MECHANICAL BEHAVIOR OF THIN FILMS

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KEY WORDS: stress, strain, mechanical testing, mechanical properties

ABSTRACT

The mechanical properties of thin films differ significantly from those of bulk materials due, largely, to the effects of interfaces, microstructure, and thick underlying substrates. Specialized experimental techniques have been developed to determine mechanical properties and the strain and stress states of both free-standing films and films bonded to substrates. In this review, recent innovations in measurement of thin film properties are considered. Included are discussions of nanoindentation, bulge testing, substrate curvature, X-ray diffraction, micro-Raman spectroscopy, and electron diffraction contrast imaging. Selected recent advances in the understanding of continuous thin film and patterned thin film mechanical behavior are also presented. Examples include low-temperature and high-temperature plastic flow in Al and Cu films, strain energy effects on film morphology, hardness enhancement in multilayers, and stress in passivated Al lines.

INTRODUCTION

Although certain thin film materials have been developed specifically for mechanical applications such as wear-resistant coatings, the primary uses of thin films have exploited their electronic, magnetic, and optical properties. Most of these applications have not included mechanical loads as part of their design and, as a result, the mechanical characteris-
tics of the films have often been ignored. However, because even films without direct structural applications are often subjected to large mechanical stresses resulting from specific deposition conditions, temperature changes during processing and in service, and from the presence of additional protective coatings, it is now recognized that an understanding of mechanical properties is essential for improving the reliability and lifetime of thin films. Furthermore, in the microelectronics and data storage industries, thin film mechanical behavior is becoming increasingly important for the basic function of certain thin film devices (e.g. films fashioned into microsensors and actuators or films for which crucial magnetic or electrical properties are dependent on the state of stress). Thus it has become progressively more important to understand the mechanical properties of thin film materials both for extensive applications as nonstructural elements and for use as tool coatings and microelectromechanical devices.

An understanding of the connections between processing, microstructure, and material deformation characteristics is the key to predicting and modifying the mechanical properties of thin films. Much of the early work in the mechanical properties of thin films was concerned with stresses and failures in integrated circuit structures, particularly in interconnect lines and dielectrics. Hoffman (1) and Campbell (2) have summarized much of this work. More recently, the examination of intrinsic stresses in microelectronic films has progressed to the study of the mechanical properties of microelectronic structures and a broad range of thin film materials. Much of this work through the late 1980s has been summarized by Nix (3), and Alexopoulos & O'Sullivan (4). Detailed reviews of nanoindentation (5), micromechanical testing (6), and X-ray diffraction techniques (7) are also recommended.

The recent work in thin film mechanical properties has been driven by the need for a reliable predictive capability of processing and properties. This capability is crucial for determining materials-related opportunities and limitations in design of future thin film devices. The knowledge required for accurate prediction of properties must be based on inexpensive, accurate tests of properties as a function of processing. This, in turn, has driven efforts to improve established measurement techniques and, where necessary, to develop new methods. Publication in this area has increased significantly since the late 1980s, so this review can in no way be comprehensive. As a result, developments from the past five years are emphasized to highlight selected recent advances in technique development and in understanding of thin film behavior.

A number of standard testing methods now exist, but work on these
techniques continues, to better characterize their limitations and to extend their applicability. We identify some of the significant improvements that have been made to existing methods. Detailed descriptions of the techniques are not presented because they are available in the referenced literature. Also, we examine a number of intriguing new techniques that have been recently introduced. This discussion concentrates on techniques intended to explore mechanical properties at submicrometer resolution, many of which are in the early stages of development. Finally, a few recently developed advances in the understanding of thin film mechanical behavior are presented as an introduction to the existing literature and ongoing challenges.

EXPERIMENTAL TECHNIQUES FOR MEASURING THIN FILM PROPERTIES

Review of Thin Film Mechanical Properties

It is now widely understood that mechanical properties of thin film materials may not be indiscriminately extrapolated from bulk material properties. Bulk testing methods are generally difficult or impossible to apply directly to thin films, necessitating the use of new methods that actually measure different quantities. For instance, uniaxial testing of films is difficult, so the biaxial modulus, rather than Young's uniaxial modulus, is more commonly measured (8–10). With thin films, particular care must be taken not to introduce instabilities in the form of tearing or buckling when these behaviors may compromise the intended test.

Furthermore, bulk testing generally provides average properties over a large section of a material, whereas thin film measurements are often taken over a very small volume. Thin films are also much more sensitive to the influences of interfaces and adjoining materials than are bulk materials. It is common to see different behavior in a free-standing film and an identical film attached to a substrate (3).

Microstructural changes induced in thin films during testing are often very different from those seen in the bulk. Grain structure, dislocation density, and vacancy concentration differences are often significant, invalidating many assumptions made about plastic deformation that are appropriate for bulk materials. Mechanical property calculations for many measurement techniques are based on deformation models that assume elastic isotropy. This is usually valid in the plane of a film for small-grained, polycrystalline materials or for many bulk polycrystalline materials. Epitaxial, highly textured, or finely patterned films with lateral dimensions on the order of the grain size are more accurately
described by anisotropic elasticity, complicating the process of deducing mechanical properties from recorded behavior. Despite these differences, however, the general approach to studying thin film mechanical properties remains similar to that employed in studying bulk properties. Elastic, plastic, and time-dependent thin film properties can all be determined with current thin film testing methods.

