Effect of Impurities on Stress Induced Void Formation in Al-1%Si Conductors

Seong-Min Lee
Department of Materials Science and Engineering University of Inchon
177 Dohwa-dong, Nam-ku, Inchon, 402-749 South Korea

E-mail: smlee@inchon.ac.kr

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It is shown in the present study that during the HTS (hot temperature storage) test, the metal contamination by impure elements can be highly susceptible to the void formation, leading to the open failure of the power line in the memory device. Such a functional failure associated with the metal contamination was investigated to be dominant in the early stages of the HTS test while the formation of a stress-driven void is mainly observed in the later stages. In particular, it was found that the void formed in the contaminated metal takes on a slit-like shape which has been known to be characteristic of the stress-related voiding. The impure elements leading to the metal degradation were identified to be carbon and oxygen introduced during the metal sputtering process. The experimental works show that the device reliability was significantly improved by reducing the level of such impure elements within metal.

Keywords : metal conductor, stress migration, contamination, void, reliability

1 . INTRODUCTION

Nowadays, VLSI(very large scale integrated) devices consist of the extremely fine conductor line to maximize their memory capacity whose width is even less than 1㎛. Consequently, the metal open failure in such VLSI devices has been a serious concern because of their high susceptibility to the stress-related voiding. In particular, the stress-driven void formation is mainly found in high density memory devices using aluminium conductors whose grain structure takes on a "bamboo" appearance which grain boundaries run perpendicular to the length of metal lines. Moreover, it has been well known that the void formation in such Al conductors has a strong functional dependence on the compressive stress of its overlying passivation layer, which mainly carries out a role to protect the metal mechanically. The passivation layer is usually composed of quite dense and rigid materials such as silicon nitride to resist against absorption of moisture from outside through the plastic molding compound. Thus, it has been widely accepted that the stress-related voiding is a result of a tendency to relieve the residual stress caused by the thermal expansion mismatch between the metal and its passivation layer[1-4]. Therefore, the mechanical properties of the passivation layer has been considered to be the most critical factor in determining the metallic open failure. However, there are few reports showing that the open circuit failure of the Al conductor is also influenced by the presence of impurities in metals or corrosive environment[5,6]. It is not clear how and why the contaminated metal is highly susceptible to the open failure. So, it is studied in the present work whether there exists any correlationship between residual stress of the passivation layer and metal contamination and, if any, how it works.

2. EXPERIMENTS

Most of test devices were prepared by conventional VLSI circuit fabrication where Al alloys containing 1% Si as a metal conductor are sputter-deposited to a thickness of 0.8㎛ on thermally grown oxide (SiO_2) and patterned to a width of 0.5㎛ and then, passivated by the PECVD (plasma enhanced chemical vapor deposition) technique. However, for the evaluation of the stress-related voiding potential of Al interconnection lines during the HTS test, some of them were prepared under a controlled fabrication procedure so that the metal might contain a higher impurity level or the passivation layer might be subjected to a higher residual stress. That is, in order to investigate the migration response of the contaminated metal, the metal deposition process was changed in a way of controlling amount of impure elements within the metal. The quantitative analysis of the impurity level in such a metal was done utilizing SIMS (secondary ion mass spectrometry) where the relative quantities of the measured secondary ions are converted to concentrations to reveal the composition and trace its impurity content as a function of sputtering time. The mechanical property of passivation was also controlled by the variation of the vapor deposition conditions such as the reaction gas ratio, pressure and dual frequency ratio and then, its residual stress was estimated by the use of a Tencor FLX laser cantilever beam system. This laser beam technique utilizes the deformation of a bare Si by a passivation layer deposited on it as a measure of the passivation stress. In other wards, the passivation stress is calculated from the radius of curvature of a Si substrate with a given thickness and covered with passivation.7 The radius measurement is performed by scanning a laser beam across the Si substrate and measuring the change of the reflected angle with distance along it. Assuming the stress to be isotropic in the passivation layer, its stress can be expressed by

$$\sigma_p = \frac{\left(E_{Si}/1-v\right)\, t_{Si}^2}{6t_p\left(1/R_p - 1/R_{Si}\right)} \qquad (1)$$

where
Ep : young's modulus of the Si substrate
v : poisson's ratio
t_{Si} : the Si thickness
tp : the passivation thickness
R_{Si} : the radius of curvature of the origin bare Si
Rp : the radius of curvature after the passivation deposition

