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Citation: J. Appl. Phys. 98, 114911 (2005); doi: 10.1063/1.2139835
View online: http://dx.doi.org/10.1063/1.2139835
View Table of Contents: http://jap.aip.org/resource/1/JAPIAU/v98/i11
Published by the American Institute of Physics.

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Transmission electron microscopy observations of Cu-induced directional crystallization of amorphous silicon

Sung Bo Lee and Duck-Kyun Choi
Department of Ceramic Engineering, Hanyang University, Seoul 133-791, South Korea

Dong Nyung Lee
School of Materials Science and Engineering, Seoul National University, Seoul 151–744, South Korea

(Received 4 April 2005; accepted 31 October 2005; published online 14 December 2005)

By annealing at 500 °C for 1 h with an electric field of 180 V/cm, a Cu-deposited a-Si/glass sample undergoes a partial crystallization of amorphous Si, whose microstructure is characterized by conventional and high-resolution transmission electron microscopy. The Si crystallites grow in the (111) and (211) directions with their {011} planes parallel to the amorphous Si film surface, assuming a needlelike shape. Copper silicides are not observed at the leading edges of the crystalline Si needles. The growth directions of (111) and the (011) surface plane orientations are accepted to arise from elastic anisotropy of c-Si and can be explained by considering the strain energy under the uniaxial stress state and the plane stress, equibiaxial strain state. Crystallization behavior of a-Si in the Ni/a-Si interface of NiSi2/a-Si is nearly equal to that of crystalline Si (c-Si) (0.5406 nm). At about 500 °C, the epitaxial c-Si is nucleated on one or more of the NiSi2 {111} surfaces and the c-Si needles grow in the (111) directions in the amorphous matrix with the [011] orientations parallel to the film surface normal. The precipitates are observed to be always present at the leading edges of c-Si needles. Hayzelden and Batstone suggested that the chemical potential of the Ni atoms is lower at the interface of NiSi2/a-Si than that at the interface of NiSi2/c-Si, and the chemical potential of Si atoms shows the exact reverse. The chemical potential differences drive the diffusion of Ni atoms into the NiSi2/a-Si and that of Si atoms into the NiSi2/c-Si interface, leading to the migration of the NiSi2 precipitate toward the a-Si and the formation of c-Si behind the precipitate. Lee et al. examined effects of various metals on the crystallization rate in the FALC process. Mask patterns with various shapes and dimensions from 5 to 120 μm were used. After annealing at 500 °C with an electric field of 2.1 V/cm, Ni produces the crystallization rate of 9.5 μm/h, but Cu induces the crystallization of 35 μm/h, much higher than that of Ni. For Cu, the crystallization rate with the electric field of 2.1 V/cm is much higher than that without an electric field (1.5–2 μm/h). However, the mechanism of crystallization enhanced by Cu has not been clearly resolved.

Our purpose in this study is to explore how the Cu atoms enhance the crystallization. In the present study, Si crystallites grow in specific crystallographic orientations, having a needlelike shape. Based on the strain energy consideration, we predict growth directions of c-Si and find that the prediction agrees with the observations. The analysis is also successfully applied to the NiSi2-mediated crystallization. A specimen fabricated by the FALC process has been observed by transmission electron microscopy (TEM) for the study. As indicated before, without any electric field, crystal-
lization does not occur very well, and therefore, it is expected that the observation of crystallization is facilitated in a crystallized specimen fabricated with an electric field.

II. EXPERIMENTAL PROCEDURE

Thin 80 nm a-Si was deposited on Corning 1737 glass by plasma-enhanced chemical vapor deposition (PECVD) at 280 °C using SiH4 and H2 as source gases. The wafer was cut into 3 × 3 cm2 square specimens for the subsequent process. To deposit a metal (Cu) on the a-Si film, the photore sist (PR) was spin-coated and patterned by a photolithographic process. In this study rectangular mask patterns of 60 × 30 μm2 were used. A thin 2 nm Cu layer was deposited in a dc sputtering system. Copper on the PR patterns was removed by the lift-off method and only that deposited on the a-Si was left.