The most basic requirements for any mechanical test are a controllable change in film strain and a method for measurement of the corresponding change in stress. An applied elastic film strain can be used to investigate elastic moduli. If plastic strains are applied, the measured response will indicate the flow stress of the film. Mechanical loading, reminiscent of bulk testing, may be used to induce stresses as long as care is taken to isolate the effects of the testing apparatus from actual film properties. Two examples of this are nanoindentation and bulge testing. Recent advances in both techniques have allowed repeatable, straightforward measurement of elastic and plastic properties of volumes ranging from nanometers to several square millimeters. Because thin films are frequently bonded to substrates, thermal expansion coefficient mismatch is often used to induce controlled strains, typically providing information on elastic, plastic, and time-dependent properties. Recent extensions of the widely used substrate curvature technique have attempted to apply this method to patterned films. Probing of thermally strained patterned structures at the submicrometer scale has also been attempted through the use of X-ray, optical, and electron beams.

**Nanoindentation**

Nanoindentation is one of the most popular techniques for measuring the mechanical properties of thin films. In a typical indentation experiment, a pyramidal or spherical diamond indenter is driven into a material while applied load and displacement are continuously monitored. Interest in the indentation test as a means of studying thin film mechanical properties has grown only recently with the development of very low-load, depth-sensing indentation instruments (11–13). These instruments allow one to make indentations as shallow as a few nanometers. The stiffness of the contact between the indenter tip and the material is then relatively small and can be measured accurately. Using an expression for the contact stiffness of an axisymmetric indenter on an isotropic halfspace (14, 15), the indentation modulus of the film $E/(1-\nu^2)$ can be determined. If the indenter is not axisymmetric, a correction factor is used to account for the shape of the indenter (16, 17). The hardness of the film is taken to be the applied load divided by the
projected contact area between indenter and material. Several models have been developed in order to estimate the contact area from the elastic unloading segment in the load-displacement curves (18, 19). In the model by Doerner & Nix, the indenter is modeled as a flat punch on a half space, whereas in the model by Oliver & Pharr, the indenter is modeled as a paraboloid. The Oliver-Pharr model describes the unloading curve better, and it results in contact areas that are slightly larger than those determined from the Doerner-Nix model.

Recently, the influence of elastic anisotropy on the indentation modulus has been investigated (17, 20, 21). The indentation modulus of an anisotropic solid is typically close to that of an isotropic, polycrystalline aggregate consisting of grains with the same elastic properties as the anisotropic solid. Correction factors for textured or single-crystalline films with arbitrary crystal symmetry can be readily calculated (17). The effect of substrate stiffness on contact compliance and indentation modulus has been investigated by a number of researchers (16, 22–24). An approximate analytic solution for this problem has been found by Gao et al., whereas the most accurate results have been obtained by Yu et al. With these results, it is possible to estimate how much the substrate contributes to the measured indentation modulus and hence get an estimate of the accuracy of the measurement. If the elastic properties of both film and substrate are known, the contact area between indenter and sample can be calculated precisely by equating the experimental contact compliance to the predicted value. This is useful when measuring the hardnesses of materials that pile up around the indenter. Neither the Doerner-Nix nor the Oliver-Pharr model gives a good estimate for the contact area in that case, and calculated values of hardness and indentation modulus can differ substantially from the actual values.

The pile-up behavior of a material depends on a number of factors. Materials that undergo little work hardening or that have large stiffness-to-yield stress ratios are especially prone to pile-up (25, 26). The residual stress in a material with a small work hardening coefficient has a profound influence on its pile-up behavior. A compressive stress in a film typically results in pile-up, whereas a tensile stress causes the material to sink in. This effect has been studied systematically by Bolsnakov et al. (27) and Syed Asif et al. (28) through use of finite element modeling.

Indentations made with a pyramidal indenter are all geometrically similar, and the average strain in the deformed regions around the indentations is independent of their depths. Pyramidal indenters are therefore ideal for determining the mechanical properties of a film as a function of indentation depth. The average strain in indentations made with
spherical indenters, on the other hand, increases with indentation depth (29). This makes spherical indenters less suited for an investigation of depth-dependent film properties. This type of indenter, however, can be used to estimate the strain hardening coefficient of a material (30–32).

**Bulge Testing**

The bulge test was one of the first techniques introduced for the study of thin film mechanical properties (33). In its original form, a thin film was clamped over a circular orifice and a uniform pressure applied to one side of the film. The deflection of the film was then measured as a function of the applied pressure and the pressure-deflection data were converted into a stress-strain curve for the film. Traditionally, the test has been plagued by a number of problems. The results are very sensitive to small variations in the dimensions of the sample (34). Failure to measure the deflection of the membrane accurately may result in apparent nonlinear elastic behavior of the film (35). The initial stress in the membrane has to be tensile. Because of its small bending stiffness, a free-standing film cannot support any compressive stresses, and the presence of such stresses causes the film to buckle or wrinkle. Finite element studies (36) of the deflection of circular membranes in compression have shown that the circumferential stress near the edge of the film remains compressive even at large applied pressures. Consequently, wrinkles near the edge of the film in compression disappear only gradually as the pressure is applied, resulting in meaningless data.

