predamaging phase, we now add the green “hypothetical” data points (i.e., the “+” and “-” signs) to indicate the shortened lifetime of the predamaged samples if the predamaging phase degrades the MTF more than its nominal 50 h due to very aggressive void growth.

The “equivalent” lifetimes used by the predamaging phase (the green “+” and “-” signs) here were calculated using the proportionality assumption [i.e., $\text{MTF} \cdot (j)^n = \text{constant}$; $\text{MTF}' = (j/j')^n \text{MTF}$; for the “+” signs, as the predamaging phase was actually done at $j = 3.5 \text{ MA/cm}^2$ for 50 h, the “equivalent” lifetime used at $j = 0.5 \text{ MA/cm}^2$ is 350 h, that is 7 times— 3.5 MA/cm^2 divided by 0.5 MA/cm^2 with $n = 1$ —the nominal 50 h]. Even if we assume extremely aggressive void growth during the predamaging phase and thus introduce a value of $n = 1.7$, such as actually determined from the data shown in Fig. 2a for the conditions in the predamaging phase, into the above proportionality calculation, the hypothetically degraded MTF would then be shown as the green “-” signs in Fig. 3b. Evidently, there remains a significant gap with the MTFs of the second set of samples.

These data sets thus strongly indicate that Joule heating cannot act alone here in these Cu interconnect samples. In the range of $\ln j/j_{\min} < 1.95$, the MTFs of the predamaged samples (the red square data points) are still significantly reduced from those of the fresh samples even after considering the shortened lifetime due to the predamaging phase (the green “+” and “-” signs). Such a significant difference can be explained by EM-induced plasticity, which introduces some permanent effects, perhaps in addition to the Joule heating effect. The significant difference here strongly suggests that the second set of samples have also suffered substantially higher EM fluxes than those of the first set of samples, even though both were tested at the same low j , which further indicates there might be effects of a structural permanent difference between the two sets of samples. We believe that the high- j predamaging phase for 50 h had created dislocation configurations at such high densities that they had accordingly aggravated the EM fluxes in the second set of samples beyond the nominal 50 h or even beyond the hypothetically degraded lifetimes due to the aggressive/extremely aggressive void growth scenarios during the predamaging phase.

Even though Joule heating is widely cited as the source of the deviation from $n = 1$ at high j , recent studies,^{20–24} both experimental as well as computational, have shown that its effects on the global transport of atoms along the metal interconnect lines are somewhat modest. In the absence of extreme local instabilities (such as hot spots or local meltdown due to current-crowding effects, for instance), the effect of Joule heating is predicted to be merely a rise of between 5% and 10% at high j compared with at low j in the global temperature of the interconnect lines.^{20–25} This is insufficient to cause a large drop in MTF at high j that would be required to cause n to deviate significantly from 1. This is especially true for the case of the Cu-SiO₂ interconnect scheme, such as used in the present study, due to the high thermal conductivity of SiO₂.²¹ For the Cu-SiO₂ interconnect scheme, Ref. 21, for instance, using a combination of an analytical thermal model with a two-dimensional (2D) numerical simulation using the finite-element method (FEM), reported a less than 7% rise in global interconnect temperature at $j = 4.5 \text{ MA/cm}^2$ compared with at $j = 0.5 \text{ MA/cm}^2$ and consequently a factor of less than 2 in the MTF reduction due to the temperature rise. The corresponding test conditions in the control data sets (i.e., the fresh samples) in the present study show at least a factor of 20 in the MTF reduction.