The crystallization annealing was done at 500 °C for 1 h in N2 ambient. During the crystallization annealing an electric field of 180 V/cm was applied to the above metal-deposited specimen by the dc power supply. Details concerning the specimen preparation are presented elsewhere.12

From the crystallized specimen, 3 mm disks were cut using an ultrasonic cutter for TEM specimen preparation. For the plan-view studies of the a-Si film, the disks were mechanically back-thinned to a thickness of about 10 μm, and then ion milled on the glass substrate side at an accelerating voltage of 5 kV using a PIPS. The crystallization microstructure was characterized by transmission electron microscopy (TEM; JEM2010, JEOL Co., Ltd., Japan). High-resolution TEM (HRTEM) was performed at 400 kV in a JEM4010 (point-to-point resolution: 0.15 nm, JEOL Co., Ltd., Japan). Energy dispersive x-ray spectroscopy (EDXS) was applied for composition analysis of a-Si and c-Si regions.

III. RESULTS

Figure 1(a) is a schematic diagram of the patterned structure composed of the small patterns of 60 × 30 μm2 in size. An enlarged view of the patterns between the electrodes is shown in Fig. 1(b). The crystallization front migrates from the negative electrode side to the positive electrode side. A needlelike morphology of c-Si was clearly observed at the interface between the fully crystallized and amorphous regions [indicated by an “O” in Fig. 1(b)], as shown in Fig. 2(a). Interestingly, as revealed in Figs. 2(b) and 2(c), almost all the c-Si needles exhibited (011) orientations with respect to the a-Si film surface normal (i.e., {011} surface plane orientations) and grew in the (111) and (211) directions, forming an angle of 90° between the needles. The [011] diffraction pattern taken from the needle “A” in Fig. 2(b) [Fig. 2(c)] shows streaks along the [111] direction, indicating that there are many stacking faults or twins normal to that direction. (The diffraction pattern taken from the needle “B” is not shown here.) The a-Si region was measured to contain (2.99±0.97) at. % Cu on average by EDXS measurement of five areas. In fully crystallized regions many needles were interwoven, as shown in Fig. 3. The regions contained (1.63±0.15) at. % Cu on average by EDXS measurements of four areas. The solubility of Cu in c-Si is negligible, and therefore, the detected Cu is believed to arise from trapped Cu solutes in the interfaces between the interwoven c-Si needles. Figure 4(a) shows a high-resolution TEM image

FIG. 1. (Color online) (a) A schematic diagram of the patterned structure composed of the small patterns of 60 × 30 μm2 in size and (b) an optical image of the patterns between the electrodes.
also revealing the growth directions of (111) and (211) with the \{011\} surface orientation. At a higher magnification [Fig. 4(b)], three \{111\} twins and a stacking fault were seen, in agreement with the indication of Fig. 2(c). In order to consider the role of any Cu silicides on the nucleation and growth of the c-Si needles, the edges of the c-Si needles were examined. Figure 5 displays the typical edge, where Cu silicides were not observed, and instead, the Si lattice image was shown. Furthermore, Cu silicides were not observed, even in the a-Si matrix and the crystallized region within the limited field of view. The c-Si region at the edge contained many \{111\} twins and stacking faults indicated by white triangles and arrows, respectively. There also seemed to be some lattice distortion in some areas, due to which defects (twins or stacking faults) were not clearly visible.

