Internal Friction Study of Microplasticity of Aluminum Thin Films on Silicon Substrates

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Internal friction in aluminum thin films 0.2 to 2.0 μm thick on silicon substrates has been investigated between 180 and 360 K as a function of strain amplitude by means of a free-decay method of flexural vibration. According to the constitutive equation, the internal friction in the film alone can be evaluated separately from the data on the film/substrate composite. The amplitude-dependent part of internal friction in aluminum films is found in the strain range approximately two orders of magnitude higher than that for bulk aluminum. On the basis of the microplasticity theory, the amplitude-dependent internal friction can be converted into the plastic strain as a function of the effective stress on dislocation motion. The mechanical responses thus obtained for aluminum films show that the plastic strain of the order of 10^{-9} increases nonlinearly with increasing stress. These curves tend to shift to a higher stress with decreasing film thickness and also with decreasing temperature, both indicating a suppression of the microplastic deformation. At all temperatures examined, the microflow stress at a constant level of the plastic strain varies inversely with the film thickness, which qualitatively agrees with the variation in macroscopic yield stress.

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I. Introduction

The mechanical properties of thin films are very different from those of corresponding bulk materials. It is known that many metallic films exhibit yield stresses much higher than their bulk values, which are partly related to the high densities of grown-in defects that pin the dislocation motion and/or the lack of operating dislocation sources⁽¹⁾. Because thin films are usually bonded to rigid substrates, the study of their mechanical behavior requires the development and use of nontraditional testing methods, as reviewed by Nix⁽²⁾. For such thin-layer materials, two common techniques have so far been used: a substrate curvature technique⁽³⁾⁻⁽⁵⁾ for the stress measurements and a submicron indentation technique⁽⁴⁾ for the hardness measurements. However, both methods have an inherent limitation for the absolute evaluation of stress-strain relationships, which can be obtained by uniaxial tensile tests of free-standing films detached from a substrate⁽⁶⁾. It should also be pointed out that, according to Doerner *et al.*⁽⁴⁾, the strength of the film on a substrate differs from that of the film alone.

In recent years, internal friction measurements have successfully been applied to the study of thin-layer materials. Since the thin-layer study has been introduced and pursued by Berry and Pritchett⁽⁷⁾⁽⁸⁾, the temperature dependence of internal friction has been investigated to date⁽⁹⁾⁻⁽¹³⁾: *e.g.*, point-defect reorientation and grain-boundary sliding. A substantially new area for the thin-layer study is the amplitude dependence of internal friction, which is caused by microplasticity owing to dis-

location motion in the pre-yield range. As shown in our previous study⁽¹⁴⁾, the analysis of the amplitude-dependent internal friction in thin-layer materials requires two theoretical procedures. First it is necessary to calculate the internal friction in the film separately from the data on the composite system. This relationship is given in terms of the constitutive equation of internal friction⁽¹⁵⁾. Secondly it is necessary to convert the amplitude-dependent internal friction in the film into typical mechanical responses such as stress-strain curves in the microplastic deformation. The analysis is based on the microplasticity theory, which has first been proposed by Asano⁽¹⁶⁾ and later developed by Nishino and Asano⁽¹⁷⁾⁽¹⁸⁾. This is a phenomenological approach and can therefore be applied to various internal friction data not only on metals and alloys⁽¹⁹⁾⁽²⁰⁾, but also on ceramics⁽²¹⁾⁽²²⁾ and thin-film materials⁽¹⁴⁾⁽²³⁾ where precise mechanisms are still unknown for the amplitude dependence.

In the present study, the strain-amplitude dependence of internal friction in aluminum thin films on silicon substrates has been examined at various temperatures from the analytical point of view. A typical example of the experimental data is presented in Section II. The results obtained show that the amplitude-dependent internal friction in aluminum films differs significantly from that of bulk aluminum. Procedures for the data analysis are briefly summarized in Section III, and we demonstrate that the present analytical approach is applicable for evaluation of the plastic strain in aluminum films as a function of the effective stress on dislocation motion, and could be comparable with tensile testing of free-standing films. Finally, the variations in the microflow stress with the film thickness and also with the temperature are obtained in Section IV to clarify the mechanical

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properties of aluminum films in the microplastic range.