To estimate the temperature dependence of the passivation on stress, the combined stack of passivation on a bare Si substrate was heated to 450℃ with a 3℃/minute ramp rate and cooled down to room temperature. All stress measurements were performed using 8 " diameter Si wafer. The present work adopts four different types of devices and voiding susceptibility of their metal lines, which have a length longer than 10 ㎛, was evaluated by measuring the cumulative electrical failure during the HTS test. Three types of devices have passivation composing of double layered combinations of PECVD PEOX with a thickness of 1500Å followed by PECVD Si_3N_4 with a thickness of 6000Å, while one type of device includes an additional PSG layer used as a stress buffer layer to consist of triple layered combinations of 1500Å PEOX/3000Å PSG/6000Å Si_3N_4.

The reliability test, which is confined to the HTS (hot temperature storage), was performed at 150℃. Voiding susceptibility of metal lines, which have a length longer than 10㎛, was evaluated by measuring the cumulative electrical failure during the HTS test. Predetermined time required for the qualification of reliability was 1000 hours. For each set of experiments, nominally identical specimens were placed in a hot oven, then individuals were functionally tested after predetermined time (i.e. 150, 300, 500, 1000 hours). Once any functional failure was found, the corresponding specimens were decapsulated and examined under a scanning electron microscope (SEM) to identify whether the metal open failure takes place.

3. RESULTS

Table 1. The HTS test result.

	No. of specimen	No. of failure				Failure Rate (%)
		150 HR	300 HR	500 HR	1000 HR	
A type device	360	12	4	6	13	9.7
B type device	360	0	0	0	0	0

Two different types of devices have been used for the HTS test. The results summarized in Table 1 indicate that most of A type devices exhibit a functional failure associated with the metal voiding without any dependence on the test time while B type devices reveal excellent reliability up to 1000 hours. Fig. 1 shows a SEM micrograph of a metal open failure observed in A type devices where the electrical failure is detected during the HTS test. This figure indicates that a slit-like void forms across metal line without causing any damage to its overlying passivation. The microscopic examination also showed that the metal line of A type specimens has larger grain size which is enough for the grain boundary to cross it completely and the intergranular failure takes place without introducing any significant damage to the interior of a grain. Based on these observations, we see that the failure mode of A type samples is characteristic of stress migration.

In order to test the impact of the passivation stress on the present metal failure, passivation of two different types of devices was analyzed for the stress measurement utilizing the Tencor FLX laser cantilever beam system. Fig. 2 shows the measured stress of each passivation layer as a function of temperature. In this figure, we see that passivation of A type devices exhibits the compressive stress level higher than that of B type devices by 2-3 times. The relaxation rate of each p-layer stress was also plotted in Fig. 3 to show that passivation

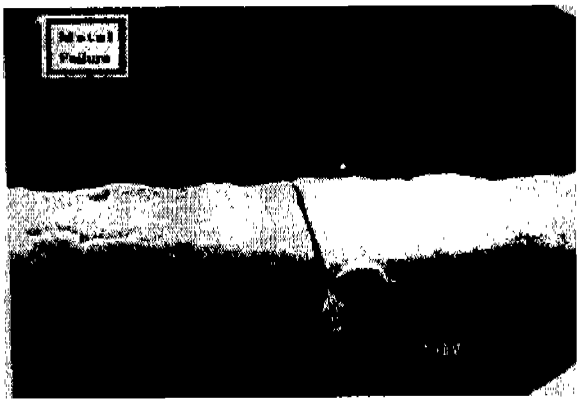

Fig. 1. SEM micrograph showing the metal open failure.

