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Wafer-Level Pulsed-DC Electromigration Response at Very High Frequencies

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Abstract - DC and pulsed-DC electromigration tests were performed at the wafer level using standard and self-stressing test structures. DC characterization tests over a very large temperature range (180 to 560 °C) were consistent with an interface diffusion mechanism in parallel with lattice diffusion. That data allowed for extraction of the respective activation energies and the diffusion coefficient of the rapid mechanism. The ability to extract simultaneously a defect-based diffusion coefficient and activation energy is significant given the extreme difficulty in making those measurements in aluminum. The pulsed-DC experiments were conducted over a range that includes the highest frequency to date, from DC to 500 MHz. Measurements were also made as a function of duty factor from 15% to 100% at selected frequencies. The data shows that the pulsed-DC lifetime is consistent with the average current density model at high (> 10 MHz) frequencies and showed no additional effects at the highest frequency tested (500 MHz). At low frequencies, we attribute the lessened enhancement to thermal effects rather than vacancy relaxation effects. Finally, the deviation in lifetime from the expected current density dependence, characterized over 1½ orders of magnitude in current density, is explained in terms of a shift in the boundary condition for electromigration as the current density is decreased.

density, n is the current density exponent normally assumed to be 2, r is the duty factor, m is a constant, E_a is the activation energy, k is Boltzmann's constant and T is the temperature. If electromigration depends on frequency we would expect A to be frequency-dependent, as that term contains factors relating to the void nucleation mechanism (e.g., see Lloyd [2]). If the electromigration duty factor dependence is simply explained by damage occurring only during the "on" portion of the current pulse with no annealing of damage during the "off" portion, then $m=1$ in Equation 1; this is sometimes called the "on-time" model. If annealing (or back diffusion) does occur, then m will be greater than 1, and the pulse response is said to be enhanced. If $m = n$, then Equation 1 yields the average current density model where the time average of the current governs the duty factor response.

Electromigration frequency response measurements reported in the literature have yielded consistent results. The measurements to date have been limited to moderate frequencies (<10 MHz) with the exception of Kwok et al. [3], who extended the range to 200 MHz. Unfortunately, their lowest frequency was 50 MHz, which makes it difficult to compare that work to other studies. In general, the lifetime is constant with respect to frequency, usually a factor of 2 above DC lifetime for 50% duty factor, until a transition frequency (f_0) is reached, where it increases to a higher value [4-7]. It is generally assumed that the increase in lifetime occurs when the "on" pulse width during one period falls below the characteristic time constant for the vacancy concentration to reach equilibrium [6]. That increase usually occurs in the range of 1 Hz to a few kHz. One notable exception to constant lifetime above the transition frequency was observed by Suehle and Schafft [5], who found a decrease in lifetime around 1 MHz with a subsequent increase for higher frequencies. That response seemed to be linked to current density and disappeared as the current density decreased. Unfortunately, their frequency range was limited to 10 MHz so the behavior beyond that point could not be ascertained.

INTRODUCTION

The reliability of integrated circuits under pulsed-DC and AC conditions is of great interest to manufacturers and users of those products. This is because most IC structures are subjected to time-varying signals under operating conditions, while reliability characterization is usually done under DC conditions. Of added interest is potential enhancement in the electromigration lifetime at low duty factors during pulsed operation. Enhancement is of practical interest because it allows for less stringent current density design rules for metal lines operating in that mode. The interest in these factors is borne out by the considerable body of experimental and theoretical work on electromigration response under pulsed-DC and AC conditions, discussed briefly below.

