Scanning thermal microscope study of a metal film under current stressing: role of temperature inhomogeneity in the damage process

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Abstract
In this paper direct observation of the evolution of local temperature inhomogeneity and the resulting atomic migration in a metal film (Ag on Si) stressed by a high electrical current is reported. Experiments were carried out by simultaneous temperature mapping and topography imaging using a scanning thermal microscope in combination with resistance measurements. The experimental observation is analysed using a model based simulation. The experimental observation and the simulation show that due to current stressing the temperature of the film becomes significantly inhomogeneous over time (with the local temperature deviating strongly from the mean). This creates local stress as well as local temperature gradients (as high as \( \approx 3 \, \mu \text{m}^{-1} \)) that lead to mass migration in addition to electromigration. The local temperature inhomogeneity serves as one of the main agents for local atomic migration which leads to a change in the film microstructure. The migration leads to damage and eventual failure as simultaneously monitored by \textit{in situ} resistance measurements.

(Some figures in this article are in colour only in the electronic version)

1. Introduction

In the presence of a large electric current through a metal film, ions in the film are acted upon by a number of forces (in addition to the direct electrostatic force) that lead to migration of ions. These forces can be due to momentum transfer from electrons to ions leading to electromigration (EM) or due to stress (stress migration) or stress resulting from thermal gradients [1–6]. Whichever may be the cause of this mass migration, it leads to changes in the film’s microstructure and the formation of voids or hillocks leading to complete damage when the current density is very high, typically of the order of \( 10^6 \, \text{A cm}^{-2} \). In such damage formation, temperature plays an important role. In some earlier works, dependences of damage formation on the equilibrium temperature of the film and the change in the average film temperature (that is the temperature measured on a macroscopic scale) caused by the Joule heating effect have been investigated [7, 8]. However, owing to the initial microstructural variation in the film and its evolution with time during damage formation, it is very unlikely that the temperature remains constant and uniform over the film during mass migration. The local temperature gradient developed in this situation could influence the damage mechanism in many ways. For instance, atomic migration in the film can be expected to take place in a very different fashion than it would do with uniform temperature, because of the thermally activated nature of atomic migration. Moreover, spatial inhomogeneity of the temperature can strongly influence the atomic migration due to two other facts, namely, it can alter the distribution of stress in the film and the local temperature gradient (\( \nabla T \)) acts as an additional thermodynamical force for atomic migration. An investigation of simultaneous evolution of microstructure and spatial variation in local temperature in the microscopic regime over the film during the progress of
the damage process would help to understand the correlation among them and to justify such a speculation. To our knowledge, such a study has not been attempted before. In this investigation we would specifically like to investigate the role of temperature inhomogeneity as it develops during the course of current stressing and its relation to the damage process.

This paper reports our attempt to address this issue directly with an experimental investigation of the evolution of microstructural changes with the temperature distribution in the metal film during the development of the damage process caused by high current stressing. Also, since atomic migration in a current stressed film is a complicated process, we carried out a simulation to compare our experimental result with theory. We did the experiment by simultaneous topographic imaging and spatially resolved temperature mapping using a scanning thermal microscope (SThM) along with the in situ resistance measurement with the progress in the damage process. In the past, a few imaging techniques such as infrared microscopy [9], SThM [10], laser-reflection thermometry [11] and scanning Joule expansion microscopy [12] have been used to image the thermal inhomogeneity of metal interconnects with different degrees of lateral spatial resolution. SThM has been demonstrated as a powerful thermal imaging tool to map the temperature distribution due to Joule heating in metal lines of 5 μm length, 0.030 μm width and 2.5 μm thickness [10], though it has not been used for the investigation of the evolution of the damage process in conjunction with thermal mapping. Our simulation was based on a finite element model (FEM) where all possible forces influencing atomic migration were considered. Since the initial microstructure has a great impact on the damage process, in this simulation we used experimentally obtained information about the sample’s microstructure as an input. This method has not been used in any earlier studies. Our investigation shows clear evidence of correlation between the evolution of the damage process and the spatial variation in temperature besides the temperature averaged over a large area or the average temperature.