To complete the analysis of the MTFs between the two data sets in Fig. 3b, we note that, at higher current densities, the difference in failure times tends to get smaller, until eventually in the special case of $\ln j/j_{\min} = 1.95$ or at $j = 3.5 \text{ MA/cm}^2$ (i.e., the j at which the predamaging phase was done), the difference should theoretically be only 50 h (i.e., the duration of the predamaging phase). That is, the predamaged samples should fail just 50 h earlier than the fresh samples, as both the predamaging phase and the subsequent EM testing here were conducted under the same conditions ($j = 3.5 \text{ MA/cm}^2$ and $T = 350^\circ\text{C}$). This is indeed what is shown by the green “hypothetical” data points (both green signs—the “+” and the “-”—are exactly on top of each other) in Fig. 3b at $\ln j/j_{\min} = 1.95$ or $j = 3.5 \text{ MA/cm}^2$. The green data points there indicate the shortened lifetime of the predamaged sample (the blue circle data point) if the predamaging phase simply used up its nominal lifetime of 50 h, and Fig. 3b shows evidently that it coincides very closely with the actual predamaged sample data (the red square data point). The green “hypothetical” data points are not given in Fig. 3b for the range of $\ln j/j_{\min} > 1.95$ or $j > 3.5 \text{ MA/cm}^2$ for clarity purposes, but they showed the same consistent trend (within the experimental error margin of the experiments) as expected.

Finally, having recognized that any value of n larger than 1 obtained from accelerated test conditions (i.e., high j values) is due to extrinsic effects, we reiterate the danger of overestimating device

lifetime using the current methodology. If, for instance, we use $n = 1.7$ as observed in the present study to extrapolate from the accelerated condition (high j) to the use condition (low j), that extrapolation would clearly lead to overestimation of the actual device lifetime (approximated by the actual MTF data point at low current density). To improve the accuracy of the reliability assessment of devices under use conditions, we thus propose that the extrinsic effect has to be removed from the EM lifetime equation. This can be done simply by insisting on $n = 1$ in our lifetime assessment, which in most typical EM test conditions will result in a more conservative prediction of device lifetime. This is true no matter whether Joule heating alone or, as we have proposed in the present study, EM-induced plasticity (in addition to Joule heating) is the root cause of deviation of MTF at high j in the electromigration of Cu interconnects.

CONCLUSIONS

We recognize the extrinsic effects leading to the artificially large (> 1) value of n obtained from the accelerated test conditions following the current methodology of EM lifetime prediction. By splitting Cu interconnect samples into two groups (fresh versus predamaged) in the present study, we studied and confirmed the signature behavior of EM-induced plasticity in causing a permanent effect on the MTF in this particular set of Cu interconnect samples. We then propose to insist on using $n = 1$ in EM lifetime prediction to avoid the danger of overestimating device reliability.

ACKNOWLEDGEMENTS

The authors would like to thank Spansion, Inc. for experimental resources and for fabricating the samples, as well as Advanced Micro Devices (AMD) for a valuable summer internship opportunity for one of the authors (A.S.B.) in 2006, during which the core idea of this study was developed in collaboration with co-authors C.S.H. and P.R.B., when both were still with AMD. Professor B.M. Clemens of Stanford is also acknowledged for having suggested some of the comparisons in the present study. A.S.B. is currently supported through the Los Alamos National Laboratory (LANL) Director's Research Fellowship Program and by the US Department of Energy, Office of Basic Energy Sciences under Grant No. DE-AC52-06NA25396. Both A.S.B. and W.D.N. gratefully acknowledge support by the US Department of Energy, Office of Basic Energy Sciences through Grant No. (DE-FG02-04ER46163).

OPEN ACCESS

This article is distributed under the terms of the Creative Commons Attribution Noncommercial License which permits any noncommercial use, distribution, and reproduction in any medium, provided the original author(s) and source are credited.