The effects of frequency and duty factor on pulsed-DC electromigration are conveniently discussed using a modified form of Black's Equation [1]

$$t_f = \frac{A}{j^n r^m} e^{\frac{E_a}{kT}} \quad (1)$$

where t_f is the median time-to-failure of the metallization, A is a material and geometry-dependent constant, j is the current

The data on duty factor response has shown far more variation. The early work of Miller [8] generated considerable excitement, as he found an exponential dependence of lifetime with duty factor, with very large enhancements at low duty factors. Experimental work since then [3,5,7,9-14] has not shown such large enhancements, with the value of m usually around 2. Larger values of m (from 3 to 5) have also been reported [3,9,13]. When m deviates from 2, it appears to be strongly correlated to the current density exponent, n . Based on those results, it is generally assumed that the electromigration

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Figure 3. A time profile of resistance change during a typical DC electromigration test shows little sensitivity to the failure criterion chosen.

electromigration times-to-failure. Finally, wafer-level testing allows for access to a much broader range in temperature and current density than is typically available in packaged-structure tests.

Test structures from several wafers from two lots were used in the experiments. Because of possible lot-to-lot and wafer-to-wafer variations, the temperature coefficients of resistance (α) and thermal resistance (θ_{sc}) were characterized for each structure type and wafer. The temperature coefficient of resistance is key to the thermometry techniques used, and θ_{sc} is needed to calculate stripe temperature under pulsed conditions.

Resistance and cross-section measurements showed a 10% variation in line cross-sectional area between the two lots. Because resistance variations were due to cross-section variations and not material variations, we used the measured resistance to calculate stripe area and hence the applied current density for each stripe. We made no corrections for current flow through the barrier layer since its sheet resistance was estimated as 500 times the sheet resistance of the Al-Cu film.

DC Electromigration

We performed electromigration tests on the DC structures over very wide temperature (180 to 560 °C) and current density (6×10^5 to 2×10^7 A/cm²) ranges in order to bracket the self-stressing test conditions and to get as close to use conditions as was practical. The average sample size was 20 though smaller sample sizes were used at the lower temperatures and current densities because of the large times-to-failure. The bulk of the experiments were "isothermal" electromigration tests at wafer-level, with Joule heating and hence, very high current densities, used to attain the test temperature. The word isothermal is placed in quotes because the temperature is controlled by using the stripe as a thermometer. Since the basal resistance of the stripe will increase as local damage accumulates, hot spots will occur and the temperature of the stripe is in question beyond that point. Also, as damage accumulates, the control algorithm will reduce the applied current to keep the measured resistance constant. Thus, the test would be more aptly named iso-resistive rather

Figure 4. Typical time profiles of resistance changes for self-stressing tests show marked similarity to the "isothermal" profile in Figure 3. A duty 50% duty factor was used.

than isothermal. While these characteristics would seem to increase test uncertainty with regard to more benign packaged-structure electromigration tests, we have seen excellent correlation to packaged-structure tests (e.g. [20]) for several metallization systems. Further details of the test approach and algorithms used can be found in [21].

In order to investigate the current density effects, DC tests were also done at 420 °C with the polysilicon heater used to attain part or all of the test temperature. These tests were done either in the "isothermal" mode or the self-stressing constant current mode, depending on the amount of Joule heating desired.

A typical time profile of the change in room temperature stripe resistance (ΔR_0) for a wafer-level test at 320 °C is shown in Figure 3. The room temperature resistance is extracted *in situ* using a thermally derived algorithm [21]. Two important features should be noted from the figure. First, the resistance changes little over most of the test. Thus, in that regime, the test is essentially an isothermal, constant-current test. Second, failure occurs abruptly as an open after a resistance change of little more than 2%. That behavior leads to estimates of the time-to-failure that are insensitive to the failure criterion, which was chosen as a 5% increase in resistance. That is, the same time-to-failure would have been obtained if the failure criterion was 10%, 20% or an open. All of the tests showed this behavior with resultant opens with the exception of those done at the two lowest temperatures, 180 and 200 °C. For some of the stripes under those conditions the 5% increase failure criterion was reached before the stripe opened, though onset of failure was still rapid.