Use of silicon micromachining techniques has made it possible to overcome many of the problems associated with the bulge test (34, 37–44). Anisotropic etchants are used to fabricate perfectly square or rectangular membranes on (100) silicon substrates. These membranes can be tested readily, and results are typically very reproducible. The sample preparation technique has the additional advantage that the residual stress in the film is retained and may be measured. If the film is in compression, a composite membrane may be fabricated consisting of the film of interest and an additional film to keep the membrane in tension (21, 41).

Finite element calculations (36, 45) have shown that the pressure-deflection relationship for a thin circular membrane with a residual stress $\sigma_0$ is well approximated by

$$P = (1 - 0.241\nu) \left( \frac{8}{3} \right) \left( \frac{E}{1 - \nu} \right) \left( \frac{t}{a^4} \right) w_0^3 + 4 \left( \frac{\sigma_0 t}{a^2} \right) w_0$$

in the elastic regime, where $w_0$ is the deflection of the center of the
membrane, \( P \) is the applied pressure, and \( t \) and \( a \) are the membrane thickness and radius, respectively. Fitting Equation 1 to experimental data allows one to determine the biaxial modulus \( E/(1-\nu) \) and the residual stress \( \sigma_0 \) in the film. A series solution for the deflection of a thin circular film based on an analysis by Hencky (46) has been presented by Vlassak (21). This solution agrees well with Equation 1 and other numerical calculations (47) and holds as long as the bending stiffness of the membrane can be neglected. The importance of the membrane bending stiffness as compared to the stiffness due to the membrane-action of the residual stress is determined by the dimensionless number \((E/\sigma_0)(t/a)^2\). If this number is small (e.g. smaller than \(10^{-2}\)), the residual stress is the main contributor to the initial membrane stiffness, and the bending stiffness can be safely neglected. It should be noted that there is a bending moment present in the membrane due to the finite bending stiffness of the film. This bending moment is maximum near the edge of the membrane and decays within a few film thicknesses to almost zero in the rest of the membrane. This edge effect reduces the overall deflection of the membrane by an amount less than the film thickness. For large deflections, the membrane stiffness increases and may become comparable to the bending stiffness of the substrate. In this case, the deflection of the substrate has to be measured and must be subtracted from the membrane displacement (21).

The deflection of square membranes has been evaluated with finite elements (45) and by a method based on the principle of virtual work (21, 34, 38, 45). The deflection of a square membrane is governed by a cubic polynomial similar to Equation 1. The virtual work analysis agrees well with the finite element results if a polynomial displacement field is assumed (21, 34), but errors can be large for other displacement fields (38, 45).

The stress state in a long, rectangular membrane approaches a state of plane strain (21, 34). This makes it straightforward to analyze the membrane deflection for both elastic and plastic deformation (21). Therefore, rectangular membranes are particularly useful if one is interested in the complete stress-strain curve of a film (44). Figure 1 shows a typical plane-strain yield curve for a thin gold film measured with a rectangular membrane (SL Neumann, JJ Vlassak, WD Nix, unpublished data). Because the stress state is one of plane strain, the plain-strain modulus \( E/(1-\nu^2) \), rather than the biaxial modulus \( E/(1-\nu) \), is measured in this type of experiment.

**Substrate Curvature**

One of the most simple and widely used stress evaluation techniques for films attached to substrates involves measuring the curvature induced in
the substrate by the bending moment applied by the stressed film (10, 48, 49). The film stress may be related to the substrate curvature through the well-known expression (1, 50)

\[
\sigma = \left( \frac{E_s}{1 - \nu_s} \right) \left( \frac{t_s^2}{6t_f} \right) K_f,
\]

which describes the biaxial stress in the thin film in terms of \( t_f \), the film thickness; \( t_s \), the substrate thickness; \( E_s/(1 - \nu_s) \), the biaxial modulus of the substrate; and \( K_f \), the change in curvature due to the presence of the film. This simple expression is true (51) when the substrate is
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a thin, elastically isotropic plate that is free to bend, and the film has a uniform thickness with \( t_f \ll t_s \). Furthermore, the stress in the film must be isotropic in the plane, and the maximum deflection due to bending may be no larger than approximately \( t_f/2 \). Residual stresses in continuous films may be determined directly by this technique, whereas elastic and plastic properties may be evaluated by applying a load by varying the temperature. For films on silicon wafers, or other large film/substrate structures, curvature is typically evaluated by scanning the film surface with a laser beam and determining the angle of reflection (48, 49). The average curvature of the substrate is determined, and an average stress in the films may then be calculated. This technique has been successfully applied to a variety of continuous thin films (49, 52–54) on various substrates (48, 55) for a number of different applications.