II. Experimental

1. Sample preparation

Aluminum thin films were deposited onto one face of (100)-oriented silicon wafers 0.5 mm thick by magnetron sputtering at room temperature. Prior to the film deposition, the wafers were cleaned in a HF solution. The base pressure of the sputtering system was 2×10^{-5} Pa, and the deposition rate was about 8 nm/s. Polycrystalline films with five different film thicknesses ranging from 0.2 to $2.0 \mu\text{m}$ were deposited using an aluminum target with a purity of 99.999%. Transmission electron microscopy revealed that the films consist of a single layer of grains with the boundaries perpendicular to the film surface and that the average grain size in the as-deposited state, as determined by the linear-intercept method, is roughly the same as the film thickness. After the film deposition, the wafers were cut with a diamond wheel saw into a rectangular shape of 10 mm in width and 80 mm in length.

2. Internal friction measurements

Internal friction in the composite sample, consisting of an aluminum film and a silicon substrate, was measured by means of a free-decay method of flexural vibration with both ends free and at a frequency around 700 Hz in the fundamental resonant mode. The measurements were performed in vacuum at temperatures ranging from 180 to 360 K. After a steady-state vibration for more than 60 s at a desired temperature, the driving signal was turned off and then a free-decay curve was measured using a high-speed level recorder, where the logarithm of the vibration amplitude is recorded against time. As a measure of internal friction, the logarithmic decrement was determined from a slope of the tangent to the smooth envelope of the free decay curve. The internal friction was measured as a function of the maximum strain amplitude that was calculated from the displacement amplitude in the middle of the composite sample.

Because of a nondestructive nature of internal friction measurements, the testing temperatures can be changed successively for a given sample. As a typical example, Fig. 1 shows the strain-amplitude dependence of internal friction in the composite samples with aluminum films of thickness $t=0.5$ to $2.0 \mu\text{m}$, all of which were measured at 240 K. The internal friction in the silicon substrate, as shown by the full circles, remained constant, independent of the maximum strain amplitude ϵ_{max} . In contrast, the internal friction in the composite samples, δ_{av} , always increased with increasing film thickness and became sensitive to ϵ_{max} : the amplitude dependence is stronger for a larger film thickness. These $\delta_{\text{av}}(\epsilon_{\text{max}})$ curves were reproducible when repeatedly measured under the same conditions. Such reproducibility suggests that the generation and multiplication of dislocations is unlikely to occur in the films during the measurements.

The film-thickness effect on the amplitude dependence

Fig. 1 The measured internal friction δ_{av} at 240 K plotted as a function of the maximum strain amplitude ϵ_{max} in the composite samples with aluminum films of thickness $t=0.5$ to $2.0 \mu\text{m}$. The data on the silicon substrate is shown by the full circles.

shown in Fig. 1 is generally the same as our previous result⁽¹⁴⁾ obtained at room temperature. On further cooling, the level of internal friction was considerably reduced: e.g., at 180 K, the value for the silicon substrate was as low as 3×10^{-5} , indicating that the thermoelastic damping is included in the measured internal friction. On the other hand, once heated to a temperature above 380 K, the amplitude dependence obtained at room temperature was no longer reproducible and much more suppressed than that before the heating. This is due to the alloying effect, presumably arising from a dissolution of silicon into the aluminum films during heating. In order to avoid this solid-state reaction at the interface, the internal friction measurements were made in the temperature range below 380 K.

III. Data Analysis

1. Evaluation of internal friction in thin films

When the film of thickness t is deposited onto one face of the substrate of thickness d , the measured internal friction δ_{av} in the composite sample is a weighted average of the internal friction in the substrate and the film, δ_s and δ_f , respectively⁽⁷⁾. The relationship among them is given by the constitutive equation of internal friction, and analytical expressions have been derived systematically for almost all vibration modes⁽¹⁵⁾. Especially for the case of free-free flexural vibration, the constitutive equation is

$$\delta_{\text{av}} = \delta_s + \frac{3t}{d} \frac{E_f}{E_s} \delta_f, \quad (1)$$

where E_s and E_f denote the Young's moduli of the substrate and the film, respectively. This relation for the free-

free beam is in accordance with that obtained by Berry and Pritchett⁽⁷⁾ for a clamped-free beam in a vibrating-reed method. Since the internal friction in the silicon substrate is much lower than that of aluminum, one can accurately evaluate the value of δ_f in the aluminum film. According to Mizubayashi *et al.*⁽¹⁰⁾, the Young's modulus of the aluminum film, E_f , for thicknesses of $t > 0.1 \mu\text{m}$, can be regarded simply as that of bulk aluminum. It must further be assumed in eq. (1) that any parasitic damping would not arise from the interface between the film and the substrate, i.e., the film adheres perfectly to the substrate. The case of imperfect adhesion has been treated by Wuttig and Su⁽²⁴⁾, who take into account the degree of adhesion at the interface.