Self-Stressing Tests

The self-stressing tests were done at very high temperatures with the polysilicon heater set to 400 °C. That base temperature, coupled with Joule heating, yielded average stress temperatures ranging from 405 to 420 °C. We chose such extreme temperatures because they yielded times-to-failure short enough

Figure 5. Cumulative time-to-failure distributions for the isothermal tests at very high current densities. The data is well behaved with unimodal failure distributions.

Figure 6. Arrhenius plot of DC electromigration data showing two diffusion mechanisms. Solid circles are DC tests using standard test structures. Open circles are DC test results using the self-stressing test structures with the on-board buffer.

to make the experiment feasible with reasonable sample sizes (10) and a wide range of frequency and duty factor conditions.

There are several key differences in the self-stressing tests as compared to the DC wafer-level tests. The self-stressing tests relied on the polysilicon heater to attain the test temperature with modest amounts of Joule heating (5 to 20 °C). The polysilicon heater was actively controlled to ± 1 °C through use of the monitor line as a thermometer. Because of the pass gate used to provide parasitic isolation of the high side of the structure from the bond pad, *in situ* resistance monitoring was not possible. Instead, at logarithmic intervals, we removed the stress and temperature from the DUT and measured the change in resistance at room temperature. The failure time was calculated as the logarithmic average of the pre and post failure times. We were concerned that the thermal cycling would damage the DUT [22]. However, several tests conducted with no applied stress current showed no degradation over the test times used.

The self-stressing tests were done at constant applied current, whereas the wafer-level DC test algorithm controlled the current to maintain constant stripe resistance. That should have minimal impact on the time-to-failure as long as the behavior shown in Figure 3 holds; that is, the stripe resistance does not change over the bulk of the test. That behavior was also observed for the self-stressing tests, as shown by the resistance change time profile at several pulse frequencies, shown in Figure 4. Like Figure 3, the resistance remains constant until the point of failure, where a rapid increase occurs. Most of the failures resulted in opens.

While the Joule heating was modest in the self-stressing test structures, it is still necessary to correct the data for that increase in temperature, particularly at pulse widths in the range of the thermal response time. The algorithm we used to make those corrections is summarized in Appendix A.

RESULTS

DC Electromigration Tests

The results of the DC electromigration characterization tests where Joule heating alone was used to attain the test temperature are shown as lognormal probability plots in Figure 5. The lines through the data are lognormal least-squares fits. The metal is very well-behaved with unimodal distributions and no evidence of a defect population. The times-to-failure range from 10 to 450,000 seconds.

The data is replotted in Figure 6 in Arrhenius format in order to estimate the activation energy for diffusion in the metal. Note that in the plot we assume that the current density exponent, n , in Equation 1 is 2, which is confirmed by measurements at lower current densities. The 90% confidence bounds on the median times-to-failure are included for each data point though they cannot be seen for most points as they are smaller than the symbols used to denote the data. The data exhibits curvature which indicates that more than one diffusion mechanism is active over the range, presumably lattice diffusion and another, faster, diffusion mechanism. Because of the curvature we fit the data to a simple dual diffusion mechanism model described in Appendix B. The solid line in Figure 6 is the fit while the dashed lines denote the contribution from each diffusion mechanism. The fit was excellent with a reduced χ^2 of 0.094 with 14 degrees of freedom. The values of activation energies obtained were 0.79 eV and 1.41 eV. The larger value is in the range expected for lattice diffusion while the smaller value is somewhat high for grain boundary diffusion. Because of the bamboo structure of the metal, we suspect that a diffusion mechanism other than grain boundary diffusion is responsible for low-temperature diffusion in the metal tested. That mechanism, and the extraction of its diffusion coefficient, is discussed in Appendix B.

Figure 7. Current density dependence of median times-to-failure at 420 °C. Vertical lines are 90% confidence bounds.

The curvature in the solid line shown in Figure 6 implies a temperature-dependent activation energy. Because the pulsed-DC electromigration tests were done at temperatures (400-420 °C) where the electromigration response deviates slightly from low temperature behavior, an effective activation energy, defined in Appendix B, was used to compensate pulsed-DC data for small variations in stripe temperature.