If there is nonuniformity in film stress or thickness, measurement of curvature over less than the full substrate diameter will lead to error in the calculated stress (51). This is because the local curvature is partially dependent on the curvature of the remainder of the substrate. This was demonstrated by von Preissig, using an example of a film partially covering a substrate. For the case in which a film of radius \( r^* \) is deposited on a larger substrate of radius \( a \), the thermal strain in the film should still be given by \( \Delta a \Delta T \), the differences between the thermal expansion coefficients of the substrate and film, and the change in temperature. The bending moment induced by the residual stress in the film must be applied at the film’s edge (at \( r = r^* \)) rather than at the substrate’s edge (at \( r = a \)). The curvature from \( r = 0 \) to \( r = r^* \) is constant but is no longer described by the basic equation. Instead, the local curvature is reduced by a factor of

\[
1 - \left( \frac{1 - \nu}{2} \right) \left[ 1 - \left( \frac{r^*}{a} \right)^2 \right].
\]

This reduction may be quite large (approximately 25% for an Al film with \( r^* = a/2 \)) and could lead to large error in stress calculation if not considered. This formalism may be extended to the case of nonuniform stresses or thickness and demonstrates the danger of attempting to relate “local curvature” to “local stress.”

Substrate curvature techniques are rapid and nondestructive, so there is an incentive to attempt to extend the approach to the evaluation of patterned films. As demonstrated, evaluation of a non-continuous film relies on an average stress to produce an average curvature. Thus for quantitative evaluations, it is applicable only to arrays of identical structures with well-defined orientation relationships (e.g. parallel lines). If the length and width dimensions of the regions are not equal (e.g. lines
with small width/length ratio), then the curvature must be evaluated in two orthogonal directions. The areal density of the patterned film is then used to calculate stresses in the relevant coordinate system. Only the stresses in the plane of the substrate are measurable by substrate curvature. Reports of such measurements of bare Al alloy lines (56, 57), deposited on relatively large cantilever beams, agree well with finite element calculations and X-ray stress measurements of similar lines.

The addition of an encapsulating film constrains the patterned structures and induces a normal stress in the lines, considerably complicating analysis. The total curvature induced by multiple continuous films is just the sum of the individual contributions:

\[ K_{\text{total}} = \left( 1 - \frac{v}{E} \right) \frac{6}{r_s} \left( \sigma_1 t_1 + \sigma_2 t_2 + \cdots + \sigma_n t_n \right). \]

In the case of passivated lines, the encapsulating film is nonplanar, and its stress state is complex and depends upon the geometry of the underlying lines. Thus its contribution to the total curvature may not be simply evaluated. The curvatures perpendicular and parallel to the lines may be described as sums of the stress contributions of each layer adjusted by an effective film thickness for each film.

One of the most successful approaches for dealing with the encapsulating layer is that presented by Burges et al (56, 58), in which finite element modeling is used to approximate the encapsulant stresses in the direction perpendicular to the lines in the plane of the substrate. The encapsulant stress parallel to the lines is similar to that in a blanket film of the same material, therefore, experimental data were used. The validity of the finite element solution was tested with simulation and measurement of SiO₂ lines and a Si₃N₄ passivation that both behaved elastically throughout a thermal cycle (58). The reported values of stress perpendicular and parallel to Al alloy lines under an identical passivation agree well with X-ray data taken from the same samples.

More ambitious attempts to derive the full stress state of metal lines and a perfectly planarized passivation are based on a large number of assumptions about the relative stresses in these materials (57, 59). Although results of this type would be quite useful, the assumptions are not well supported by finite element analysis of similar structures (AS Mack & RP Vinci, unpublished calculations). Such calculations show that some of these assumptions may be valid at particular locations (e.g. at the line/passivation interfaces) but are not accurate on average. As an example, one of the cited assumptions is that the stress normal to the substrate in the top portion of the passivation (Figure 2) is, on average, equal to half the normal stress in the lines (59). This is
Figure 2  Schematic of passivated metal lines.

Based on stress continuity at the metal/passivation interface with a linear gradient to zero at the passivation free surface. This is a reasonable approximation for the stress directly above lines with an aspect ratio of 1:1. However, in that case, the normal stress in the top portion of the passivation between the lines is opposite in sign to that directly above the lines, leading to a volume average normal stress close to zero. Similar inconsistencies can be found in several other critical assumptions, particularly when lines of different aspect ratio are considered. Also, the assumptions do not address conformal (non-planarized) passivations. As a result of these difficulties, this beam bending analysis produces line stress data that resemble certain X-ray data, but not others.

X-Ray Diffraction
X-ray diffraction (60–62) is widely used for evaluating the mechanical behavior of thin film materials because of its accuracy and because a
wide range of sample sizes may be tested. X-ray techniques evaluate strains from measured interplanar spacings of crystalline materials using the simple expression

$$\varepsilon_{\psi\phi} = \frac{d_{\psi\phi} - d_0}{d_{\psi\phi}},$$

5.

in which the strain in any direction, $\varepsilon_{\psi\phi}$, defined by the angle $\psi$ from the film normal and $\phi$ in the plane of the film, is determined from the measured interplanar spacing in the same direction, $d_{\psi\phi}$, and the corresponding lattice spacing when unstrained $d_0$.

Strains in crystalline materials may be determined directly (60–62), while stresses in amorphous materials may be determined by measuring the strain induced in a crystalline substrate (62). The calculated strains are then converted to stresses using the elastic constants and crystallographic orientation of the film or substrate, as appropriate (63). For an isotropic crystalline film under equal biaxial strain, the interplanar spacing is given by

$$d_{\psi} = \left[ \sigma d_0 \left( \frac{1 + \nu}{E} \right) \right] \sin^2\psi + d_0 \left( 1 - \frac{2\nu}{E} \sigma \right),$$

6.

in which the angle $\phi$ is no longer needed. The stress may be readily derived from the slope of a plot of $d_{\psi}$ and $\sin^2\psi$.