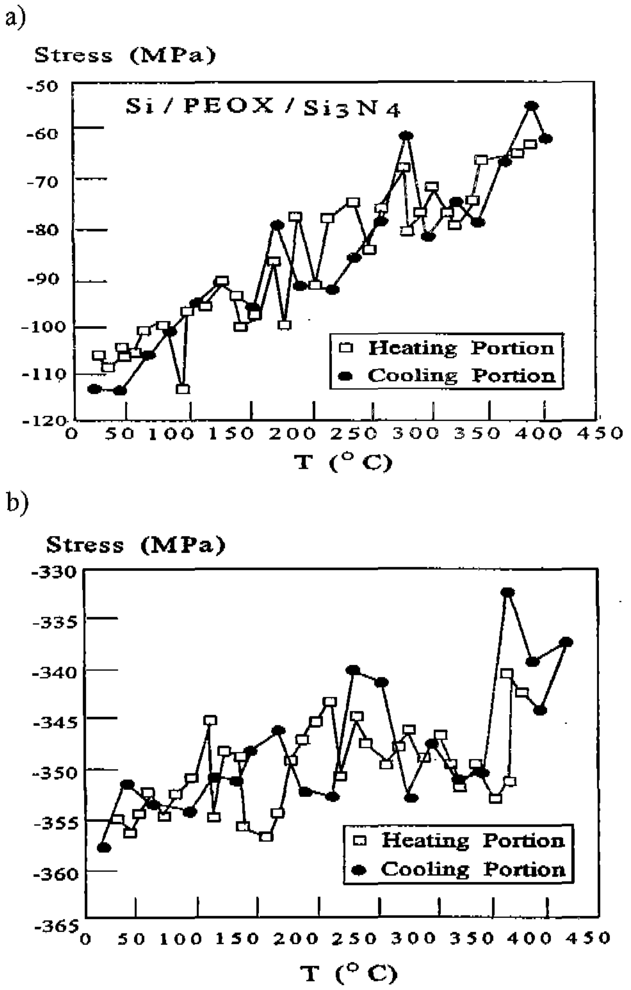

Fig. 2. Graphs showing the measured stress of passivation as a function of temperature; a) A type device and b) B type device.

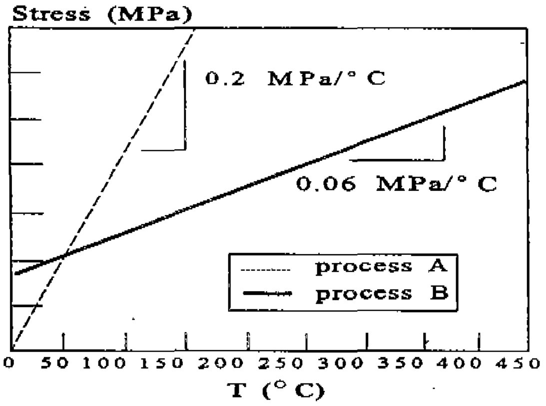

Fig. 3. The relaxation rate of the passivation stress as a function of temperature in two different processes.

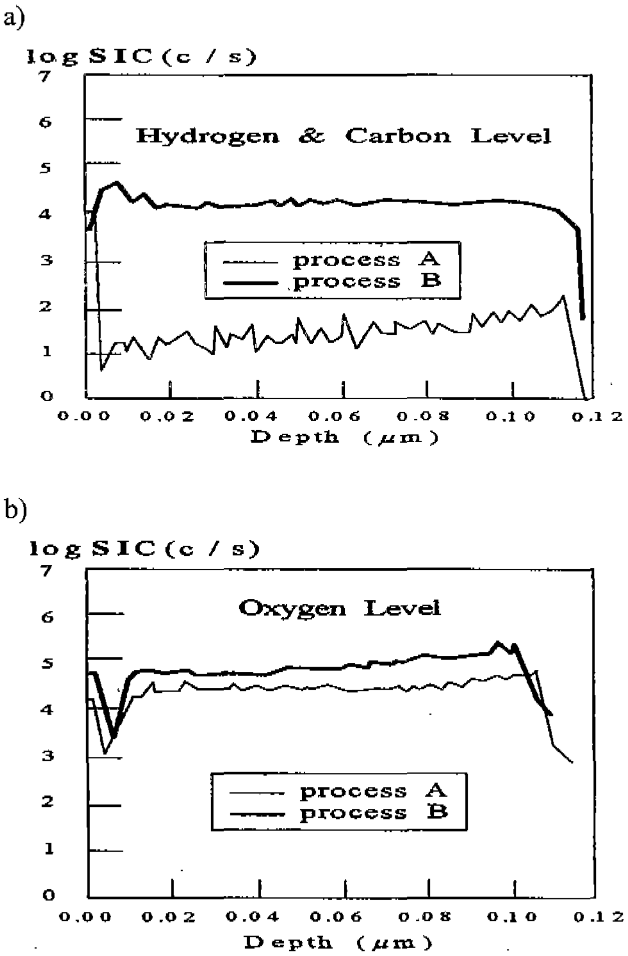

Fig. 4. The SIMS data showing the level of impure elements in two different metal layers; a) A type device and b) B type device.

of A type devices keeps its internal stress longer than that of B type devices by 2 times, indicating that A type passivation dose not have sufficient elasticity as compared to B type passivation. Therefore, there is no doubt that the underlying metal layer overcoated by such mechanically dissimilar layer as A type passivation will be highly susceptible to the stress-related migration.