Also shown in Figure 6 are the results of testing the self-stressing structure under DC conditions. The current density used was roughly a factor of 6 less than that used for the "isothermal" test at that point. The agreement with the DC structure test results validates the use of 2 for the current density exponent in the analysis shown in Figure 6.

Suehle and Schafft [6] have pointed out that very highly accelerated wafer-level tests, with commensurate short failure times, may overestimate the time-to-failure. Their arguments are based on vacancy response time considerations where, if the failure time is less than the vacancy response time, then the excess vacancy concentration will be less than the equilibrium value for the stress conditions used. That in turn should lead to an increased time-to-failure at high stress conditions, with a break in the median time-to-failure versus current density characteristic curve. Because our experiments were done under highly accelerated conditions, such an effect could contaminate the results given the wide range in times-to-failure expected.

We characterized the electromigration response as a function of current density to address the concerns outlined above. The experiments were done at current densities ranging from 5×10^5 to 2×10^7 A/cm². The results of that characterization are shown in Figure 7 as a plot of median time-to-failure versus current density. The solid lines in the figure have slopes of -2. The data clearly shows two regimes with the break between the regimes occurring around 5×10^6 A/cm². This is similar to the behavior observed by Suehle and Schafft, who observed the break around 2×10^6 A/cm².

The remarkable thing about the data shown in Figure 7 is that the current density at the break is close to that observed by Suehle and Schafft [6] yet occurs at a much higher temperature

Figure 8. Correction factor for stripes with peak temperatures that do not produce diffusion barriers at the end of the line. Vertical lines are 90% confidence bounds.

(420 versus 277 °C). If the vacancy response time exhibits an activation energy similar to that of diffusion, then the response time we observed would be on the order of 3×10^6 seconds (830 hours) at 180 °C. That is far outside the range of .01 seconds to 3 hours at 180 °C cited by Suehle and Schafft based upon their survey of the literature. Such a large value defies physical intuition and we doubt that vacancy response processes are responsible for the behavior shown in Figure 7 and that observed by Suehle and Schafft.

A more straightforward explanation is obtained by noting that the temperature rise above ambient increases as the current density increases. At high stress currents, the temperature drop at the widened ends of the test structure is large and kills the diffusion coefficient in that region. Thus, there is a diffusion barrier at the ends of the line with a zero-flux boundary condition. As the current density is reduced, the temperature gradient-induced barrier becomes less effective until, in the limit of low current densities, it ceases to exist altogether. Thus, the break in Figure 7 simply reflects electromigration under greatly different boundary conditions. Note that if the test structures had deliberate diffusion barriers placed at the ends of the lines then we would expect no break in the response and the data would fall along line a).

Given the model outlined above, we replotted the data in Figure 7 as a function of the increase in stripe temperature above background. The result is shown in Figure 8 where the vertical axis is the line labeled a) in Figure 7 divided by the experimental results. That is, the vertical axis is the "correction factor" needed to bring all of the data shown in Figure 7 into agreement with the high current density trend. We could have normalized Figure 8 to low current densities. However, since the bulk of the DC data was taken with large amounts of Joule heating, normalization to the high current density limit provides for less extrapolation of the data.

The horizontal dashed lines in Figure 8 are the high and low current density trend lines shown in Figure 7. The slanted line is a simple power law fit to the transition region. The decision to use a power law is purely empirical and reflects no underlying

Figure 9. Frequency response of the self-stressing test structures at 50% duty factor. Vertical lines are 90% confidence bounds. The increase with frequency is due to thermal effects, rather than characteristic vacancy relaxation times.

physics. The solid line in the figure was used to correct the pulsed-DC data for the effect of changing boundary conditions.