By the use of eq. (1), the value of δ_f in the film alone can be calculated from the measured internal friction, δ_{av} and δ_s , in Fig. 1. The internal friction δ_f in the aluminum films of thickness $t = 0.5$ to $2.0 \mu\text{m}$ at 240 K is shown in Fig. 2 as a function of the maximum strain amplitude ϵ_{max} . It should be remarked that the calculated values of δ_f in Fig. 2 are almost 10^2 times larger than the measured values of δ_{av} in Fig. 1. For comparison, the amplitude dependence of internal friction in bulk polycrystalline aluminum is shown by the broken curve. The magnitude of the internal friction in the films is of the same order as that in bulk aluminum. However, the amplitude-dependent part of internal friction in the films can be found only at a strain approximately two orders of magnitude higher than that for the bulk. It is also seen that the amplitude-independent part for the films decreases as the film thickness decreases, while the amplitude-dependent part shifts to a higher strain amplitude. In particular, the latter is caused by microplasticity owing to dislocation motion in

Fig. 2 The calculated internal friction δ_f at 240 K plotted as a function of the maximum strain amplitude ϵ_{max} in aluminum films of thickness $t = 0.5$ to $2.0 \mu\text{m}$. The measured internal friction in bulk aluminum is represented by the broken curve.

Fig. 3 The calculated internal friction δ_f plotted as a function of the maximum strain amplitude ϵ_{max} in aluminum film $1.0 \mu\text{m}$ thick at temperatures $T = 180$ to 360 K .

the films, as will be discussed later. It might be curious that the former appears to go away from that for the bulk as the thickness increases, but within the scope of δ_f , the result can be considered to be characteristic of thin films. The amplitude-independent part would be associated with the grain-boundary relaxation in the films, according to Prieler *et al.*⁽¹¹⁾, who have suggested that reduction in the relaxation peak with decreasing film thickness may be attributed to the influence of the substrate, which, owing to adhesion, could hinder relative displacements of neighboring grains near the interface.

The effect of temperature on the internal friction in the aluminum films has been examined by analyzing internal friction data in the same manner as described above. As an example, Fig. 3 shows the internal friction δ_f in the aluminum film $1.0 \mu\text{m}$ thick as a function of the maximum strain amplitude ϵ_{max} at temperatures $T = 180$ to 360 K . Note that a series of $\delta_f(\epsilon_{\text{max}})$ curves have been obtained from internal friction data measured on the same sample. It can be seen that the amplitude-independent part increases as the temperature is raised, and also the amplitude-dependent part appears more remarkably. As mentioned before, the temperature effect on the amplitude dependence was reproducible in this temperature range. The above results therefore indicate that the microplastic deformation in the films occurs more easily at a higher temperature. In the following analysis, we evaluate the stress-strain responses in the microplastic range from the amplitude-dependent internal friction in the films.

2. Conversion to stress-strain responses

In the case of the thin film under the free-free flexural vibration, the strain distribution is supposed to be

uniform through the thickness but inhomogeneous along the length. Internal friction data, measured in the amplitude-dependent range, should be formulated in the volume integration of local internal friction⁽²⁵⁾. When the value of δ_f is obtained as a function of the maximum strain amplitude $\varepsilon_{\max}$, the intrinsic internal friction $\delta(\varepsilon_0)$ in the film, that would be measured under the ideal condition of a homogeneous strain, is expressed in the form,

$$\delta(\varepsilon_0) = \frac{1}{2\varepsilon_0^2} \frac{d}{d\varepsilon_0} \int_0^{\varepsilon_0} \frac{\varepsilon_{\max}^3 \delta_f(\varepsilon_{\max})}{\sqrt{\varepsilon_0^2 - \varepsilon_{\max}^2}} d\varepsilon_{\max}, \quad (2)$$

where ε_0 is the strain amplitude under the homogeneous strain. Because of the thin-film approximation for the strain distribution⁽¹⁴⁾ as assumed above, this expression for the flexural vibration is formally of the same type as that for longitudinal vibration⁽²⁵⁾. The correction due to an inhomogeneous strain is important in the data analysis for evaluating absolute values of the plastic strain as a function of stress.