Numerous studies have been performed on continuous films and multilayers that have extended the use and understanding of X-ray techniques (64–70), but many of the interesting advances during the last five years have come in determining strain in discontinuous films. Specifically, much progress has been made determining stresses in patterned structures such as those of concern to the integrated circuit (IC) industry (71–75). X-ray techniques offer a significant benefit over other methods for this particular application. It is possible to directly determine strains in metallic lines (such as IC interconnects) that lie beneath layers of other materials (such as interlayer dielectrics); furthermore, stresses in such structures are generally triaxial in nature, and X-rays permit direct determination of strain as a function of orientation.

Standard methods are limited, however, by spatial resolution. Typical X-ray beam diameters of several square millimeters are much larger than the micrometer scale of ICs, thus measurements represent averages over the areas of multiple features. Improved spatial resolution, on the micrometer or submicrometer scale, would allow greatly improved correlation between microstructure and local behavior such as voiding and hillocking.
Recently, there have been several efforts to provide a localized analysis capability. Dehaven et al. have presented measurements made with a 50-μm diameter beam (76). Modifications to the apparatus have allowed beam sizes as small as 10-μm in diameter (70). Yamamoto et al. have reported strain measurements from a white X-ray beam focused to approximately 0.8 μm diameter, using a parabolic glass capillary (77–80). Energy dispersion, rather than angular dispersion, was used to determine interplanar spacings of patterned lines. In this method, rather than evaluating spatial shifts in a diffraction peak using Bragg’s law, \( d_{hkl} = \lambda/(2\sin \theta_{hkl}) \), energy shifts are used: \( d_{hkl} = (hc/E_{hkl})/(2\sin \theta) \), where \( h \) and \( c \) are Planck’s constant and the speed of light, respectively. With this method, the precision is not sensitive to small errors in the 2θ angle of the detector. Only symmetric reflections were measured, so only strains normal to the sample surface were reported. It is unclear how a single symmetric measurement can provide the unstrained lattice parameter, and no comparison to more conventional measurements is yet provided. Thus it is difficult to evaluate the accuracy of the results.

A similar approach by Wang et al. recently used a tapered glass capillary to reduce white X-rays from a synchrotron source to a spot on the order of 30-μm in diameter (81). This technique requires a capillary to sample distance on the micrometer scale to minimize the effect of beam divergence. In this case, asymmetric reflection over a limited angular range allowed extrapolation of the full strain state of a measured continuous film. The few published results have large errors, however, and are difficult to compare to more reliable data.

X-ray strain measurements require a high degree of accuracy that may be unobtainable from micrometer-scale measurements. Nonetheless, if the inherent difficulties can be overcome, techniques such as these may provide extremely useful measurements. If accurate strain analysis remains elusive, however, microbeam diffraction techniques can still be very useful for microstructure mapping. This capability was demonstrated by Laue patterns produced from a single Al grain by Wang et al. using a 0.3-μm diameter beam (81).

**Optical Spectroscopy**

The need to determine strains in specific features on the micrometer scale has led to the investigation of a number of optical techniques based on piezospectroscopic phenomena (82). These techniques have been developed over the past 25 years, but two variations have arisen within the last five years that may expand the use of high resolution optical techniques, particularly for microelectronics technologies: the combination of finite element calculations and Raman spectroscopy.
measurements of a substrate for indirect determination of the stress in overlying metallic structures (83, 84), and the implementation of a stimulated luminescence technique with Raman microprobe optics (85, 86).

The Raman microprobe, developed over twenty years ago, has made it possible to examine stresses in many materials with micrometer scale lateral spatial resolution. Raman spectroscopy is based on the inelastic scattering of intense monochromatic high-frequency radiation (visible, ultraviolet, or near infrared laser light) that results from excitation of vibrational modes in molecular and crystalline materials. Stresses in crystalline solids produce small shifts in the Raman spectra that may be used to characterize localized stresses.

Optical spectroscopy is unsuitable for direct measurement of stress in metallic thin films and structures because the high reflectivity of the metallic surfaces minimizes the signal due to scattering. The use of micro-Raman spectroscopy for determining the stress in silicon is possible, however, and allows the indirect determination of stress in metal structures (83, 87). Silicon is highly reflective to visible wavelengths but is largely transparent to the infrared. The Raman probe is focused next to the metallic structure of interest (requiring optically transparent overlayers if such are to be present). In the unstrained lattice, the three silicon vibration modes have the same frequency [for example, see De Wolf et al (87) for details]. Strain lowers the symmetry of the lattice so that the three modes shift by different amounts. For the most general case, in which all strain tensor components are nonzero, the mode shifts can be related to the strain components by a series of simultaneous equations. Because of the large number of unknowns in this relation, there is generally insufficient information available to deduce the individual strain components. Significant simplifications and assumptions about the strain state must be made in order to deduce the stress distribution.

