Anomalous properties found in Cu films near below 100 nm thick deposited on Ta

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Abstract

For the Cu films sputter-deposited at ambient temperature on a Ta buffer layer with a clean surface (the \textit{nex}-Cu film) and those on a Ta buffer layer with native oxide (the \textit{ex}-Cu film), the internal stress along the film surface, \(\sigma_{il}\), the Young’s modulus, \(E_f\), the internal friction, \(Q^\prime\), the decrease in \(Q^\prime\) after annealing at 400 K, \(\Delta Q^\prime_{100 \text{K}}/Q^\prime\), and the root mean square of surface roughness (the RMS roughness) were studied for the Cu film thickness, \(t_{\text{Cu}}\), between 5 and 1000 nm. For the \textit{nex}-Cu films with \(t_{\text{Cu}}\) below 100 nm, \(E_f\) showed a deviatory decrease from its theoretical value, \(E_{f,\text{th}}\), which was estimated from the crystallographic texture and \(\sigma_{il}\), suggesting that the constituent static anelastic strain due to grain boundaries (GBs), \(\varepsilon_{\text{GB}}\), was, e.g. as large as about 25\% of the constituent elastic strain, \(\varepsilon_s\), at \(t_{\text{Cu}}\) of 10 nm. The local maximum of the RMS roughness and the local minimum of \(\Delta Q^\prime_{100 \text{K}}/Q^\prime\) were found near below 100 nm, indicating that the properties of GBs in already deposited Cu film were modified by additional deposition (the capping effect, below) and the anomalous capping effects took place near below 100 nm in \(t_{\text{Cu}}\). For the \textit{ex}-Cu films, \(E_f\) showed good agreement with \(E_{f,\text{th}}\) except that a local decrease in \(E_f\) was found for \(t_{\text{Cu}}\) near below 100 nm. The local maxima of \(\Delta Q^\prime_{100 \text{K}}/Q^\prime\) and the RMS roughness were found for \(t_{\text{Cu}}\) near below 100 nm too. It is suggested that the properties of GBs in the \textit{ex}-Cu films were of those in bulk metal for \(t_{\text{Cu}}\) below a few tens of nanometer and then changed to those in nanocrystalline metal with increasing \(t_{\text{Cu}}\). The capping effects were also observed for the \textit{ex}-Cu films.

Keywords: Cu thin film; Elastic property; Morphology; Internal friction; Texture; Thermal stability; Ta buffer layer; Surface oxidation

1. Introduction

Due to the lower electric resistivity and the higher melting point, interconnections in the high performance integrated circuits are changed from Al-based alloy films to Cu films. Understanding and control of the film property are important for the further downscaling in future because of a decrease in the circuit reliability [1]. Electromigration and stressmigration failures are of the most important issue in metallization, where the film thickness dependence of the Young’s modulus, the internal stress, the crystallographic texture, the grain size and the property of the grain boundaries should be understood [1–3]. It is known that the adhesion behavior of Cu on a diffusion barrier is one of the critical issues in Cu metallization [4,5]. For Cu deposition on Ta buffer layers [6], it is reported that Cu will grow conformally on the clean Ta surface or in wetting condition. In this case, Cu grows in a layer-by-layer growth mode (Frank–Van de Merwe (FM) mode) at first and its growth mode changes from the initial layer-by-layer growth to the three-dimensional-nuclei formation (Stransky–Krasstanov (SK) mode) due to increased internal stress with increasing Cu thickness. On the other hand, in the dewetting condition such as the deposition on the oxidized Ta surface, Cu will form three-dimensional islands on the surface (Volmer–Weber (VM) mode). Since semiconductor fabrication is typically carried out under conditions in which incidental contamination of the barrier surface by oxygen is relatively uncontrolled [6], the effects of such surface contamination should be pursued for the elastic property of Cu films.