### 2. Experimental techniques

Experiments were carried out in Ag films (thickness ≈ 0.160 μm) as test samples (dimension ≈ 20 μm square). Although metals such as aluminum or copper are normally used as interconnect, we used Ag for our work to avoid undesired surface oxidation in air ambience. The films used in this work were prepared by thermal evaporation on Si (held at 175 °C) at a base pressure of 10⁻⁹ mbar and post-growth annealing at 250 °C for 6 h in the same vacuum, because it was observed that the films prepared under these conditions had the lowest room temperature resistivity (ρ₃₀₀K) in comparison with the films annealed at lower temperatures, while those annealed at higher temperatures were rougher with poor grain connectivity. The x-ray diffraction (XRD) studies of the films showed that the films were mostly oriented in the (1 1 1) direction and well textured. The initial uniformity of the film’s surface was studied with atomic force microscopy (AFM) and the root mean square (rms) value of the surface roughness was found to be 4.8 nm. The room temperature resistivity and residual resistivity ratio (ρ₃₀₀K/ρ₄.₂K) for the films were typically around 1.65 μΩ cm and ~8, respectively.

Figure 1 shows the schematic for in situ resistance and SThM as well as topography measurements. The SThM and topography measurement was based on a contact mode AFM technique with an additional temperature measurement unit (SThM module) and using the cantilever of AFM itself as a thermometer. We used a commercial AFM from the Veeco Metrology Group (Model: Autoprobe CP Research) and a cantilever with a built-in resistive thermal element made of Pt/10% Rh alloy of diameter 5 μm which acts as a probe for the scanning thermal microscope. A mirror, cemented on the cantilever, acted as the reflector for the laser light to measure the deflection of the cantilever. To measure the temperature of the sample at the point where the tip of the probe touches it, the probe was connected as an arm of a balanced Wheatstone bridge powered by a constant current from the SThM module. The error voltage, Vₑ (see figure 1), from the bridge varies linearly with the local temperature (T) of the point where the tip touches the film and is given as

\[ Vₑ = V₀ + iα R₀ C(T − T₀), \]

where \( V₀ \) is the bridge voltage at the reference temperature, \( i \) the current applied to the probe, \( α \) the temperature coefficient of resistance of the probe material, \( R₀ \) the resistance of the probe at a reference temperature \( T₀ \) and \( C \) the calibration gain of the SThM module. Since the leads of the cantilever beyond the sensor head were made of Ag wire of a much larger diameter (75 μm), they had very low resistances in comparison with the sensor head and hence the measurements were insensitive to the temperature dependence of the thermometer leads. Before carrying out measurements on the actual sample, the SThM probe was calibrated against \( Vₑ \). This was done by using a Pt100 film thermometer as a sample which was placed on a heater and recording \( Vₑ \) for different reference temperatures as measured with the Pt100 film thermometer serving as the sample. From a fit through these calibration points the absolute temperature could be calculated according to the equation

\[ T = \left( T₀ - \frac{1}{m} V₀ \right) + \frac{1}{m} Vₑ \]

\[ (2) \]
for any value of $V_{out}$. From the calibration curve we find the slope $m = 8.3$ mV K$^{-1}$. The spatial resolution of the thermal (local $T$) images was limited by the finite size of the tip to $\approx 200$ nm. The test sample size was kept in the range of $\approx 20\,\mu$m due to this spatial resolution. From the temperature profile in the thermal map of the sample it was possible to estimate the thermal gradient, the minimum value of which was limited by the noise of the measurement, and it was $0.1\,K\,\mu$m$^{-1}$. To facilitate the resistance measurement, this film was connected by a Ag film which was deposited in a four-lead pattern. We applied a constant direct current (dc) of $0.65\,A$ across the current leads to stress the film while the voltage drop across two other leads was recorded to measure the resistance. Thus the current in the test sample was in the plane of the film with the calculated initial current density, $J_{dc} = 2 \times 10^7\,A\,cm^{-2}$. With this current the initial power dissipation at the film due to Joule heating was $\approx 0.05\,W$. In the experiment care was taken so that no other component of the current combines with this stressing current. To ensure that damage does not take place in the current leads, the current density at the leads was kept at a value as small as $\approx 4 \times 10^4\,A\,cm^{-2}$ by making their widths and thickness 4 mm and 0.5 $\mu$m, respectively. The damage process was monitored along with imaging by directly recording the resistance ($R$) at regular time intervals automatically until the film was damaged, as evidenced by a substantial increase in $R$. With this experimental setup we could systematically investigate the evolution of the damage process with in situ AFM and SThM.