Pulsed-DC Tests

The median time-to-failure for the self-stressing structures is plotted in Figure 9 versus frequency at 50% duty factor. The frequencies range from 1 Hz to 500 MHz. The 90% confidence bounds on the medians are shown as vertical lines. Also shown are two horizontal lines at twice and four times the DC time-to-failure. Those lines represent the predictions of the "on-time" and average current density models for 50% duty factor and a current density exponent of 2. The data is tightly clustered about the "on-time" model line for frequencies less than 1 MHz and rises to the average current density model at high frequencies. That dependence is due to the difference in stripe temperature when the pulse is on or off. That is, when there is significant Joule heating the temperature of the stripe in the off-portion of the pulse is so low as compared to the on-portion that back diffusion is not effective in annealing damage. The effect does not occur at high frequencies because the temperature does not change significantly between the on and off portion of the waveform, as discussed in Appendix A. This behavior shows that test techniques that allow significant Joule heating (5 °C or so) should not be used for pulsed-DC electromigration tests with periods below the structure's thermal time constant. The agreement with the average current density model at high frequencies shows no unusual effects up to 500 MHz.

Note that the correction factor shown in Figure 8 played an important role in the data analysis because at high frequencies we were operating in the transition region between the boundary conditions. Without that consideration, the data would show no frequency dependence and would fall along the "on-time" model line. Again, care should be taken when testing structures with no chemical diffusion barriers under conditions with significant Joule heating.

The median time-to-failure for the self-stressing test structures is plotted in Figure 10 versus duty factor for frequencies of 10 kHz and 200 MHz. Again, the impact of Joule

Figure 10. Duty factor response of the median time to failure. The data agrees well with the average current density model at 200 MHz. The vertical lines are 90% confidence bounds.

heating shows up in the low frequency data. While the data does show some enhancement at low duty factors, it is below that expected from the average current density model. The enhancement does show that, with a temperature differential of 20 °C between on and off states, some back diffusion occurs. In the large Joule heating limit, we would expect the 10 kHz data to follow the "on-time" model.

The data at 200 MHz follows the average current model remarkably well. Thus, experiments with large amounts of Joule heating appear to be appropriate when operating at pulse periods below the thermal time constant. Again note that the correction factors given in Figure 8 played a key role in yielding that agreement. In this instance high current densities are advantageous as they avoid the problem of changing boundary conditions until very low duty factors are reached.

CONCLUSIONS

We found that the high frequency (up to 500 MHz) pulsed-DC electromigration response of the metal system tested followed the average current density model. At lower frequencies, thermal effects masked the response and yielded smaller amounts of enhancement than expected. Those effects are not important under use conditions unless design rules allow for significant Joule heating.

We also observed the break in the electromigration lifetime dependence on current density observed in [6]. We attribute that break to a change in boundary conditions due to the decrease in the temperature gradient at the ends of the test structure with decreasing current density. That effect would not show up in test structures that incorporate deliberate diffusion barriers at the transition from narrow to wide line width. Knowing the quantitative dependence of lifetime on current density due to that effect proved crucial in proper analysis of the data and highlights the need for vigilance when performing electromigration tests where even modest Joule heating is a factor.

Perhaps the most interesting aspect of the experiments was not the high frequency work but the DC stress test with the analysis described in Appendix B. The electromigration response agreed very well with simple mass transport theory and

Figure 11. Temperature rise as a function of time for a current step. Note that the temperature comes close to steady state at a few μs while adiabatic behavior is apparent at times below a few hundred ns.

allowed us to extract the activation energies for "interface" and lattice diffusion, estimate the "interface" diffusion coefficient and determine the relative impact of the competing diffusion mechanisms as a function of temperature. As electromigration activation energies continue to rise due to process improvements, the relative influence of competing diffusion mechanisms will become more important in packaged-part testing. Wafer-level testing, if done properly, could prove a useful tool in delineating those diffusion mechanisms.

Finally, the self-stressing test structures proved capable of the task of yielding accurate, high quality electromigration data at a wide range of pulse frequencies. As the structures can be used in both the wafer-level (high stress) and packaged-structure (moderate stress) environments, self-stressing test structures should provide a viable alternative to the costly and unwieldy high frequency test systems used for electromigration tests under AC and pulsed conditions.