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Meanwhile it is reported that the property of grain boundaries (GBs) in nanostructured fcc metals are considerably different from those in bulk fcc metals and GBs in nanostructured fcc metals play an important role on the elastic property and the material transport process. The pioneer work on the elastic property of 100 nm-thick Al films reported that the anelastic relaxation associated with GBs was observed at much lower temperatures than in bulk polycrystalline specimens [7]. For Al films thinner than 100 nm, it was reported that the internal friction associated with the grain boundary anelastic process (GBAP) starts to increase at about 200 K and the Young’s modulus at 300 K shows a decrease with decreasing film thickness, especially below ~20 nm [8–10]. The similar elastic responses were observed for Ag [11–13] and Cu [14] films. In the previous work on Cu films, the elastic property and thermal stability of Cu films deposited on Ta buffer layer were found to show anomalous behaviors at the Cu thickness near below 100 nm. In the present work, we studied on the effects of Ta buffer surface contamination and the Cu thickness on the Cu film growth through the elastic property and the surface morphology of Cu films.

2. Experimental procedure

The elasticity measurements of thin Cu films were carried out by means of the composite reed method. The shape of a Si reed-substrate used here is shown in Fig. 1, where a Cu film was deposited on the reed-substrate with a 30 nm thick Ta buffer layer. After deposition of the Cu film, the resonant frequency, \( f \), and the internal friction, \( Q^{-1} \), of the composite reed were measured at around 300 K. Then the warm up measurements were repeated twice between 80 and 400 K with the heating rate of about 1 K/min.

The Young’s modulus, \( E_y \), and the internal friction, \( Q_y^{-1} \), of a specimen film may be determined by the following equations [7,8,14],

\[
E_y = \left( \frac{E_s}{3} \right) \left[ 2 \left( \frac{f_1}{f_2} \right) + \frac{t_1}{t_2} \left( \frac{\rho_s}{\rho_y} \right) \right] \frac{1 - \nu_s^2}{1 - \nu_y^2} \tag{1}
\]

and

\[
Q_y^{-1} = \Delta Q^{-1} \left( \frac{t_1}{3t_f} \right) \left( \frac{E_y}{E_s} \right) \tag{2}
\]

where the subscripts ‘f’ and ‘s’ denote the quantities for the specimen film and the Si reed-substrate with the Ta buffer layer, respectively. \( E, f, \rho \) and \( \nu \) denote the Young’s modulus, the film thickness, the density and the Poisson ratio, respectively. \( \Delta f/f_s = (f-f_s)/f_s \) and \( \Delta Q^{-1} = Q^{-1} - Q_s^{-1} \), where \( f_s \) and \( Q_s^{-1} \) are the resonant frequency and the internal friction of the reed-substrate and \( f \) and \( Q^{-1} \) are those of the composite vibrating reed with the specimen film, respectively. For characterization of the elastic property of the specimen films, calibration for \( f_s \) and \( Q_s^{-1} \) is requested.

Deposition of a Ta buffer layer and a Cu film to the reed substrate was carried out by DC magnetron sputtering at ambient temperature in 1.3×10⁻¹⁵ Pa Ar with zero bias-voltage. The film growth rates used for Ta and Cu depositions were about 2 and 4.5 nm/min, respectively. The nominal purity of Ar gas was 6 N. Before deposition, the Si reed substrate was chemically etched and terminated by hydrogen. In the present study, a Cu film was deposited on a Ta film without or with native oxide of Ta as schematically shown in Fig. 2. After deposition of a Ta buffer layer 30 nm thick, calibration for \( f_s \) and \( Q_s^{-1} \) was made, where the Si reed-substrate with the Ta buffer layer was exposed to the atmosphere during its transfer to the measurement chamber resulting in the formation of native oxide on the Ta surface. Then a Cu film was deposited on the air-exposed Ta surface of Si reed-substrate and the elasticity measurement of the composite vibrating reed was carried out. Such Cu films will be called the exposed (ex-) Cu films below. In order to minimize the effect of surface oxidation of the Ta buffer layer, after calibration for \( f_s \) and \( Q_s^{-1} \) of the Si reed-substrate with the Ta buffer layer 30 nm thick, an additional Ta buffer layer 5 nm thick was deposited on the air-exposed Ta buffer layer and then, a Cu film specimen was subsequently deposited. These films will be called the not-exposed (nex-) Cu films below. In this case, the changes in \( f_s \) and \( Q_s^{-1} \) due to the additional Ta buffer layer were separately calibrated.