3. Result

In figure 2 we plot the evolution of the resistance of a film as a function of time with the progress in the damage process. Resistance is plotted in a logarithmic scale to enhance the initial trend but with a much larger slope $R \approx 2 \times 10^{-3}\,h^{-1} i_0$ is the slope of the log$R$–$t$ plot. After this point, resistance followed a similar trend but with a much larger slope $p_2 \approx 3.6 \times 10^{-2}\,h^{-1}$. This trend was seen to be maintained till around 120 h (marked as $t_2$ in the plot) after which the resistance also kept increasing at a similar average rate. In the late stage of the damage process, as shown in figure 2, there are steps in resistance changes. We generally take the SThM and topography images in these regions because the resistance stays constant over a time period. The steps are marked as $t_0$, $t_1$, ... $t_7$ in the figure. In figure 3, three representative topography, SThM images and temperature profiles are shown. The data obtained at different stages of current stressing are shown in three different rows. The progress in damage can also be seen from the change in $R$ of the film with time (figure 2). Resistance data in figure 2 and the AFM based imaging were taken simultaneously. The first figure in row (a) shows the topography of the whole film, including the contact pad, while in all the other images only the part of the film marked by a square is shown in order to elaborate the changes that occur as a result of current stressing. Row (a) corresponds to the film at unstressed condition (at $t_0$). The line scan is a typical example of temperature inhomogeneity in an unstressed film and shows the noise limited resolution. As the damage process progresses one sees the appearance of relatively hot areas surrounded by relatively cold contours as indicated by arrows in the SThM image in row (b), which was taken after 146 h (at $t_3$) of current stressing. A typical line scan on the SThM image shows that the spatial temperature fluctuation has gone up. Simultaneous contact mode AFM mapping shows that these local hot spots act as regions where the films’ microstructure has changed considerably. This becomes more visible (see row (c)) in later stages of the damage process where there were sufficient changes in the microstructure and the film becomes rough (with the formation of hillocks and voids) at these positions and a more nonuniform temperature profile develops. Row (c) images were taken after 163 h (at $t_5$) of current stressing. As the damage process progresses, the temperature over the film becomes more nonuniform and a larger part of the film is damaged. (Note: the line scans, shown in figure 3 at different stages of current stressing, were taken at the same sites to show gradual evolution of temperature inhomogeneity.)

The result of in situ SThM measurement qualitatively shows that the progress in damage is accompanied by an increase in surface roughness, enhanced local temperature variation and also an enhanced average temperature. A quantitative evaluation of these observations can be obtained from the development of these parameters with the progress in the damage process. Correlations among these parameters will lead to a clearer picture. To achieve this, we calculate the rms roughness of the film ($\sigma(h)$), the average temperature ($\langle T \rangle$) and the relative variances in temperature ($\langle (\Delta T)^2 \rangle / \langle T^2 \rangle$) for an arbitrarily chosen part of the film at different stages of the damage process. The averages are taken over the whole surface of the film. We have marked such an area with a square in the topography image in row (a) of figure 3. The dimensions of this area are $12\,\mu$m $\times 12\,\mu$m. All further calculations presented here were done for this area.

To calculate the rms roughness of the film we find out the height profiles along some lines over the selected part of
the film. The standard deviation of the height profile gives the rms roughness along a line. We drew 60 such lines in the horizontal direction and 60 lines in the vertical direction. Thus each line is separated by \(\approx 200\) nm, a value close to the resolution of the SThM tip. For each stage of damage we calculated the rms roughness averaged for all the lines. This rms roughness \(\sigma(h)\) (averaged for the area) for the stages marked as \(t_0, t_1, t_2, \ldots , t_7\) in figure 2 is plotted in figure 4(a) as a function of time along with the resistances of the film at the respective stages. From this plot the increase in \(\sigma(h)\) with the time of current stressing can be clearly seen. Since the roughness of the surface has resulted from the mass flux in the film due to the migration process, the quantities \(\sigma(h)\) at any stage are a measure of the mass migration of the film that occurred at the corresponding stage due to the current stressing. Initially \(\sigma(h)\) increased very slowly, but became very rapid when the resistance change is also rapid. The simulation, described later on, shows that this happens when there is significant mass migration. In the example of the film shown here this happens after 146 h of stressing. At this stage the rms roughness is 24.5 nm, which is 5 times more than its initial value and is around 15% of initial thickness. Towards the end of the experiments the \(\sigma(h)\) increased by more than one order of magnitude.