IV. DISCUSSION

As for the NiSi$_2$-mediated crystallization\cite{10,13-15} in this study, crystallization proceeded in the \{111\} directions with a shape of needles as shown in Figs. 2(b) and 4(a). The c-Si needles had the \{011\} surface plane orientations. Additionally, as clearly shown in Figs. 2(b) and 4(a), the growth occurred also in the \{211\} directions. At variance with the NiSi$_2$-mediated crystallization\cite{10,13-15}, however, as shown in Fig. 5, Cu silicides were not observed at the leading edges of the c-Si needles, indicating that the crystallization of the Cu/a-Si system in this study was not mediated by any Cu silicides. Radnoczi et al.\cite{4} suggested that Au, Sb, In, and Al, which form eutectic with Si, dissolved in the a-Si film may loosen the covalent bonds in Si and make the a-Si even unstable, enhancing crystallization. Even though Cu is a silicide-forming metal, Cu atoms appear to enhance crystallization in a similar way to the eutectic-forming metals. Since Cu has a negligible solubility in crystalline Si,\cite{16} the Cu...
solute atoms are repelled by the $c$-Si. The Cu atoms will diffuse into the $a$-Si matrix, making the $a$-Si unstable, and the Si atoms will migrate into the $c$-Si side from the $a$-Si, resulting in crystallization. Diffusion of Cu atoms in $a$-Si seems to be rate controlling. Without the application of any electric field during annealing at 500 °C [i.e., in the MILC process], the crystallization rate is reported to be $1.5–2$ μm/h.$^{12}$ Copper is reported to diffuse in $a$-Si with $D > 10^{17}$ cm$^2$/s in a temperature range between 400 °C and 600 °C.$^{13}$ From the diffusivity value of $10^{12}$ cm$^2$/s, we calculate the minimum diffusion length ($\sqrt{D t}$) to be 0.6 μm for 1 h at 500 °C. This diffusion length qualitatively agrees with the crystallization rate obtained without any electric field in the MILC process ($1.5–2$ μm/h). Lee et al.$^{12}$ showed the increase in the crystallization rate with an electric field of 2.1 V/cm to 35 μm/h. Although some crystallization parameters such as pattern size, shape, and the applied electric field intensity are different in the present study, a high crystallization rate of $\sim 40$ μm/h was obtained, as shown in Fig. 1(b). The crystallized pattern image in Fig. 1(b) reveals that the macroscopic crystallization direction under an applied bias corresponds to that of an electric field in the pattern.

It is suggested that the electric field in the FALC process can cause a kind of electron wind effect (electromigration) via current flow in the pattern at the elevated temperature.$^{18}$ In Ref. 18, the electron wind effect is proposed to explain the NiSi$_2$-mediated crystallization$^{3,14}$ and is limited to the diffusion of Ni in NiSi$_2$. When the current density is high enough, the momentum exchange induces the (Ni) atoms in NiSi$_2$ to migrate from the negative electrode side to the positive one (along the electron flow direction).$^{18}$ In the present study, because no Cu silicides were observed at the leading edges of the $c$-Si needles, the application of the above-mentioned suggestion is not plausible. However, the electron wind effect is likely to hold for the diffusion of Cu in $a$-Si at the high temperature of 500 °C in the present study, because the resistivity of $a$-Si decreases with the increase of temperature because of the exponential increase in intrinsic carrier concentration, and at such a high temperature, the electron flow in $a$-Si can cause the electron wind effect. (A more theoretical consideration of the electron wind effect in $a$-Si at high temperatures is in progress.) The effect will accelerate the diffusion of Ni along the electron flow direction, and consequently, the crystallization rate.