The X-ray diffraction (XRD) measurement was performed with the Cu Kα radiation, where the scattering vector was normal to the flat surface of the Cu film and the reflections from Si powders put on the film surface were used as reference. As the most case, the XRD measurements were made for the Cu films after the elasticity measurements up to about 400 K in order to minimize an effect of
oxidation of the Cu surface on the elastic property. Separately, in order to investigate the crystallographic changes due to heating up to about 400 K, the XRD measurements were carried out for the Cu films in the as-prepared state and those after 400 K annealing. The surface morphologies of the Si reed substrate prior to any deposition, after deposition of a Ta buffer layer and after deposition of a Cu film were sequentially observed by STM operating at the constant current mode.

3. Results

3.1. Characterization by XRD

Fig. 3 shows examples of the XRD spectra observed for the \( \text{nex-Cu} \) film with the Cu film thickness, \( t_{\text{Cu}} \), of 60 nm and the \( \text{ex-Cu} \) film with \( t_{\text{Cu}} \) of 80 nm after 400 K annealing, where the (111) reflection from fcc Cu was predominant for both the films and the (200) reflection was also observed for the \( \text{ex-Cu} \) film. Fig. 4 shows the Cu film thickness dependence of the fractional intensity of the XRD (200) reflection, \( \frac{I_{\text{200}}}{I_{\text{111}} + I_{\text{200}}} \), observed for the \( \text{nex-Cu} \) films and the \( \text{ex-Cu} \) films. For the \( \text{nex-Cu} \) films, the (200) reflection was not detected for \( t_{\text{Cu}} \) thinner than about 100 nm and showed an increase for the films thicker than about 100 nm. It is noted that such strong (111) texture is also reported for \( \text{nex-Cu} \) films sputtered on a Ta layer [3,15,16]. For the \( \text{ex-Cu} \) films, the (200) reflection was observed for the whole thickness range between 20 and 1000 nm, where \( \frac{I_{\text{200}}}{I_{\text{111}} + I_{\text{200}}} \) was about 0.2 and slightly smaller than 0.3 expected for the random texture. The mean grain size, \( D_{\text{111}} \), of the Cu films was estimated from the peak width of the (111) reflection [17]. It is not shown here but for the \( \text{nex-Cu} \) films, \( D_{\text{111}} \) was almost the same to the film thickness at around \( t_{\text{Cu}} \) of 10 nm and gradually increased with increasing \( t_{\text{Cu}} \) to about 100 nm at \( t_{\text{Cu}} \) of 1000 nm. \( D_{\text{111}} \) after 400 K annealing remained almost unchanged for the \( \text{nex-Cu} \) films thinner than 100 nm and showed a slight increase for \( t_{\text{Cu}} \) thicker than 100 nm. The film thickness dependence of \( D_{\text{111}} \) observed for the \( \text{ex-Cu} \) films was very similar to that mentioned for the \( \text{nex-Cu} \) films thinner than 100 nm but showed saturation at about 50 nm for the \( \text{ex-Cu} \) films thicker than about 100 nm (not shown here).
Fig. 5 (a) and (b) show the (111) plane spacing, $d_{111}$, normal to the flat surface of the $nex$-Cu films and the $ex$-Cu films, respectively, where $d_{111}$ was determined from the (111) reflection. For the $nex$-Cu films, $d_{111}$ observed for $t_{Cu}$ thinner than about 20 nm was about 0.3% shorter than $d_{111}$ of bulk Cu. With increasing $t_{Cu}$, $d_{111}$ showed an increase followed by saturation at $d_{111}$ slightly longer than the bulk Cu value. For the $ex$-Cu films, $d_{111}$ observed for the films thinner than 30 nm was about 0.4% shorter than $d_{111}$ of bulk Cu. With increasing $t_{Cu}$, $d_{111}$ showed an increase followed by saturation at $d_{111}$ slightly longer than the bulk Cu value. These changes in $d_{111}$ may be associated with the internal tensile stress, $\sigma_{i//}$, of the Cu film along the substrate as indicated in the right-side scale of Fig. 5(a) and (b) [18]. In Fig. 5(a) and (b), the curves 1 and 2 are tentatively fitted to the $d_{111}$ vs. $t_{Cu}$ data, respectively. The thickness dependence of $d_{111}$ observed for both the $nex$-Cu films and the $ex$-Cu films showed inflection near below 100 nm. The changes in $d_{111}$ or $\sigma_{i//}$ observed after annealing at 400 K may be due to the grain growth [19].