To study the evolution of the average temperature with the progress in the damage process, we calculate the mean temperatures along the lines.

Figure 3. (Row-wise) Topography, thermal map and typical temperature line profile (a) before current stressing, (b) after 146 h and (c) after 163 h for the part of the film marked by the rectangle. All lengths are in \(\mu m\) and the temperatures are in Kelvin scale. Dashed lines indicate the mean temperatures along the lines.

Figure 4. Evolution of (a) the rms roughness \(\sigma(h)\) and (b) the mean temperature \(\langle T \rangle\) as a function of the stressing time. Change in the resistance is also plotted to show the progress in the damage process.
temperature $⟨T⟩$ for the same part of the film in the same way as was done for the calculation of $\sigma(h)$. Here we take the thermal map of this part of the film and find the temperature profile along 60 horizontal lines and 60 vertical lines. Some typical temperature line profiles are already shown in figure 3. The mean temperatures $⟨T⟩$ for various stages of damage, calculated from the mean temperature for each line and averaged over all lines, give the average temperature of this film area at corresponding stages of damage. The evolution of $⟨T⟩$ is plotted in figure 4(a) as a function of the stressing time. With the increase in the stressing time $⟨T⟩$ increases, initially rather slowly but rapidly after 120 h of stressing. Starting from an initial value 300 K, $⟨T⟩$ becomes 313 K after 120 h and increases more rapidly to a value around 335 K after 146 h.

Relative variances of temperature $((ΔT)^2)$ for any stage were obtained by calculating the variance of temperature for each line and averaging over all the lines and dividing the resultant quantity by the average of the squares of temperature along all the lines. The quantity $((ΔT)^2)$ represents the spatial inhomogeneity of temperature over the selected area irrespective of the background temperature. The change in the relative variances in temperature is shown in figure 5(a) as a function of time. Corresponding resistance values are also shown there. Relative variance of the temperature after 146 h of stressing is around 50 times more than its initial value. Almost throughout the stressing time, all these quantities were seen to increase with the increase in damage.

4. Discussion and simulation

4.1. General discussions

The result of the experiment shows that thermal inhomogeneity over the film, which is developed during current stressing, is an important physical parameter and it plays an important role in the failure process of the film. This can be explained from the following point of view. Due to the presence of various driving forces in a current stressed film, thermally activated motion of atoms takes place [13]. At the same time, the Joule heating effect causes an increase in the average temperature, $⟨T⟩$, of the film. Now, since the resistance of any part of the film depends on the microstructure of that part, the microstructural changes in the film caused by atomic migration during the damage process result in local resistance changes that subsequently lead to spatial variation of the current density. This will lead to local variation in heat dissipation which will give rise to the observed spatial variation of the local temperature in the film. With the increase in the stressing time, enhanced atomic migration causes surface depression (this causes nucleation and the developments of voids) and mass accumulation (this causes formation of hillocks) in the film, resulting in enhanced surface roughness and a significant variation in local resistances over the film. Consequently, the temperature distribution is expected to show stronger spatial variation, irrespective of the background average temperature. This is what was observed in this experiment. Comparing the evolution of $((ΔT)^2)/T^2$ with that of $σ(h)$, it could be seen that for most of the stressing time, the relative spatial variation of the temperature increased along with the increased surface roughness.

In a current stressed film, the mass migration and the local atomic flux are greatly influenced by the temperature. In particular, atomic migration resulting from local temperature gradient or thermomigration becomes one of the main components of net mass migration, while the other components are electromigration and stress migration. Interestingly, stress migration itself can be influenced by temperature inhomogeneity, because the local variation in the temperature produces a mechanical stress field in the film due to a mismatch of the thermal coefficients of expansion of the film material (Ag) and the substrate (Si). The stress field thus generated also acts as an additional strong driving force for atomic migration. The effective result of the temperature variation is enhanced atomic migration. Thus since mass migration results in the development of temperature inhomogeneity in the film which in turn enhances the mass migration itself, hence, after prolonged stressing of the film the migration process and the temperature inhomogeneity become more and more significant and accelerated. This results in a rapid increase in the parameters such as $σ(h)$ and $((ΔT)^2)/T^2$ which was observed in this experiment after around 120 h. On further stressing, the rate of mass migration increases many times and eventually there occurs a ‘runaway’ situation in all the parameters and the film gets completely damaged.