The crystallization behavior of $a$-Si in the Cu/$a$-Si bilayer (without any electric field in this case) was studied by Russell et al.$^{19}$ On heating to 175 °C, the Cu$_3$Si phase appears and subsequent heating to 485 °C results in the crystallization of the $a$-Si in the form of dendrites in the Cu$_3$Si matrix. Different from the results of Russell et al.$^{19}$ the matrix was observed to be still $a$-Si in the present study and discernible Cu silicides, including Cu$_2$Si phase, were not observed. In the study by Russell et al.$^{19}$ the thickness ratio of Cu to $a$-Si was determined to obtain an average composition of Cu$_{1.0}$Si$_{0.6}$, where $\delta$ is from 0.5 to 1. In this case the matrix (continuous phase) could become Cu$_2$Si phase. In the present study, however, the deposited Cu film (2 nm thick) is very thin, with the thickness ratio to the $a$-Si underlying layer (80 nm) of 1:40. Therefore, as mentioned in the Results section, they could not be easily observed in the present study. It may be that they act as heterogeneous nucleation sites for crystallization. However, the difficulty in observing the phases indicates that their nucleation density was too low and their size was too small to explain the observed, overall crystallization behavior, strongly implying that the presence of Cu solutes, not Cu silicides, enhances crystallization.

The $\{111\}$ growth directions and the $\{011\}$ surface plane orientations observed in the present study may be caused by anisotropic elastic properties of $c$-Si, although $a$-Si is isotropic, because the strain energy can influence the crystallization rate. The higher strain energy will give rise to the higher crystallization rate. One of the major strain energy sources may be the thermal strain energy due to differences in thermal expansion coefficient between $a$-Si and the substrate glass. The thermal strain energy can influence the crystallization rate, but is not related to the directional crystallization. Since the thermal expansion coefficients of $a$-Si and $c$-Si are expected to be almost the same, the thermal strain energy between $a$-Si and $c$-Si can be negligible.

Another strain energy may arise from the accommodation strain between $a$-Si and $c$-Si due to different structures and densities [the densities of $a$-Si and $c$-Si are measured to be 2.1–2.3 (Ref. 20) and 2.32–2.34, $^{21}$ respectively]. This strain energy can be anisotropic because of anisotropic elastic properties of $c$-Si, even though $a$-Si is isotropic. Since the thickness of the Si film, including $a$-Si and $c$-Si, is much smaller than the dimensions along the film surface, the stress in the plane normal direction (ND) is negligible, and the film is under the plane stress. Initial Si crystallites will form in the surface layer, where the Cu concentration is highest, and they could be of a disk shape. The surface orientation of $c$-Si disks is presumably random, because they formed from amorphous Si. However, their growth rates will vary with their orientations, because different orientations give rise to different strain energies. The strain energy of the disk is approximated by that of a thin film under plane stress-equibiaxial strain. When $c$-Si grows with the needlelike shape, however, the stress is expected to be much higher in the axial direction (AD) than in the transverse direction (TD) because of the longer axial dimension, and therefore, the stress state of $c$-Si needles can be approximated by the uniaxial stress state. That is, the stress state of the initial stage of the crystallization of $a$-Si (disk formation) is approximated by the biaxial stress and that of the growing stage (needlelike shape) by the uniaxial stress.

Elastic properties of thin films of the cubic system have been studied by Lee.$^{22}$ For a material under the uniaxial stress, we define Young’s modulus, $E$, in the direction of stressing, as the uniaxial stress divided by the longitudinal strain in that direction. For a material with cubic symmetry, Young’s modulus $E$ is given by

$$\frac{1}{E} = S_{11} + \frac{S_{44} - 2(S_{11} - S_{12})}{a_{11}a_{12}^2 + a_{12}^2a_{13}^2 + a_{13}^2a_{11}^2},$$

(1)

where $S_{11}$, $S_{44}$, and $S_{12}$ are the compliances referred to the symmetric axes and $a_{ii}$ are the direction cosines of the uniaxial stress direction referred to the symmetric axes. For
the \([hkl]\) direction, \(a_{11} = h/\sqrt{h^2+k^2+l^2}\), \(a_{12} = k/\sqrt{h^2+k^2+l^2}\), and \(a_{13} = l/\sqrt{h^2+k^2+l^2}\).