3.2. Elasticity measurements

Fig. 6(a) and (b) show the thickness dependence of $E_f$ observed for the $nex$-Cu films and the $ex$-Cu films, respectively. For the $nex$-Cu films, the error bars shown in Fig. 6(a) were mainly due to the experimental error for the calibration of $f_s$ associated with the additional Ta buffer.
layer, which were separately made (see Fig. 2). Such error decreased with increasing Cu film thickness. For the ex-Cu films, the experimental error for the calibration of $f_j$ was less than the size of symbols shown in Fig. 6. In Fig. 6(a) and (b), the solid curves 1 and 2 show the theoretical values of $E_x$, $E_{fth}$, estimated for the nex-Cu films and the ex-Cu films, respectively, where the crystallographic texture shown in Fig. 4 and $\sigma_{||}$ shown in Fig. 5 were taken into account. (See [10] for the evaluation of the Young’s modulus in nanostructured thin films but for Al films.) For the curve 1 in Fig. 6(a), the slight decrease in the theoretical $E_x$ with decreasing $t_{Cu}$ below 100 nm was associated with an increase in $\sigma_{||}$ and that with increasing $t_{Cu}$ above 100 nm was mainly due to an increase in $I_{200}/(I_{111} + I_{200})$. For the curve 2 in Fig. 6(b), the thickness dependence of $E_{fth}$ was mainly associated with a change in $\sigma_{||}$. $E_{fth}$ for the ex-Cu films (the curve 2) was, in general, lower than that for the nex-Cu films (the curve 1) because $I_{200}/(I_{111} + I_{200})$ observed for the ex-Cu films was larger than that for the nex-Cu films. For the nex-Cu films shown in Fig. 6(a), $E_x$ showed a gradual decrease from $E_{fth}$ (the curve 1) for $t_{Cu}$ below 100 nm. For the ex-Cu films shown in Fig. 6(b), $E_x$ showed a local decrease near below 100 nm as shown by the bold dashed curve 3. These decreases in $E_x$ will be discussed in terms of GBAP below.

Fig. 7 shows the $Q^{-1}$ spectra observed for the Si reed-substrate with the 30 nm Ta buffer layer alone and the Si reed-substrate with the ex-Cu ($t_{Cu} = 70$ nm) film. For the Si reed-substrate with the Ta buffer layer alone, the change in $Q^{-1}$ spectra due to 400 K annealing was very small. Referring to the $Q^{-1}$ spectra observed for the Si reed-substrate with the Ta buffer layer alone, those for the Si reed-substrate with the ex-Cu ($t_{Cu} = 70$ nm) film showed the strong increase above 200 K. The strong increase in $Q^{-1}$ above 200 K was commonly observed in nanostructured fcc metals, Al[8,10], Cu[20], Ag[11], Au[21], and attributed to GBAP in nanostructured fcc metals. $Q^{-1}_{f}$ was determined from $\Delta Q^{-1}$ shown in Fig. 7 and Eq. (2), where $Q^{-1}_{f}$ at 300 K was mainly associated with GBAP in the present Cu films. As seen in Fig. 7, $Q^{-1}_{f}$ in the as-prepared Cu film showed a decrease above about 340 K, indicating that modification of the grain boundaries took place. From $\Delta Q^{-1}_{f,400 K}$ shown in Fig. 7 and Eq. (2), the decrease in $Q^{-1}_{f}$ at 300 K after 400 K annealing, $\Delta Q^{-1}_{f,400 K}$, was also determined.