The component of mass flow arising from the thermal gradient scales with $((1/T^2)∇T) e^{-E_a/k_b T}$ where $E_a$ is the activation energy for diffusion [14]. To check the connection of the thermal gradient with the damage process, we calculated the local temperature gradient from the temperature profiles obtained from the SThM images at different stages of the damage process. We obtained the magnitude of the temperature gradient $|∇T|$ by averaging over the same
regions that we used for the calculation of rms roughness. It can be noted that with the change in temperature in the film, the thermal conductivity of the film may change, which may result in artefacts in the measured temperature distribution. However, since for most of the stressing period the change in the average temperature is rather too small to cause any significant change in the thermal conductivity of the film, the degree of such an artefact can be neglected. To calculate $|\nabla T|$ we used the same temperature profiles for all the lines used as described in the last section. We obtained $(\partial T/\partial x)$ and $(\partial T/\partial y)$ from the horizontal lines and vertical lines, respectively, for each of their cross-points and then we calculated the local value of $|\nabla T|$ at that point by taking the square root of $(\partial T/\partial x)^2 + (\partial T/\partial y)^2$. When this quantity was averaged over all cross-points, $|\nabla T|$ gave the magnitude of the temperature gradient averaged for the selected region of the film. In figure 5(b) we have plotted $|\nabla T|$ as a function of time. To show the development of surface nonuniformity with the evolution of the temperature gradient, we have also plotted $\sigma(h)$ in the same figure. It can be clearly seen from this plot that the local gradient, $|\nabla T|$, increases monotonically with time and that the rms roughness follows the same trend. Thus, the local temperature distribution and the thermal gradient are related and both have a significant effect on the migration process. Though it is not possible to single out the measure of damage caused by the temperature gradient field alone from other causes, the results of in situ SThM have clearly shown that the local temperature variation has a strong influence on the local atomic mass flux.

The argument that local resistance inhomogeneity leads to inhomogeneity in local heat dissipation and hence the local temperature variation will be justified in our simulation described below. Also the explanation that mass migration is controlled by the local temperature gradient in a significant way will be justified using the same simulation.

4.2. Simulation

From the above discussion it is clear that the damage process of a current stressed metal film is a highly complicated process where various influencing phenomena take place simultaneously and they are dependent on a number of parameters such as current density, operating temperature, thermo-mechanical properties of the film and other materials in physical contact with it. It is also important to consider that the film cannot be considered as smooth. It has topological features and thus local variation in resistance. It is thus not possible to obtain an accurate analytical result on the evolution of the damage process. However, with a good model the damage process can be simulated. In the simulation part of this investigation we considered all the important driving forces and their dependences on these parameters in a realistic model. We will see that the model, though simple, captures most of the features observed in the experiment. Below we describe the model briefly and present the results. The detailed model and computational methods are beyond the scope of this paper and have been presented separately [15].

4.2.1. Model and method of the simulation. Our simulation is based on the calculation of the local atomic flux and the local temperature distribution in the presence of a stressing current. This was done by using a combination of finite element analysis (FEA) and solution of an electrical network. For this, the film was modelled as a resistance network (see figure 6) and the film/substrate system was modelled to be composed of a number of finite elements (figure 7). To make such a model more realistic, it is necessary to start with a realistic input to the model. In this simulation we used the topography of the film, taken by AFM, in its pristine condition as the input and we derived the resistance network and the size of the elements from this input. The value of each local resistor was obtained from the height profile of the AFM image. We made an assumption that, on average, the resistivity is homogeneous across the film but local resistance variation arises from the local height variation (in other words the cross-section of the
local resistance element) which is captured by the actual AFM image. The equivalent resistance of the film was obtained by solving the network. This equivalent resistance was then compared with the experimentally measured resistance of the film and a scaling factor was obtained. The scaling factor was then used to scale the values of all the resistance elements of the network. We believe that in this way we get a more realistic value of the local resistance elements that comprise the network and that is compatible with the resistance of the actual film. To our knowledge, modelling the atomic migration process in the current stressed metal film using an experimentally obtained microstructure as the input has never been done before.