Measurement on many ceramics of technological interest has been hampered by their low Raman intensity. Fluorescence spectroscopy, a technique that grew out of early laser research (88, 89), has been used to examine stresses in alumina lightly doped with chromium (ruby) (90). In this technique, shifts of strong chromium fluorescence lines are used to evaluate stresses. The combination of this technique with Raman microprobe optics was reported by Molis & Clarke (85) and has since been used for examining a wide variety of structures (86, 91, 92). A recent example of both theoretical and technological interest is an investigation of stress redistribution during cracking of multilayers of sapphire and either aluminum or copper (93). Note that this is based
on different principles than those of an earlier piezospectroscopic study of stress fields around crack tips in magnesia partially stabilized zirconia (94). In this earlier study, standard Raman shifts were used to assess the tetragonal to monoclinic phase transformation that is indicative of the high stress field ahead of an advancing crack tip in this material. The two phases produce unique Raman spectra that are readily differentiated.

The most significant drawbacks to the use of piezospectroscopy for strain determination are the indirect nature of the technique, in the case of the metal lines, and the complexity of the shift/strain relations. Since insufficient information is available for determination of individual stress components, only relatively simple systems that lend themselves to accurate simulation are good candidates. In these cases, the Raman shifts may be compared with the simulated shifts with a reasonable degree of confidence. For materials that are not well understood, such as highly anisotropic materials, or for complicated stress states, the accuracy of finite element simulations must be fully evaluated using more conventional techniques.

**Electron Spectroscopy**

Stress evaluation of submicrometer and deep submicrometer features, such as those associated with crystalline defects or stress concentrators near thin film edges and corners, is extremely difficult due to the resolution limits of conventional techniques. Electron diffraction in a transmission electron microscope (TEM) is one of the few options available for examining the deep submicrometer regime. While it is possible to use convergent beam electron diffraction (CBED) to obtain local lattice displacements with a high degree of accuracy at a particular point under ideal conditions (95), numerous measurements are required to fully characterize a displacement field. An alternative approach is to use electron diffraction contrast imaging (EDCI) to map a displacement field over a two-dimensional region. This technique has a theoretical resolution limit of approximately one nanometer. The resulting image does not provide quantitative information directly. Instead, a quantitative displacement field is inferred from comparisons to EDC images produced by theoretical models and simulation tools.

Recent work in this area has been published by Janssens and colleagues. They are developing modular software (96) (SIMCON) for the simulation of EDC images of localized displacement fields associated with a wide variety of defects and geometries (e.g. dislocations, precipitates, and lithographically defined microscopic structures). For example, with their technique they have demonstrated measurements...
near the corners of micrometer scale microelectronics structures (96–99). Reasonable quantitative agreement with CBED and micro-Raman measurements of the same structures is presented. Work on error reduction through improved imaging parameter analysis and displacement field simulation is continuing.

Potential sources of error for the EDCI technique are numerous and fall into two general categories (95, 97, 98): experimental difficulties and limitations of image and displacement field simulations. First, sample preparation is critical, particularly where thickness variations or changes in the stress state are concerned. Sample preparation inherently involves severe thinning of the features of interest, which may produce stresses not present in the original structure. Sample thinning also leads to changes in the displacement fields. A plane stress condition develops, and rapid relaxation may occur. Thus the relevance to realistic structures may be difficult to determine. An elastic solution may be able to account for some of this change, but, as Armigliato and coworkers have reported (99), adjusted EDCI measurements do not fully match those from other techniques such as micro-Raman spectroscopy. Cooling of the sample during imaging, necessary if sharp CBED patterns are to be produced for calibration of the simulation, will also change the stress state. Finally, thickness variations will add thickness fringes to the image and may also alter the strain field.

If conventional photographs are used for imaging, comparison with simulations may be limited to qualitative agreement. Uncertainty associated with nonlinear photographic emulsions and processes may lead to considerable error in quantitative analysis. A linear image recording device such as a CCD camera could significantly reduce this problem (97) and allow automated image comparison. Even with automated image matching, however, it may be difficult to correctly deal with subtle differences in the simulated images. A complete understanding of experimental imaging conditions is also necessary for accurate image modeling, which includes a number of factors such as material absorption parameters that are often difficult to determine. Finally, other modeling limitations, such as those resulting from the use of conventional elastic solutions that do not apply near defect cores, add significant sources of error.

EDCI has the potential to provide quantitative strain data not readily available by other techniques, but it is susceptible to many of the general problems associated with TEM sample preparation and imaging. A number of challenges in image collection and simulation must still be met before the technique can be reliably applied to general problems.
RECENT ADVANCES IN THE UNDERSTANDING OF THIN FILM BEHAVIOR

Continuous Films

LOW-TEMPERATURE PLASTIC FLOW IN AL AND CU FILMS It is well known that the behavior of thin film materials is significantly different from the same materials in bulk form. While elastic properties remain unchanged, thin films often support stresses that would ordinarily cause rapid relaxation through plastic flow. It is now recognized that much of the strengthening effect is caused by the film/substrate interface and the small grain size typically encountered.