Fig. 8(a) shows $Q^{-1}_{f}$ at 300 K in the as-prepared state observed for the nex-Cu films and the ex-Cu films. $Q^{-1}_{f}$ at 300 K was the order of $10^{-3}$ and similar to that observed for GBAP commonly in nanostructured fcc metals. In the present thickness range shown in Fig. 8(a), $Q^{-1}_{f}$ at 300 K observed for the nex-Cu films remained almost constant and $Q^{-1}_{f}$ at 300 K for the ex-Cu films showed a small local minimum near below 100 nm. Fig. 8(b) shows the decrease in $Q^{-1}_{f}$ observed after annealing at 400 K, where the normalized value, $\Delta Q^{-1}_{f,400 K}/Q^{-1}_{f}$, is plotted. $\Delta Q^{-1}_{f,400 K}/Q^{-1}_{f}$ showed a local minimum near below 100 nm for the nex-Cu films and a local maximum near below 100 nm for the ex-Cu films, respectively. These changes in $\Delta Q^{-1}_{f,400 K}/Q^{-1}_{f}$ observed near below 100 nm will be discussed in terms of GBAP below.

![Fig. 7](image7.png)

Fig. 7. Examples of the internal friction, $Q^{-1}$, spectra observed for the ex-Cu film with the Cu film thickness of 70 nm (1) before and (2) after about 400 K annealing and those for the Si reed-substrate with the Ta buffer layer (30 nm thick) alone (1') before and (2') after annealing, respectively.

![Fig. 8](image8.png)

Fig. 8. (a) The Cu film thickness dependence of the internal friction, $Q^{-1}_{f}$, of a Cu film in the nex-Cu films and the ex-Cu films, where the $Q^{-1}_{f}$ data observed at 300 K are plotted. (b) The Cu film thickness dependence of a decrease in $Q^{-1}_{f}$ observed at 300 K after 400 K annealing, $\Delta Q^{-1}_{f,400 K}$, where the normalized data, $\Delta Q^{-1}_{f,400 K}/Q^{-1}_{f}$, are plotted.
3.3. STM observation