From the definition of internal friction, $\delta(\varepsilon_0)$ or $\delta(\sigma_0)$ is related with the total strain $\varepsilon(\sigma)$ as follows:

$$\delta(\sigma_0) = \frac{E_f}{\sigma_0^2} \oint \varepsilon(\sigma) d\sigma, \quad (3)$$

where $\sigma_0 = E_f \varepsilon_0$, and the integration is performed for one cycle of vibration, corresponding to the loop area of stress-strain hysteresis. In addition, $\varepsilon(\sigma)$ is regarded as the sum of elastic and plastic components,

$$\varepsilon(\sigma) = \frac{\sigma}{E_f} + \varepsilon_p(\sigma), \quad (4)$$

where ε_p is the plastic strain due to dislocation motion, giving $\varepsilon(\sigma)$ a nonlinear hysteretic character. Direct observations of mechanical hysteresis in the microplastic deformation⁽²⁶⁾ demonstrate that steady-state cyclic deformation results in a single closed loop with a symmetrical shape, i.e., a friction-type hysteresis as classified by Asano⁽¹⁶⁾. Such a hysteresis loop has also been observed for aluminum films under a loading-unloading condition of tensile testing⁽²⁷⁾. Then, according to the microplasticity theory⁽¹⁶⁾⁽¹⁷⁾, the analytical solution of eq. (3) together with eq. (4) is expressed as

$$\varepsilon_p(\sigma) = \frac{\sigma}{E_f} \left[\frac{1}{4} \delta(\sigma/2) + \frac{1}{2} \int_0^{\sigma/2} \frac{\delta(\sigma_0)}{\sigma_0} d\sigma_0 \right]. \quad (5)$$

Nishino and Asano⁽¹⁷⁾ have pointed out that ε_p should be defined as a function of the effective stress σ on dislocation motion, whose origin is settled at the lower corner point of the hysteresis loop under a steady-state vibration.

In order to apply the general formulae described above, the $\delta_f(\varepsilon_{\max})$ curves shown in Figs. 2 and 3 are approximated by a power function of $\varepsilon_{\max}$,

$$\delta_f(\varepsilon_{\max}) = A \varepsilon_{\max}^n + B, \quad (6)$$

where A , B and n are constants. Thus, eq. (6) is substituted into eq. (2) to give the explicit form of the intrinsic internal friction $\delta(\varepsilon_0)$,

$$\delta(\varepsilon_0) = A K(n) \varepsilon_0^n + B, \quad (7)$$

and the characteristic function $K(n)$ is expressed in the form⁽²⁵⁾,

$$K(n) = \frac{\sqrt{\pi}}{2} \Gamma\left(\frac{n+4}{2}\right) \Gamma\left(\frac{n+3}{2}\right), \quad (8)$$

where Γ represents the gamma function. When the functional form of $\delta(\varepsilon_0)$ or $\delta(\sigma_0)$ is known, substitution of the amplitude-dependent term of eq. (7) into eq. (5) yields

$$\varepsilon_p(\sigma) = \frac{A(n+2)}{2^{n+2}n} K(n) \left(\frac{\sigma}{E_f}\right)^{n+1}. \quad (9)$$

This simple expression is applicable to evaluate the stress-strain responses from the internal friction data expressed by eq. (6).

Figure 4 shows the relation between ε_p and σ at 240 K for the aluminum films of thickness $t=0.5$ to $2.0 \mu\text{m}$, where the effective stress σ is taken on the abscissa: the curves are obtained from the results in Fig. 2. The value of ε_p is only of the order of 10^{-9} and as low as 0.01% of the total strain given by eq. (4). Nevertheless, microplastic flow indeed occurs in the films and the plastic strain increases nonlinearly with increasing stress. Such a nonlinear hysteretic response in the microplastic range may have the same origin as that obtained by tensile testing of free-standing films⁽²⁷⁾. It should be noted that, although not shown in Fig. 4, the microplastic flow in bulk aluminum occurs only at a stress level about an order of magnitude lower than that for the films. There is also a general tendency for the films that the curves shift to a higher stress with decreasing film thickness, thus indicating a suppression of the microplastic flow in thinner films.

Figure 5 shows the relation between ε_p and σ for the aluminum film $1.0 \mu\text{m}$ thick at temperatures $T=180$ to 360 K : the curves are obtained from the results in Fig. 3.

Fig. 4 The plastic strain ε_p expressed as a function of stress σ in aluminum films of thickness $t=0.5$ to $2.0 \mu\text{m}$ at 240 K. The curves are obtained from the results in Fig. 2.