REFERENCES

1. J.R. Black, *6th Annual IEEE Intl. Rel. Phys. Symp. Proc.* (1967), p. 148.
2. R. Kirchheim and U. Kaeber, *J. Appl. Phys.* 70, 172 (1991).
3. M. Shatzkes and J.R. Lloyd, *J. Appl. Phys.* 59, 3890 (1986).
4. C.S. Hau-Riege, *Microelectron. Reliab.* 44, 195 (2004).
5. C.K. Hu, R. Rosenberg, H.S. Rathore, D.B. Nguyen, and B. Agarwala, *IITC Proc.* (1999), pp. 267–269.
6. A. Roy and C.M. Tan, *J. Appl. Phys.* 103, 093707 (2008).
7. C.K. Hu, L.M. Gignac, B. Baker-O'Neal, E. Liniger, R. Yu, P. Flaitz, and A.K. Stamper, *9th Int. Workshop on Stress-induced Phenomena in Metallization*, ed. S. Ogawa, P.S. Ho, and E. Zschech (2007), pp. 27–41.
8. A.S. Budiman, P.R. Besser, C.S. Hau-Riege, A. Marathe, Y.-C. Joo, N. Tamura, J.R. Patel, and W.D. Nix, *J. Electron. Mater.* 38, 379 (2009).
9. J.D. Verhoeven, *Metall. Rev.* 8, 311 (1963).
10. C.S. Hau-Riege, A.P. Marathe, and V. Pham, *2002 Adv. Metall. Conf. Proc.* (2002), p. 169.
11. E. Zschech, M.A. Meyer, and E. Langer, *Mater. Res. Soc. Proc.* 812, F.7.5.1 (2004).
12. B.C. Valek, N. Tamura, R. Spolenak, W.A. Caldwell, A. MacDowell, R.S. Celestre, H.A. Padmore, J.C. Bravman, B.W. Batterman, W.D. Nix, and J.R. Patel, *J. Appl. Phys.* 94, 3757 (2003).
13. A.S. Budiman, N. Tamura, B.C. Valek, K. Gadre, J. Maiz, R. Spolenak, W.D. Nix, and J.R. Patel, *Appl. Phys. Lett.* 88, 233515 (2006).
14. K. Chen, N. Tamura, B.C. Valek, and K.N. Tu, *J. Appl. Phys.* 104, 013513 (2008).
15. N. Tamura, A.A. MacDowell, R. Spolenak, B.C. Valek, J.C. Bravman, W.L. Brown, R.S. Celestre, H.A. Padmore, B.W. Batterman, and J.R. Patel, *J. Synch. Rad.* 10, 137 (2003).
16. G. Feng, A.S. Budiman, N. Tamura, J.R. Patel, and W.D. Nix, *J. Appl. Phys.* 104, 043501 (2008).
17. A.S. Budiman, S.M. Han, J.R. Greer, N. Tamura, J.R. Patel, and W.D. Nix, *Acta Mater.* 56, 602 (2008).
18. G. Lee, J.-Y. Kim, A.S. Budiman, N. Tamura, M. Kunz, K. Chen, M.J. Burek, J.R. Greer, and T.Y. Tsui, *Acta Mater.* 58, 1361 (2010).
19. C.S. Hau-Riege, S.P. Hau-Riege, and A. Marathe, *J. Appl. Phys.* 96, 5792 (2004).
20. H. Wang, C. Bruynseraede, and K. Maex, *Appl. Phys. Lett.* 84, 517 (2004).
21. W. Wu, S.H. Kang, J.S. Yuan, and A.S. Oates, *Solid State Electron.* 45, 59 (2001).
22. C.S. Hau-Riege, Internal AMD Study, private communication, 2006.
23. C.W. Chang, C.L. Gan, C.V. Thompson, K.L. Pey, W.K. Choi, and M.H. Chua, *Proc. 10th Intl. Symp. Phys. Failure Analysis of Integrated Circuits (IPFA)* (2003), pp. 69–74.
24. A. Labun and K. Jagjitkumar, *IEEE Trans. Comput. Aided Des. Integrated Circ. Syst.* 27, 1840 (2008).
25. S.P. Gurrum, Y.K. Joshi, W.P. King, K. Ramakrishna, and M. Gall, *J. Electron. Packag.* 130, 031001 (2008).