When the network is biased with a constant current source, the current through each resistance element was calculated. Each resistance element also behaved like a source of local heat input due to Joule heating. The finite element modelling of the film was done to calculate correctly the local temperature at any point of the film. The height of a finite element for the film at any location was taken according to the film’s height at that location whereas the height of all the elements for the substrate were kept fixed throughout the simulation. Using FEA the local temperatures for each node of all elements were calculated for the film and the junction between the film and the substrate. The main task of the FEA was to solve the heat flow equation:

$$\rho m \frac{\partial T}{\partial t} = \nabla \cdot (k \nabla T) + Q,$$  

(3)

where $\rho m$ and $s$ are the mass density and the specific heat of the material, respectively, $T$ is the temperature, $k$ is the thermal conductivity of the material and $Q$ consists of the rate of heat generation per unit volume (as a result of Joule heating) and radiative heat loss. The solution of this equation gives the space–time dependence of temperature. In this analysis the temperature distribution inside each element was approximated as

$$T(x, y, z, t) = \sum_{i=1}^{8} T_i^C(t) \psi_i^C(x, y, z)$$  

(4)

and the ‘Galerkin weighted residual method’ was followed [16]. For this simulation both the film and the substrate were assumed to be thermally isotropic. We used $\rho m = 1.049 \times 10^3 \text{kg m}^{-3}$, $s = 235.4 \text{J kg}^{-1} \text{K}^{-1}$ and $k = 430 \text{W m}^{-1} \text{K}^{-1}$ for the silver film and $\rho m = 2.330 \times 10^4 \text{kg m}^{-3}$, $s = 700 \text{J kg}^{-1} \text{K}^{-1}$ and $k = 150 \text{W m}^{-1} \text{K}^{-1}$ for the silicon substrate.

Once the temperature distribution in the film and the substrate was calculated, the corresponding stress due to the mismatch of the thermal expansion coefficients (TEC) of the film and the substrate was found. Using the just obtained local temperature and stress, the net mass flux arising from the combined contribution of ‘electromigration’, ‘stress migration’ and ‘thermomigration’ was calculated. For this we used the formula

$$J_a = \frac{D_a C}{k_B T} \left( Z_a e \rho J - \Omega \nabla \sigma - \frac{Q^*}{N_\Lambda T} \nabla T \right) \exp[-E_a/k_B T],$$  

(5)

where $J_a$ is the atomic flux, $D_a$ is the prefactor of diffusivity, $C$ is the atomic concentration, $k_B$ is the Boltzmann constant, $Z_a^*$ is the effective valence of the material, $\sigma$ is the resistivity of the film, $J$ is the current density, $\Omega$ is the atomic volume, $\sigma$ is the stress, $Q^*$ is the heat of transport, $N_\Lambda$ is Avogadro’s number and the rest of the symbols have the usual meaning. For the Ag film, we took values of $D_a = 6.4 \times 10^{-7} \text{m}^2 \text{s}^{-1} \ [17]$ for self surface-diffusion, $Z_a^* = 21 \ [2]$, $\rho = 1.65 \times 10^{-8} \Omega \text{m}$ (experimentally obtained), $\Omega = 1.7 \times 10^{-29} \text{m}^3$, $Q^* = 0.5 \text{eV}$ and $N_\Lambda = 6.022 \times 10^{23}$. We used $J = 2 \times 10^7 \text{A cm}^{-2}$ and kept $E_a$ as an adjustable parameter. For the sake of simplicity, we made an approximation that $E_a$ does not depend on the nature of the driving forces.

The resulting mass migration was then used to recalculate the new height profile of the film. The new local height profile becomes the input for the next simulation steps. At the end of each simulation step we can obtain the new resistance of the film, the average temperature, the local height variation (which we expressed as the rms roughness) and the local temperature variation that we expressed as the rms temperature fluctuation averaged over the film surface. These are also the experimentally determined observables. This process was followed in every simulation step. Thus, in our simulation we could see the evolution of the microstructure and local variation in temperature along with the evolution of the film resistance as a function of time (which is the number of simulation steps) during the progress in the damage process caused by current stressing. The value of $E_a$ was adjusted so that the evolution of damage, as evidenced by the increase in resistance, matched better with the experimental observations.