Fig. 5 The plastic strain ϵ_p expressed as a function of stress σ in aluminum film $1.0 \mu\text{m}$ thick at temperatures $T=180$ to 360 K . The curves are obtained from the results in Fig. 3.

It is clearly seen that the curves shift to a lower stress as the temperature is raised, which means a softening of the aluminum film at a higher temperature. A similar temperature effect on the microplastic flow has been observed for all the film thicknesses examined. Although the stress levels are approximately two orders of magnitude lower than the macroscopic yield stresses, the nonlinear hysteretic behavior of the aluminum films, as shown in Figs. 4 and 5, could be caused by dislocation motion under a steady-state vibration in the microplastic range.

IV. Discussion

As demonstrated in Section III, it is possible to evaluate an absolute form of stress-strain responses in the microplastic deformation from the amplitude-dependent internal friction. This analytical approach is applicable to the internal friction data on both bulk and thin-film materials. In the case of bulk materials, similar analyses⁽²⁸⁾⁻⁽³⁰⁾ have been performed for the amplitude-dependent internal friction, but only a cross-cut stress, i.e., a relative value of stress amplitude is provided as a microyield stress. In this sense, the present analytical approach of internal friction substantially differs from other existing methods. In the following, we will make clear the variation in the film strengths by defining the microflow stresses at a certain value of the plastic strain in the stress-strain responses.

1. Temperature dependence of microflow stress

The microflow stress σ at $\epsilon_p = 1 \times 10^{-9}$, as determined from Fig. 5, is plotted against the temperature T in Fig. 6. The microflow stress should be regarded as a measure of the resistance to dislocation motion in the microplastic deformation. For the films of thickness $t=0.5$, 1.0 and $2.0 \mu\text{m}$, the respective σ values decrease considerably as the temperature is raised, and above room temperature

they become almost constant. The temperature dependence of the microflow stress in the aluminum films is qualitatively the same as that for bulk aluminum shown by the open squares. However, as compared with the bulk, the variation in the stress levels is more remarkable for the films especially of a smaller thickness. Since the generation and multiplication of dislocations would not occur during the measurements and the motion of dislocations occurs only inside the films, the large increase in the stress levels on cooling may be attributed to the constraining effect of the substrate to which the film is bonded.

According to substrate curvature measurements⁽³⁾, the stress in aluminum films is considered to be almost zero in the as-deposited state when a substrate temperature remains close to room temperature during the deposition. On cooling during the internal friction measurements, due to the difference in thermal expansion of the aluminum film and the silicon substrate, the film is supposed to be in a state of biaxial tension below room temperature, and a significant portion of the thermally-induced stress may be accommodated by elastic deformation in the film. As far as the stress relaxation would not occur during cooling, the strengthening due to the thermal expansion mismatch is expected not to depend on the film thickness. In fact, when the microflow stresses for different film thicknesses in Fig. 6 are normalized by setting the respective constant values above 300 K to unity, all the temperature trends absolutely coincide with each other. In contrast, the microflow stresses normalized as above for the bulk are still lower than those of the films at all temperatures below 300 K . Therefore, the large increase in the microflow stress on cooling could be caused by the presence of biaxial tension in the film owing to the thermal expansion mismatch.

Fig. 6 Temperature dependence of the microflow stress σ at $\epsilon_p = 1 \times 10^{-9}$ in aluminum films of thickness $t=0.5$, 1.0 and $2.0 \mu\text{m}$. The σ values for bulk aluminum are shown by the open squares.

2. Thickness effect on film strength

Figure 7 shows the relation between the microflow stress σ at $\epsilon_p = 1 \times 10^{-9}$ and the film thickness t at $T = 180, 240, 300$ and 360 K: here, σ is plotted against $1/t$. Also, the σ values for bulk aluminum are indicated by the open squares on the axis of $1/t = 0$. The solid lines are the best fit through the data points including the bulk results, and excellent linearity is observed at all temperatures examined. As a result, the microflow stress is inversely proportional to the film thickness, and hence the thickness effect on the film strength σ_{film} in the microplastic deformation is simply expressed as

$$\sigma_{\text{film}} = \sigma_{\text{bulk}} \left(1 + \frac{C}{t} \right), \quad (10)$$

where σ_{bulk} is the microflow stress for the bulk, and C is a constant that depends on the temperature. A higher C at a lower temperature, as seen in Fig. 7, is due to the effect of biaxial tension in the film, arising from the thermal expansion mismatch as mentioned above. It should be pointed out that the thickness effect on the film strength in the microplastic range qualitatively agrees with the variation in the macroscopic yield stresses as obtained from substrate curvature measurements⁽³⁾⁽⁵⁾. Although the respective stress levels are quite different from each other, the film strength always varies inversely with the film thickness, both in the macroscopic yield stresses and in the microflow stresses. This qualitative agreement is surprising if one remembers that microplasticity is controlled only by the movement of dislocations under a steady-state vibration, without accompanying the generation and multiplication of dislocations as in macroplastic-

ity. In the microplastic range studied here, dislocation motion is impeded by friction owing to randomly dispersed obstacles, in addition to intrinsic lattice resistance.