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APPENDIX A - THERMAL ANALYSIS

The transient thermal profile for the structures tested was calculated by using the solution to the one-dimensional time dependent heat equation given in [23] and dimensional parameters obtained from the metallization cross-sections. An example of those calculations for a step current input is shown in Figure 11. The vertical lines are reference points to relate the times to frequencies at 50% duty factor. As expected, adiabatic conditions hold at pulse widths below a few hundred ns, where

the temperature will vary slightly about the mean. We expect the pulsed-DC electromigration response to follow that obtained under low stress conditions in the short pulse (high frequency regime). At pulse widths larger than 10 ms, the time to steady state is short compared to the pulse width and the temperature profile closely follows the current pulse shape. In that instance, the off portion of the pulse is at ambient and allows for little diffusion as compared to the on portion of the pulse when Joule heating is significant.

All data was corrected for temperature variations and related to a common temperature T_c through the relation:

$$t_{fc} = \frac{t_{fm}}{t_p} e^{\frac{E_a}{kT_c}} \int_0^{t_p} K(T(t) - T_{heat}) e^{-\frac{E_a}{kT(t)}} dt$$

where,

$$T(t) = T_{heat} + \overline{\Delta T} - \frac{\Delta T_{avg}}{2} + \Delta T(t)$$

and t_{fc} is the corrected time-to-failure, t_{fm} is the measured time-to-failure, t_p is the pulse width, E_a is the effective activation energy (see Appendix B), K is the boundary condition correction factor given in Figure 8, T_{heat} is the temperature of the polysilicon heater, ΔT_{avg} is average of the transient profile, D $T(t)$ is the transient temperature profile shown in Figure 11 and the average temperature increase of the line is [18]

$$\overline{\Delta T} = \frac{r \theta_{sc} I^2 R_0 [1 + \alpha(T_{heat} - T_c)]}{1 - r \alpha \theta_{sc} I^2 R_0}$$

where r is the duty factor, θ_{sc} is the stripe to chuck thermal resistance (measured), I is the applied current, T_c is the chuck temperature, R_0 is the DUT resistance at T_c and α is the temperature coefficient of resistance at T_c . The data was normalized to the test temperature at DC conditions, 412 °C, so that frequency and duty factor effects can be directly compared.

APPENDIX B - DIFFUSION CALCULATIONS

The curvature exhibited by Figure 6 clearly indicates that multiple diffusion mechanisms are active over the temperature range used in the DC electromigration tests. We modeled that curvature as the result of two parallel diffusion mechanisms. By parallel we mean that all materials are in equilibrium perpendicular to the direction of electromigration-induced transport. That is, there are no concentration gradients transverse to the stripe. That model has the satisfying outcome of yielding lattice transport at high temperatures and a faster transport mechanism at low temperatures. We assume here that interface diffusion is responsible for low temperature diffusion in this metal system. That assumption is discussed in more detail below.

Figure 12. The relative contribution to metal failure due to the two diffusion mechanisms. Note that parameters extracted from the fit in Figure 6 were used to generate the curves.

Based upon the above considerations, the solution to the electromigration equation can be written, by inspection from solutions such as those given by Shatzkes and Lloyd [24], as

$$\frac{1}{t_f j^2} = \frac{1}{A_i} e^{-\frac{E_{ai}}{kT}} + \frac{1}{A_l} e^{-\frac{E_{al}}{kT}} \quad (2)$$

where the subscripts denote interface and lattice diffusion respectively. The solid line in Figure 6 is the fit to Equation 2 while the dashed lines denote the contribution from each term in Equation 2. As discussed previously, the fit is excellent.