For silicon, \(S_{11} = 0.00786\), \(S_{12} = -0.00214\), and \(S_{44} = 0.01256\) GPa. From Eq. (1), the maximum and minimum Young’s moduli of \(-c\)-Si are calculated to be 185.9 GPa in the \(\{111\}\) directions and 127.2 GPa in the \(\{001\}\) directions, respectively. For reference, we calculate Young’s moduli along directions normal to the \([111]\) direction. The \((hkl)\) directions normal to the \([111]\) direction are obtained from a relation of \(h+k+l=0\). For \(h+k+l=0\), \((a_{11}^2 a_{12}^2 + a_{12}^2 a_{13}^2 + a_{13}^2 a_{11}^2) = 1/4\) in Eq. (1). Therefore, Young’s moduli along directions normal to the \((111)\) directions are the same and calculated to be 166.7 GPa. The strain energy \(W\) per unit volume of a specimen strained by \(e\) under the uniaxial stress is given by \(W = e^2E/2\), and the Young’s modulus, \(E\), can function as measures of the stiffness and the strain energy for the uniaxial stress state. Thus, for Si, the \(\{111\}\) directions are the stiffest and have the highest strain energy density, which indicates that the \(-a\)-Si matrix contacting the \(\{111\}\)-oriented needles receive the highest strain energy density along the needle growth directions, accelerating the crystallization along the \(\{111\}\) directions.

Likewise, for the plane stress-equibiaxial strain state, the strain energy per unit volume of a cubic-system crystal strained by \(e\) along the \((hkl)\) plane is given by

\[
W = \frac{e^2}{2} \left[ S_{11} + S_{12} - 2S_{11}^2 \right] (a_{11}^2 a_{12}^2 + a_{12}^2 a_{13}^2 + a_{13}^2 a_{11}^2),
\]

for the plane stress-equibiaxial strain state, the \(\{111\}\) disks of \(-c\)-Si needles can grow extensively and occupy large areas within the \(-a\)-Si films, resulting in more frequent observations.

We want to find the orientation of initial crystal disks formed in the surface layer that can develop into the \(\{111\}\)-oriented \(-c\)-Si needles. We can expect that the disk crystals

<table>
<thead>
<tr>
<th>((hkl))</th>
<th>(\bar{1}2)</th>
<th>(\bar{2}3)</th>
<th>(\bar{2}5)</th>
<th>(\bar{3}4)</th>
<th>(\bar{4}5)</th>
<th>(\bar{5}6)</th>
<th>(\bar{7}8)</th>
<th>(\bar{1}0)</th>
</tr>
</thead>
<tbody>
<tr>
<td>(\bar{M}) (Si), GPa</td>
<td>209.4</td>
<td>209.7</td>
<td>210.3</td>
<td>210.8</td>
<td>211.2</td>
<td>211.8</td>
<td>212.1</td>
<td>212.5</td>
</tr>
</tbody>
</table>

Table I. The \(\bar{M}\) values for planes parallel to the \([111]\) direction in Si.

\[
[a_{ij}] = \begin{bmatrix}
\frac{-k}{\sqrt{h^2+k^2+l^2}} & \frac{h}{\sqrt{h^2+k^2+l^2}} & 0 \\
\frac{-hl}{\sqrt{h^2+k^2+l^2}} & \frac{h}{\sqrt{h^2+k^2+l^2}} & \frac{h^2+k^2}{\sqrt{h^2+k^2+l^2}} \\
\frac{h}{\sqrt{h^2+k^2+l^2}} & \frac{k}{\sqrt{h^2+k^2+l^2}} & \frac{l}{\sqrt{h^2+k^2+l^2}}
\end{bmatrix}
\]

Here \(\bar{M}\) can function as measures of the stiffness and the strain energy for the plane stress, equibiaxial strain state.