As reviewed by Doerner and Nix⁽³¹⁾, the increased strength of thin films is generally ascribed to the unique microstructures with the grain size as small as the film thickness, combined with the constraining effect of a rigid substrate. Also, the Hall-Petch relationship has been validated for aluminum films⁽⁴⁾⁽³²⁾. In their study using a substrate curvature technique, Venkatraman and Bravman⁽⁵⁾ have separated the components of strengthening due to the film thickness and the grain size by assuming that the flow stress of the film is simply the sum of these two components. On the other hand, in the microplastic deformation, the grain-size strengthening is considered to be negligible because dislocation motion occurs only inside individual grains and the grain boundaries play a minor role as suggested by Lebedev⁽³³⁾. There is actually no grain-size effect on the microflow stress in bulk aluminum when the grain size is larger than several tens of micrometers, as has previously been demonstrated for bulk copper⁽³⁴⁾. In the case of the thin films, therefore, much of the film strength in the microplastic range is caused by the film-thickness effect, provided the grain size is larger or at least equal to the film thickness. We should expect that the increase in the microflow stress with decreasing film thickness is associated with the bowing of dislocation segments whose ends are fixed at the film surface as well as the film-substrate interface. Since the loop length of the dislocation becomes shorter as the film thickness decreases, the stress required to move the dislocation increases. This is consistent with eq. (10), which is also equivalent to that explained by a conventional dislocation pinning model or a force balance on the dislocation⁽³⁵⁾. The model, however, underestimates the macroscopic yield stresses measured for aluminum films⁽³⁾⁻⁽⁵⁾ by an order of magnitude, and also for lead film⁽³⁶⁾ by a factor of three to four. Although a more detailed model than eq. (10) is needed for the macroplastic deformation, the dislocation pinning mechanism described above seems to be the most reasonable one for explaining the film thickness effect in the microplastic deformation.

V. Conclusion

In our attempt to study the mechanical properties of thin films, we have examined the strain-amplitude dependence of internal friction in aluminum films 0.2 to 2.0 μm thick on silicon substrates between 180 and 360 K. Since the internal friction measurements are made on the composite sample, it is necessary first to calculate the internal friction in the film alone from the data on the composite sample, using the constitutive equation of internal friction. The amplitude-dependent part of internal friction in the aluminum films is found only at a strain approximately two orders of magnitude higher than that for bulk aluminum. Further, on the basis of the microplasticity

Fig. 7 The microflow stress σ at $\epsilon_p = 1 \times 10^{-9}$ as a function of the inverse thickness $1/t$ at temperatures $T = 180$ to 360 K. The σ values for bulk aluminum are shown by the open squares.

theory, the amplitude-dependent internal friction can be converted into the plastic strain expressed as a function of the effective stress on dislocation motion. From a practical standpoint, the present analytical approach of internal friction is applicable for non-destructive evaluation of the stress-strain responses in thin films bonded to a substrate, and is directly comparable with uniaxial tensile testing of free-standing films detached from a substrate.

The microplastic stress-strain responses thus obtained for the aluminum films show that the curves shift to a higher stress as the film thickness decreases and also as the temperature decreases. The microflow stress at the plastic strain of 1×10^{-9} increases markedly on cooling below room temperature, because of an additional thermal contribution owing to the difference in thermal expansion of the aluminum film and the silicon substrate. When the grain size is larger than the film thickness, the microflow stress is inversely proportional to the film thickness, qualitatively in agreement with that of the macroscopic yield strengths, although the stress levels are considerably different from each other. The thickness effect on the film strength in the microplastic range could be explained by the bowing of dislocation segments whose ends are fixed at the film surface and at the film-substrate interface. A consistent understanding of the relationship between microplasticity and macroplasticity is important to clarify deformation processes responsible for the high strengths of thin films on substrates.

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