Fig. 9(a)–(c) show the STM surface topographs observed for the \(\text{nex-Cu} \) films with \( t_{\text{Cu}} \) of 20, 80 and 150 nm, respectively. It is not shown here but in the STM images of the Si reed-substrate just after chemical etching, pyramidal dips with \( \sim 100 \) nm square and \( \sim 40 \) nm deep due to the chemical etching were observed. The similar pyramidal dips were also observed after the deposition of a 30 nm Ta buffer film (not shown here). It can be seen in Fig. 9(a) that ultrafine crystallites with \( \sim 10 \) nm in diameter aggregate on the Si surface. The height difference in the topograph of Fig. 9(a) is about 45 nm in maximum which is close to pyramidal dips observed for the Si reed-substrates after chemical etching. In order to minimize the effect of the built-in surface morphology of a Si reed-substrate on the characterization of the surface morphology of a specimen Cu film, the root mean square for the surface roughness (the RMS roughness) was estimated avoiding pyramidal dips. Then the RMS roughness of a specimen Cu film was investigated referring to that of a Si reed-substrate just after chemical etching and that observed after the deposition of a 30 nm Ta buffer film (the built-in RMS roughness). Fig. 10 shows the built-in RMS roughness data and the RMS roughness data observed for the \(\text{nex-Cu} \) films and the \(\text{ex-Cu} \) films. For \( t_{\text{Cu}} \) below 10 nm, the RMS roughness of the \(\text{nex-Cu} \) films was almost the same as the built-in RMS roughness, suggesting that the \(\text{nex-Cu} \) films grew in the FM mode in the thickness range below 10 nm. In contrast, the RMS roughness of the \(\text{nex-Cu} \) film 20 nm thick was larger than the built-in RMS roughness, suggesting that the \(\text{nex-Cu} \) film 20 nm thick grew in the SK mode. The surface morphology observed in Fig. 9(a) supports this point of view. In Fig. 10, the RMS roughness observed for the \(\text{nex-Cu} \) films shows a local maximum near below 100 nm in Cu film thickness. The STM surface topographies seen in Fig. 9(b) and (c) show that the local maximum for the RMS roughness near below 100 nm in Cu film thickness is associated with the RMS roughness of the Cu films. In Fig. 10, the RMS roughness observed for the \(\text{ex-Cu} \) films appears to show a local maximum near below 100 nm in Cu film thickness too.

Fig. 9. STM surface images observed for the \(\text{nex-Cu} \) film with (a) \( t_{\text{Cu}} = 20 \) nm, (b) 80 nm and (c) 150 nm. The \( X \) and \( Y \) scales are 1000 nm and the \( Z \) scale is about 50 nm.

Fig. 10. The Cu film thickness dependence of the RMS roughness estimated from the STM surface image.
4. Discussion

4.1. Not exposed (nex-) Cu films

As mentioned in Fig. 6(a), referring to $E_{f\text{th}}$ depicted by the curve 1, $E_f$ observed for the nex-Cu films decreased with decreasing $t_{Cu}$ below 100 nm, e.g. by about 20% at $t_{Cu}$ of 10 nm. In other words, the anelastic strain associated with GBAP increased with decreasing $t_{Cu}$ below 100 nm, e.g. to about 25% of the constituent elastic strain, $\varepsilon_e$. On the other hand, $Q_f^{-1}$ at 300 K shown in Fig. 8(a) remained almost constant at the order of $10^{-3}$ in the present thickness range, indicating that the constituent dynamic anelastic strain associated with GBAP, $\varepsilon_{da,GB}$, was the order of 0.1% of $\varepsilon_e$ in the present thickness range. Then it is suggested that the constituent static anelastic strain associated with GBAP, $\varepsilon_{sa,GB}$, was responsible for the deviatory decrease observed in the nex-Cu films for $t_{Cu}$ below 100 nm. It is reported for nanocrystalline Au [22] that $\varepsilon_{sa,GB}$ starts to increase strongly when the applied stress increases beyond the threshold stress of a few MPa. In the nex-Cu films, with decreasing $t_{Cu}$ below 100 nm, the mean grain size decreased [14] and $\sigma_{il}$ increased (Fig. 5(a)), respectively. The strong increase in $\varepsilon_{sa,GB}$ in the nex-Cu films may be associated with the decrease in the mean grain size and the increase in $\sigma_{il}$. It is noted that the decrease in $E_f$ in the thickness range below a few tens nm was commonly observed in Al and Al alloy films [10] and Ag films [12].

As already mentioned, the RMS roughness data (Fig. 10) suggested that the nex-Cu films thicker than 20 nm grew in the SK mode. On the other hand, the inflection in the thickness dependence of $d_{111}$ (Fig. 5(a)), the local maximum in the thickness dependence of the RMS roughness (Fig. 10) and the local minimum in the thickness dependence of $\Delta Q_f^{-1}(400 K)/Q_f^{-1}$ (Fig. 8(b)) were found near below 100 nm. The $\Delta Q_f^{-1}(400 K)/Q_f^{-1}$ data indicate that the thermal stability of the GBs was higher in the nex-Cu films with $t_{Cu}$ near below 100 nm than in the nex-Cu films with $t_{Cu}$ far below 100 nm or above 100 nm. The present results indicate that the properties of GBs in already deposited Cu film were modified by additional deposition of the same metal (the capping effect, hereafter) and the anomalous capping effects took place near below 100 nm in $t_{Cu}$.