Then, in order to find out another factor that might trigger the metal open failure, each metal film of different types of devices was examined for impurity content by secondary ion mass spectrometry. It was found that the metal line of A type device is severely contaminated by impure elements such as C and O_2 compared to that of B type device. Fig. 4 shows that the level of carbon is higher in the metal of A type device than that of B type device by more than 100 times and the oxygen level is also higher in the metal of A type device by 2-3 times. This result implies that such impure elements might segregate along a grain boundary to cause its partial embrittlement which can act as a source of the metal open failure. Consequently, reliability degradation in A type devices is due to both severe metal contamination and higher internal stress of passivation. However, it is questionable which one is more dominant factor in causing the present HTS failure. In order to see which is more significant in defining the present metallic failure, two test procedures were performed. First, the process revision was done in a way of reducing the internal stress of A type devices by the introduction of a stress buffer layer (i.e. PSG) between the metal conductor and its rigid passivation layer without any revision in the metal process. The resulting C type device composed of a still contaminated metal and passivation with triple layered combinations of 1500 Å PEOX/3000Å PSG/6000Å Si_3N_4 and so, it exhibits internal stress in its passivation layer as low as A type devices. Then, the cumulative failure rate of C type devices subjected to the HTS test was investigated to be 3.6% up to 300 hours and no failure was detected after that (Table 2). It was interesting to see that most of the failures in C type devices occur during the early stages of the HTS test.

Table 2. The HTS test result.

	No. of specimen	No. of failure				Failure Rate (%)
		150 HR	300 HR	500 HR	1K HR	
C type device	360	10	3	0	0	3.6
D type device	360	0	0	3	12	4.2

Table 3. The HTS test result.

Material	Al-1% Si	Al_2O_3	SiC
Rigidity	1	5.4	6.6

This result manifests that most of the metal open failures in A type devices, which are observed during its late stages, are due to the high stress of passivation. Now, the second approach adopted in the present work was to reduce the impurity level of the metal line in the A type devices without changing the passivation. Thus, the impurity level of the resulting D type device is similar to that of B type devices but its passivation layer is subjected to the high internal stress. The test result shows that D type devices reveal no functional failure up to 300 hours but the cumulative failure rate of 4.2% after 500 hours, observed mostly at 1000 hours (Table 2).

Therefore, it is concluded that although the high internal stress induced to the metal by the thermal constraint of passivation has been known to be a primary cause of stress-induced failure, an impure element such as carbon or oxygen could be another critical factor in determining the reliability of the metal conductor during the early stages of the HTS test.

4. DISCUSSION

During the cooling portion of the passivation process, the presence of a rigid passivation layer such as silicon nitride has a tendency to suppress the thermal expansion of its underlying metal layer. Consequently, the metallic conductor lies under a state of tensile stress after the manufacturing process of a device. Such an internal stress introduced to the metal is determined by geometry or physical properties of its passivation layer. Imagine a metal layer coupled with rigid passivation as shown in Fig. 5. The internal stress of the underlying metal can be expressed as a function of that of its passivation ($\sigma_{pass'n}$) as the following ; [8]

Above equation indicates that in a case of rigid passivation such as silicon nitride, thicker passivation can introduce higher internal stress to its underlying metal. The thick multilayer compromising an alternating sequence of mechanically dissimilar layer will also reduce the internal stress level induced to the metal. That is why the adoption of a viscous PSG layer prior to the rigid passivation coating such as silicon nitride suppresses the failure rate as shown in the result section

Various void formation mechanisms associated with internal stress have been already reported[1-5]. According to those mechanisms, the thermal displacement induced stress has been known to drive a vacancy migration toward the grain boundary running