4.2.2. Results of the simulation. The results of the simulation (plotted as the number of simulation steps) are shown in figures 8 and 9 for resultant $E_a = 0.15 \text{eV}$. Noticeably, this value of $E_a$ is very close to 0.1 eV, the activation energy for surface diffusion of the Ag atom on the Ag (1 1 1) surface [17]. The results of simulation can be compared with the experimental results shown in figures 4 and 5. In figures 8(a) and (b) we show the evolution of rms roughness ($\sigma(h)$) and the average temperature ($\langle T \rangle$) along with the evolution of resistance as obtained from the simulation. In figure 9(a) we show the evolution of the resistance of the film along with the change in the relative variances of the temperature. The evolution of the surface roughness and that of the temperature gradient can be seen from figure 9(b). The simulation showed that the surface roughness, mean temperature, temperature gradient and the resistance increase slowly with time initially, but at a very rapid rate after some instants when the resistance starts to change at a rapid rate. This result is in conformity with what has been observed experimentally. The simulation not only reproduces the experimental observation qualitatively but also has quantitative validity. As an indicator we can compare the numbers when as a result of the migration the resistance $R$ of the film reaches 1 $\Omega$. In the experimental observation the parameter surface roughness $\sigma(h)$ reaches $\approx 0.017 \mu\text{m}$. The simulation shows a value of $\approx 0.023 \mu\text{m}$. The average temperature ($T$) at that time is $\approx 539 \text{K}$. The corresponding value from the simulation
Evolution of (a) rms roughness ($\sigma(h)$) and (b) mean temperature ($\langle T \rangle$) along with resistance as a function of the stressing time with the progress in the damage process as obtained from the simulation.

Change in (a) the resistance ($R$), the relative variances of temperature ($\langle \Delta T \rangle^2/\langle T \rangle^2$) and the local temperature gradient $|\nabla T|$ with time as the damage process progresses as obtained from the simulation.

The average temperature gradients $|\nabla T|$ at that instant obtained from the experiment and from the simulation are 0.95 K $\mu$m$^{-1}$ and 1.13 K $\mu$m$^{-1}$, respectively. The relative variance of the temperature ($\langle \Delta T \rangle^2/\langle T \rangle^2$) for that stage from the experiment is $2 \times 10^{-6}$ and from the simulation is $4 \times 10^{-6}$. We believe that the slight deviation of the simulation output of these two parameters from the experimental ones is due to the complexity in modelling the underlying local heat transport process. Nevertheless, we find that the simulation reflects the experimental results in a significant way throughout the process of current stressing. This is encouraging because there are no adjustable parameters in the simulation except the activation energy $E_a$.

The simulation establishes the basic understanding about the nature of atomic migration in a complex system such as a current stressed film. In the simulation the temperature enters in three ways: first, an enhanced average temperature accelerates the atomic migration which is a thermally activated process, second, an enhanced local gradient of $T$ also enhances the atomic flux and, third, the local temperature variation results in the redistribution of stress. Thus, the nonuniform temperature causes nonuniform mass migration in the film, resulting in an enhanced surface roughness. The effect of surface roughness is realized in the simulation by the unequal resistance and thus the current density in each element, which results in a nonuniform temperature in the film. Thus, in the simulation atomic migration and the evolution of temperature over the film are tightly correlated. This can be seen in the nature of evolution of the parameters such as $\sigma(h)$, $\langle \Delta T \rangle^2/\langle T \rangle^2$ and $|\nabla T|$ during the current stressing in the simulation. In agreement with the experimental result, the simulation justifies our understanding of the underlying physical processes involved in the damage process and also the speculation that the amount of damage and the evolution of variation in local temperature are closely correlated.

5. Conclusion

To summarize, in this investigation, we have carried out a systematic study on the time evolution of the local temperature variation as well as the local topology variation of a thin metal film subjected to high current stressing using SThM. We also measured in situ the resistance change to monitor the progress in the damage to the film under current stressing. We found that although the film has a uniform temperature initially, at some point of time there appear regions with relatively large temperature inhomogeneity with hotter zones surrounded by colder contours. The migration process nucleates in these regions and significant atomic migration results here as time progresses. As time progresses, new hotter and colder zones with a higher local temperature difference appear. Our experiment demonstrates a close correlation among the atomic migration, the variation of local temperature and microstructural variation such as surface roughness. The experimental observation was analysed using a FEM based simulation. The results of the simulation show that the local heating controls the stress as well as the thermal gradient ($|\nabla T|$). The migration due to local stress and the thermal migration add up to the electromigration induced mass migration and create the total mass migration. This leads to damage to the film. Our investigation shows that not only the average temperature growth but also the local spatial temperature variation plays a predominant role in determining the nature of net mass flow into or out of a region of the film under current stressing, and thus in the evolution of the damage process.
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