A model specifically addressing the effects of a rigid substrate and free surface has been developed by Nix (3). He attributes the strength enhancement of thin films to a constraint imposed on dislocation motion by the film/substrate interface. Following the approach of Freund (100, 101), who treats the formation of misfit dislocations in epitaxial films, this model depicts dislocation glide in thin films as a process that must create a dislocation line at the film/substrate interface as it occurs. Thus in order for a dislocation to move, the work done by the stress in the film must be sufficient to deposit an interfacial dislocation. If an oxide or passivation layer is present on the film surface, a second dislocation must be created at the film/oxide interface, further raising the energy requirements for dislocation glide.

The equation describing the flow stress in a (111) textured fcc film is

$$\sigma_{\text{flow}} = 3.464 \frac{b}{2\pi(1-\nu)h} \left[ \frac{\mu_i\mu_s}{\mu_t + \mu_s} \ln \left( \frac{\beta_s h}{b} \right) + \frac{\mu_i\mu_o}{\mu_t + \mu_o} \ln \left( \frac{\beta_o t}{b} \right) \right],$$

where $b$ is the magnitude of the Burgers vector; $h$ is the film thickness; $t$ is the oxide thickness; $\mu_t$, $\mu_s$, and $\mu_o$ are the elastic shear moduli of the film, substrate, and oxide, respectively; and $\beta_s$ and $\beta_o$ are constants. If an oxide or passivation layer is not present, then the second term in the parentheses is omitted.

This model is largely athermal, with temperature dependence only in the shear modulus terms, and is therefore unsuitable for determining stress as a function of temperature throughout a thermal cycle. It also neglects grain size strengthening or other obstacles to dislocation motion. Therefore, it is common to add a grain size strengthening term, in the form of the bulk Hall-Petch relation (102, 103), to the flow stress expression for fine-grained films at low temperatures, for which this...
The total flow stress is simply the sum of the two mechanisms:

$$\sigma_{\text{flow, total}} = \sigma_{\text{flow}} + kd^{-1/2},$$

where $d$ is median grain size and $k$ is the Hall-Petch coefficient.

Thompson has developed a model (104) similar to that of Nix in which the room-temperature flow stress is given by

$$\sigma_{\text{flow}} = \frac{3.464}{4\pi(1 - \nu)} \ln \left( \frac{d}{b} \right) \left( \frac{2}{0.943d + \frac{1}{h}} \right),$$

which explicitly includes grain size strengthening with the addition of $d$, the grain size. Both models predict strengthening that is dependent on the crystallographic orientation of the film normal, but the Thompson model also predicts that the grain size effect is influenced by the film orientation.

The applicability of these models to the stress-temperature behavior of Al alloy films on oxidized silicon wafers has been tested by Venkatraman et al (103, 105-107). Dislocation glide, observed during TEM in situ heating experiments (105), was identified as an important relaxation mechanism below 300°C. Dislocation loops were seen to move along the inclined $\{111\}$ planes, leaving behind dislocations at the Al/SiO$_2$ and Al/Al$_2$O$_3$ interfaces. Pile-ups of dislocations were also seen near the grain boundaries, emphasizing the role of grain size in these small-grained films. These observations are in agreement with earlier studies of flow in Pb films (64).

Separation of the strengthening contributions is difficult because grain size is typically a function of film thickness. Venkatraman carried out two novel studies that allowed separate determination of the thickness and grain size components. First, substrate curvature stress measurements were conducted on pure Al films ranging from 0.25- to 1.52-$\mu$m thick that had been subjected to a laser reflow operation (106). Reflow increased the average grain size to 8-$\mu$m in the thinnest film and to more than 20-$\mu$m in the remaining films, well above levels at which grain boundaries should play a significant role in film strengthening. It was found that the flow stress close to room temperature varies as 1/thickness, as predicted by the dislocation models. Furthermore, the flow stresses were reduced compared with similar films with small grain sizes.

In the second set of experiments, measurements of stress as a function of temperature were performed on films with a small, fixed grain size, but varying film thickness (103). The grain size and thickness were...
controlled by growing the films to the same initial thickness, then growing and removing controlled amounts of anodic oxide. This method produced smooth surfaces and uncontaminated films. When the flow stress near room temperature was plotted against film thickness, as in the previous experiment, it was seen that the strengthening was increased compared with that observed in the reflowed films. The slope, however, was similar, indicating that the thickness strengthening component remained consistent. The offset between the sets of data is due to the contribution of grain boundary strengthening. When the Hall-Petch relation is assumed, the grain size strengthening components yield slightly higher Hall-Petch coefficients than are reported for bulk Al.

Measurement of flow stress as a function of crystallographic orientation is possible using X-ray techniques, which are inherently orientation sensitive. The stress in (111) and (100) grain populations in Cu thin films was evaluated in films of several thicknesses by Vinci et al (108). Stresses at room temperature were clearly orientation dependent, with the (111) grains supporting a higher stress, as predicted by the dislocation glide models. Surprisingly, although the (111) grains demonstrated the expected thickness dependence, the room temperature flow stress of the (100) grains appeared to be invariant with film thickness. This result has not been fully explained.

HIGH-TEMPERATURE PLASTIC FLOW IN Al AND Cu FILMS As discussed above, the dislocation glide models predict a largely athermal strengthening behavior, one which is not seen in practice. Although polycrystalline Al films show some loss of strength at high temperatures (49), high-temperature stresses in Cu thin films relax very rapidly (69, 109, 110), often dropping to zero by 400°C (Figure 3). This is surprising in light of the high melting temperature of Cu compared with Al (1083 vs 660°C). Moreover, the measurements of stress as a function of orientation in Cu films (69, 108) showed no clear orientation dependence in relaxation above approximately 300°C. Time-independent dislocation glide models, in their current forms, do not adequately address the high-temperature behavior of either material, so an alternative approach must be sought.