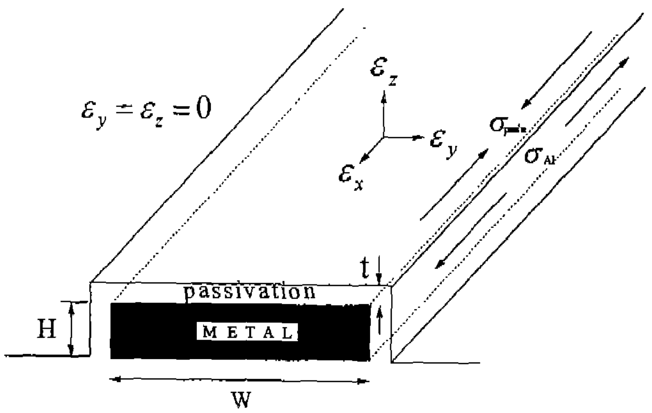

Fig. 5. The schemetic cross-section of the metallic line couples with a passivation layer.

$$\sigma_{Al} = \sigma_{pass'n} t \left(2/W + 1/H\right) \qquad (2)$$

Then,

$$\sigma_{Al} = E_{pass'n} \Delta\alpha \, t\left(2/W + 1/H\right) \qquad (3)$$

where

σ_{Al} = internal stress of the metal
t = thickness of the passivation layer
H = thickness of the metal
W = width of the metal
$E_{pass'n}$ = elastic modulus of the passivation layer
$\Delta\alpha$ = thermal expansion mismatch between the metal and passivation

perpendicular to the longitudinal direction of a metal line to release its high stored energy, resulting in the vacancy segregation along the grain boundary of the metal to form a slit-like void. Since voiding is a result of a time-dependent process, it requires enough duration time. This is the reason why the stress-related voiding is mainly observed during the later stage of the HTS test. Thus, the first step to minimize migration susceptibility will be to choose a proper passivation material with good thermal match to its underlying layer (i.e. low-stress passivation). However, this is not exactly the case of the present issue where the contamination of the metal line influences the metal open failure more significantly than the passivation stress dose (at least, during the early stage of the HTS test). Consequently, the stress-related voiding mode associated with the vacancy flow can not explain the present failure aspect clearly.

The influence of impure elements on stress voiding has been reported[5,6]. It has been known that grain growth in aluminum film is suppressed by the presence of the impure elements and such a retarding effect of impure elements on grain growth accelerates the propensity for the thermal voiding during the HTS test[5].

It was reported that high oxygen content in aluminum produces a few long, slit-like void rather than a circular shape, providing more chance of dendritic voids completely across the metal line[6]. That is presumably the reason why A type devices exibit higher failure rate than D type devices during the later stages of the HTS test. However, the reason why the contaminated metal is mainly failed during the early stages of the HTS test, in which there is not enough time for the activation process, is not clear yet.

An alternate explanation for some of the behavior observed in the present work is that the oxygen and carbon contamination could change the mechanical property of the metal, thereby increasing the stress level. For instance, carbon can easily segregate into grain boundaries of the Al metal without causing any significant strain because the atomic radius of carbon is about 0.15Å which is smaller than that of aluminum (~0.5Å). Such a segregated boundary has a tendency to suppress the plastic flow of aluminum, resulting in the local stress concentration. Therefore, if the segregation of carbon elements occurs along a particular grain boundary, which runs perpendicular to the length of metal lines, it could be a preferential site for the mechanical debonding. Consequently, the early failure is considered to occur as the metal line fails its integrity with an exposure to such local stress concentration. CO_2 gas in the sputtering chamber during the metallization process will be the obvious source of carbon contamination of the Al metal. Indeed, the decomposition of CO_2 gas into such an individual element as carbon or oxygen is thermodynamically stable at the temperature of 350℃ that the metallization process proceeds[7]. Moreover, it was identified by the SIMS data showing the high concentration of such impure elements within the metal layer just after the metallization process.

In conclusion, the early stage metal failure ($t<150$ hours) is caused by a mechanical debonding due to the segregated impure element along the grain boundary in aluminum and the later stage failure is by the conventional vacancy flow due to the high internal stress.

5. CONCLUSION

This article shows that even though the HTS failure in high density memory devices is a primary function of passivation stress, the metallic contamination can give another serious impact on the functional failure associated with a slit-like void formation. In particular, it was shown that carbon elements segregated along grain boundaries might allow the mechanical debonding of a grain boundary in metal conductors, leading to a characteristic fracture at a time of less than 150 hours corresponding to the early stage of the HTS test. The present experimental work also verified that the reduction of the carbon content in the metal significantly enhances the device reliability due to better resistance to the formation of a slit-like void.

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