The form of Equation 2 is similar to that proposed by Dreyer et al. [24]. However, our interpretation is quite different. Equation 2 implies that there is a fast diffusion mechanism always present, regardless of the grain structure of the metal. The model in [24] assumes that grain boundary and lattice diffusion mechanisms appear in series, with the response dictated by the probability density functions of each type of grain configuration. Because of the goodness of fit in Figure 6 and the observed near-bamboo grain structure of the metal, we feel that there is another diffusion mechanism other than grain boundary diffusion operative in parallel with lattice diffusion. That is, if Dreyer's model held for this metal, we would expect to see only lattice diffusion, with no curvature in Figure 6 and a higher activation energy at low temperatures. For lack of a better term, we will call this mechanism "interface diffusion." From a physical standpoint, we suspect that this diffusion mechanism will be active along the Al - TiW interface. Given the agreement between Equation 2 and the data shown in Figure 6, our assumptions appear warranted.

The result of the fit shown in Figure 6 is presented in somewhat different form in Figure 12. There the fraction of damage caused by each transport mechanism is plotted versus temperature. At temperatures where packaged structure tests are

Figure 13. Effective activation energy for the competing diffusion mechanisms versus temperature. The dashed lines denote the temperature range over which the pulsed-DC tests were done.

conducted, the "interface" diffusion mechanism dominates and results can be directly extrapolated to use conditions. At temperatures where conservative wafer-level tests are done, 250 to 350 °C, lattice diffusion will start to affect the results. At temperatures where very fast wafer-level tests such as SWEAT are done (500 °C) lattice diffusion accounts for fully half of the transport and interpretation of results are difficult without a detailed characterization as shown in Figure 6. Characterizations such as these will become more important as grain boundary diffusion is suppressed in advanced metals and hence the impact of multiple diffusion mechanisms is felt at lower temperatures.

For the purposes of data analysis, especially at high temperatures where lattice diffusion plays a role, it is useful to define an effective activation energy [24] based on the tangent to the fitted curve shown in Figure 6. That result is shown in Figure 13 versus temperature. We used the effective activation energy to correct for small (< 20 °C) temperature variations due to Joule heating in the self-stressing tests.

The pre-exponential factors from the fit shown in Figure 6 can be used to estimate the "interface" diffusion coefficient by equating the coefficients to solutions of the electromigration equation [25]

$$A_l = \frac{C_f}{D_{0l}} \left(\frac{kT}{ez_l^* \rho} \right)^2 \quad (3)$$

and

$$A_i = \frac{C_f}{\left[\frac{\delta D_{0i}}{d} \right]} \left(\frac{kT}{ez_{gb}^* \rho} \right)^2 \quad (4)$$

where C_f is the critical vacancy concentration, δ is the short-circuit diffusion path thickness, d is the short-circuit diffusion

path width, z^* is the effective charge and all other quantities have been defined. Dividing Equation 4 by Equation 3 yields

$$\delta D_{0i} = \frac{A_i}{A_i} \left(\frac{z_i^*}{z_{gb}^*} \right)^2 d D_{0i} \quad (5)$$

From the fit in Figure 6, $A_i = 1.6 \times 10^{11} \text{ s-A}^2/\text{cm}^4$ and $A_l = 9.3 \times 10^6 \text{ s-A}^2/\text{cm}^4$. Because of the near-bamboo structure of the lines, we set the defect width, d , to the linewidth, $1.5 \mu\text{m}$. From the literature, $z_{gb}^* \approx 4$ [26], $z_l^* \approx 26$ [26] and $D_{0l} = 1.71 \text{ cm}^2/\text{s}$ [27]. We chose to use the value for the effective charge associated with grain boundary diffusion for lack of a better number. The value chosen for D_{0i} was associated with an activation energy that was close to the value we measured. Those values used in Equation 5 yield $\delta D_{0i} = 7.5 \times 10^{-7} \text{ cm}^3/\text{s}$. The ability to extract simultaneously a defect-based diffusion coefficient and activation energy is significant given the extreme difficulty in making those measurements in aluminum. Generally, the diffusion coefficient is extracted from moving marker experiments while the activation energy is extracted separately from electromigration time-to-failure measurements.

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