4.2. Exposed (ex-) Cu films

As shown in Fig. 6(b), the thickness dependence of $E_f$ observed for the ex-Cu films was explained by the curve 2 for $E_{f\text{th}}$ except that $E_f$ showed a local decrease near below 100 nm (the curve 3). In the thickness range below a few tens nm, the decrease in $E_f$ was commonly observed in the nex-Cu films (Fig. 6(a)), Al and Al alloy films [10] and Ag films [12], but such decrease in $E_f$ in the thickness range below a few tens nm was not observed for the ex-Cu films. For the ex-Cu films with $t_{Cu}$ below a few tens nm, $\sigma_{il}$ was very high (Fig. 5(b)) and the thermal stability of the GBs was high too (Fig. 8). As already mentioned, the deposition of the ex-Cu films is expected to start in the VM mode followed by the SK mode. Then it is suggested that the properties of GBs in the thickness range below a few tens nm were of those in bulk metal but not of those in nanocrystalline metal [22] and then the properties of GBs changed to those in nanocrystalline metal with increasing $t_{Cu}$. The decrease in $Q_f^{-1}$ (Fig. 8(a)) with decreasing $t_{Cu}$ below 100 nm observed for the ex-Cu films supports this point of view. The local decrease in $E_f$ for $t_{Cu}$ near below 100 nm (the curve 3 in Fig. 6(b)) or the increase in $\varepsilon_{sa,GB}$ may be associated with the transition of the properties of GBs. The local maxima of $\Delta Q_f^{-1}(400 K)/Q_f^{-1}$ (Fig. 8(b)) and the RMS roughness (Fig. 10) found for $t_{Cu}$ near below 100 nm may be associated with the transition of the properties of GBs too. The present results also indicate that for the ex-Cu films, the properties of GBs in already deposited Cu film were modified by capping of the same metal.

5. Conclusion

The Young’s modulus, $E_f$, the internal friction, $Q_f^{-1}$, and the root mean square of the surface roughness (RMS roughness) were investigated for the Cu films deposited on a clean surface of Ta (the nex-Cu films) and those on a Ta with native oxide (the ex-Cu films), where the sputter-deposition of Cu and Ta was made at ambient temperature.

For the nex-Cu films, $E_f$ showed a deviatory decrease from $E_{f\text{th}}$ due to the static anelastic strain in grain boundary regions (GBs) for the Cu film thickness, $t_{Cu}$, below 100 nm, where $E_{f\text{th}}$ is the theoretical value of $E_f$ estimated by taking into account the crystallographic texture and the internal stress. The local maximum of the RMS roughness and the local minimum of the recovery in the internal friction by 400 K annealing, $\Delta Q_f^{-1}(400 K)/Q_f^{-1}$, were found for $t_{Cu}$ near below 100 nm, indicating that the properties of GBs in already deposited Cu film were modified by additional deposition of the same metal (the capping effect) and the anomalous capping effects took place near below 100 nm in $t_{Cu}$.

For the ex-Cu films, $E_f$ showed good agreement with $E_{f\text{th}}$ except for a local decrease in $E_f$ found for $t_{Cu}$ near below 100 nm. The local maxima of $\Delta Q_f^{-1}(400 K)/Q_f^{-1}$ and the RMS roughness were also found for $t_{Cu}$ near below 100 nm. It is suggested that the properties of GBs were of those in bulk metal for $t_{Cu}$ below a few tens of nanometer and then changed to those in nanocrystalline metal with increasing $t_{Cu}$. The capping effect was also observed for the ex-Cu films.

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