The \(\bar{M}\) values for the \(\{001\}\), \(\{211\}\), \(\{111\}\), and \(\{111\}\) planes of silicon are calculated to be 175, 209.4, 213, and 223 GPa, respectively. Actually, out of the biaxial elastic moduli of all orientations, \(\bar{M}_{(111)}\) is the highest and \(\bar{M}_{(001)}\) is the lowest. Therefore, the \(\{111\}\) layers in contact with \(-a\)-Si film are the stiffest and have the highest strain energy density, which also implies that the \(-a\)-Si layers in contact with the \(\{111\}\) layers receive the highest strain energy density. The \(-a\)-Si layers with the highest strain energy density are likely to undergo crystallization with the highest rate.

Because the \(\{111\}\) directions have the highest Young’s moduli and the planes normal to those directions have the highest \(\bar{M}\) value, it may happen that in the \(-a\)-Si film surface, the \(\{111\}\) disks of \(-c\)-Si are selectively nucleated parallel to the film surface and the growth would occur in the \(\{111\}\) directions (i.e., normal to the film surface). However, such needles are unlikely to be observed due to their short dimension. When the \(\{111\}\) orientations are parallel to the film surface, the \(-c\)-Si needles can grow extensively and occupy large areas within the \(-a\)-Si films, resulting in more frequent observations.
whose planes have the highest $\bar{M}$ value grow fastest among those whose planes are parallel to both the film surface and the (111)-oriented needles. In this case, planes with maximum $\bar{M}$ value among planes parallel to the (111) directions are likely to be parallel to the surface. The \{hkl\} planes parallel to the (111)-oriented needle are obtained from a relation of $h+k+l=0$. These planes and their $\bar{M}$ values are given in Table I. The maximum and minimum $\bar{M}$ values are obtained for the \{011\} and \{211\} planes, respectively. Therefore, the initial disk c-Si crystals whose \{011\} planes are parallel to the surface grow fastest in the ND and have the best chance to grow in the (111) directions, leading to the formation of the \{011\} (111)-oriented needles, as revealed in Figs. 2(b) and 4(a) and schematically shown in Fig. 6. As noted before, the Young’s moduli along directions normal to the (111) directions are equal, while $\bar{M}$ values of planes parallel to the (111) directions have the highest value for the \{011\} planes.

The growth of the (111)-oriented needles may be interrupted probably by inhomogeneity of Cu solute distribution in the a-Si. At this time, the additional growth in the (211) directions may occur with the help of many \{111\} twins and stacking faults observed in the c-Si [Figs. 4(b) and 5], which are certainly due to the low stacking fault energy of Si ($\sim 50 \text{ mJ/m}^2$), even though the (211) orientations are not favorable in terms of the strain energy consideration. These \{111\} twins and stacking fault planes can generate steps for growth at the interface between the c-Si and a-Si, enhancing the growth in directions parallel to the twin and stacking fault planes. Because the \{011\} planes have the highest $\bar{M}$ value out of the planes normal to the (111) growth directions (Table I), even though there is some change in the growth direction into (211), the \{011\} surface plane orientations will remain unchanged, as revealed in Figs. 2(b) and 4(a).

For the NiSi$_2$-mediated crystallization, the \{011\} (111)-oriented crystallization is obtained$^{13,14}$ as for the Cu-enhanced crystallization. However, for the NiSi$_2$-mediated crystallization,$^{13,14}$ crystallization of a-Si occurs on the flat \{111\} faces of the octahedral NiSi$_2$ precipitates, and during crystallization annealing, the initial octahedral shape of the NiSi$_2$ precipitates at the leading edges of the c-Si needles changes into the thin plate shape,$^{13,14}$ whereas for Cu-enhanced crystallization, the Cu$_3$Si phases are not observed and the leading edges are not flat. For the NiSi$_2$-mediated crystallization, the \{211\}-oriented needles have not been reported, which are easily observed for Cu-enhanced crystallization. The fan-shaped growth and thinning of the NiSi$_2$ plate take place for the NiSi$_2$-mediated crystallization. These differences seem to be attributable to the stability of NiSi$_2$.

We consider the effect of NiSi$_2$ on the formation of the \{011\} (111)-oriented c-Si needles. The NiSi$_2$ crystal belongs to the cubic system, and its measured elastic constants are not available. Its calculated compliances are as follows:$^{24}$

$$S_{11} = 0.012 \text{ 19 GPa}^{-1}, \quad S_{44} = 0.018 \text{ 87 GPa}^{-1},$$

$$S_{12} = -0.005 \text{ 05 GPa}^{-1}.$$  

From Eq. (1), the maximum and minimum Young’s moduli of NiSi$_2$ are calculated to be 143 GPa in the (111) directions and 82 GPa in the (100) directions, respectively. The \{001\} and \{111\} planes of NiSi$_2$ are calculated to have the lowest (140 GPa) and the highest $\bar{M}$ (220.3 GPa), respectively, as for c-Si. The $\bar{M}$ values for planes parallel to the (111) direction of NiSi$_2$, $\bar{M}$ (NiSi$_2$), are given in Table II. The maximum and minimum $\bar{M}$ (NiSi$_2$) values are obtained for the \{011\} and \{211\} planes, respectively, as for c-Si. In addition to similar anisotropic elastic trends of c-Si and NiSi$_2$, any lattice plane of c-Si is likely to have a close match with the same plane of c-Ni$_2$Si, because the Ni$_2$Si crystals have a close lattice parameter match with c-Si (−0.4%).

Among the NiSi$_2$/c-Si interfaces, the \{111\} NiSi$_2$/c-Si interfaces are elastically the stiffest, because the \{111\} layers of the both crystals are elastically the stiffest. Therefore, the region surrounding the interface edge is strained, which in turn gives rise to crystallization along the interface plane as well as along the needle axis. In this way, the c-Si needles fan out during its growth, and simultaneously, the NiSi$_2$ precipitates become thinner. As the interface area increases, the accommodation strain increases to a point such that coherency between the two crystals cannot be maintained, resulting in dissociation of the NiSi$_2$ crystal into a few smaller crystallites. These smaller precipitates continue to mediate the growth.$^{14}$ Because the growth is mediated by the NiSi$_2$ whose equibiaxial-strain elastic moduli show a stronger anisotropy than that of c-Si (Table II), the \{011\} (111)-oriented growth for the NiSi$_2$-mediated crystallization remains without growth in other directions (e.g., \{211\} for the Cu-enhanced crystallization).

### V. Concluding Remarks

The crystallization behavior of a Cu-deposited a-Si/glass sample annealed at 500 °C for 1 h with an electric field of 180 V/cm has been examined by TEM. The Si crystallites grow in the (111) and (211) directions with their \{011\} planes...
parallel to the a-Si film surface, assuming a needlelike shape. No Cu silicides are observed at the leading edges of the c-Si needles, and therefore, it seems that the Cu atoms in the a-Si film make the a-Si unstable, enhancing crystallization, as in the case of the eutectic-forming metals. The growth directions of the (111) and the {011} surface plane orientations can be explained based on the strain energy consideration using the concept of an equibiaxial-strain elastic modulus. In c-Si, the (111) directions have the highest equibiaxial-strain elastic modulus and can give the highest strain energy density to the a-Si film in that direction, resulting in the fastest crystallization in the direction. The {011} planes have the highest strain energy density out of the planes parallel to the (111) directions and are likely to be parallel to the film surface. The (211) directions will be possible with the help of many {111} twins and stacking faults in c-Si. The NiSi$_2$-mediated, directional crystallization of a-Si can be also explained in terms of the strain energy consideration.

ACKNOWLEDGMENTS

One of the authors (SBL) thanks Kyung-Sook Jeon for the TEM specimen preparation and Young-Woong Kim and Jeong-Ung Kim for the preparation of the crystallization specimens. The help of Dr. Young-Ho Lee at the JEM 4010 is appreciated. This work was financially supported by a Korea Research Foundation Grant (KRF-2004-005-D00167).