Attempts have been made to address the high-temperature behavior of thin films by applying bulk relaxation models. One such model, based on Frost-Ashby deformation maps (111), was recently modified by Thouless et al to produce results that closely resemble stress-temperature curves for 1-μm thick bare copper films (110). The deformation
map concept is based on a number of complementary and competing deformation processes that have been identified in bulk materials. These include diffusional creep, dislocation climb, and dislocation glide. Each mechanism is assumed to dominate the strain relaxation at a different combination of stress and temperature.

Thouless modified the bulk strain-rate expressions for diffusional deformation by grain boundary diffusion (Coble creep) and lattice diffusion (Nabarro-Herring creep) to account for the equibiaxial stress state common in thin films. Whereas the classic Coble creep and Nabarro-Herring creep mechanisms assume a chemical potential gradient resulting from a uniaxial stress that drives diffusion between neighboring grain boundaries, Thouless includes the flow of matter between the grain boundaries and the free surface of the film. This introduces film thickness and grain size terms into the diffusional rate equations. All other rate equations retain their classic form, remaining independent of film thickness or grain size.

The lack of dependence of the classic equations on what Frost & Ashby refer to as the internal variables (i.e. grain size, dislocation density) is due to a simplifying assumption, that of steady-state deformation (111). Under this assumption, the internal variables are determined by the external variables of stress and temperature. This assumption only holds true at high temperatures for bulk materials, and often only at
large strains. The steady-state assumption may not apply at all to thin film materials at very small strains over the temperatures and heating rates of interest to the integrated circuit industry. Thus this approach must be applied with caution.

It is reasonable to postulate that rapid surface diffusion may be driving the high temperature relaxation in Cu films, which, unlike Al films, does not develop a stable oxide. The deformation map approach does predict a rapid loss of strength in Cu films at high temperature, where relaxation by diffusion dominates the model (110). It does not, however, include the thickness component of strengthening at low temperature, where it is known to be significant. Furthermore, it has been shown that the addition of a surface layer such as Si$_3$N$_4$ significantly alters the high-temperature behavior of Cu films (69, 112, 113) (perhaps by fulfilling the role played by the Al$_2$O$_3$ layer on Al films). The presence of the surface layer leads to a degree of strength retention similar to that in Al. Removing all surface diffusion from the deformation-map model is insufficient to predict this surface effect (69), which indicates that application of the deformation-map model to thin films will require additional modification (69, 114).

The orientation dependence of flow by dislocation glide in epitaxial Al was tested by Besser et al (115). They measured stress in (111) single crystal and (110) bicrystal films during annealing cycles to 400°C using X-ray techniques. In compression, the results demonstrate a clear orientational dependence, with the (110) film yielding at approximately 200 MPa, whereas the (111) film behaved elastically to approximately 400 MPa. This difference may have consequences for the development of certain morphological features, as discussed below. It was not possible to test the yield behavior at low temperature as neither film reached a high enough tensile stress to cause plasticity, but this result seems to support an orientation dependence of the flow stress in Al films to at least 400°C, the maximum annealing temperature for this study.

STRAIN ENERGY EFFECTS ON FILM MORPHOLOGY Many polycrystalline thin films demonstrate a columnar grain structure with preferred crystallographic orientation (116). In fcc metals like Al, the preferred orientation or texture tends to align {111} planes parallel to the surface of the film. Thompson has explained this tendency in terms of surface and interface energetics that drive abnormal grain growth (116, 117). Even in strongly textured films, however, a small population of misoriented grains is likely. It was recently proposed by Sanchez & Arzt that the strain energy density difference that exists between grains of different crystallographic orientation can lead to the development of certain mor-
phological changes in film microstructure (118, 119). Hillock, sunken grains, and abnormal grain growth of orientations other than those favored by surface energy may be encouraged by the crystallographic dependence of elasticity and plastic flow (116, 120, 121).

A grain with orientation \((hkl)\) with respect to the film surface, and a biaxial thermal strain \(\varepsilon\) imposed by the thermal expansion mismatch between the film and the substrate, has an elastic strain energy density \(W_\varepsilon = M_{hkl} \varepsilon^2\), where \(M_{hkl}\) is the appropriate biaxial modulus. If elastic anisotropy exists, as it does for most metals, significant differences in \(W_\varepsilon\) can develop for materials under isostrain averaging (the assumption that the biaxial strain is the same in all grains). If yielding occurs, the strain energy density is described by \(W_\varepsilon = \sigma_y^2 / M_{hkl}\). As described above, the yield stress is also dependent on crystallographic orientation (and geometry, in the case of the Thompson model) (3, 104). Thus free-energy differences between grains of different orientation can exist in both the elastic and plastic regimes.

The effect of a strain energy driving force on grain growth was examined in a pair of studies on Cu and Ag films by Zielinski et al (55, 121-123) and Carel et al (124, 125), respectively. As part of the study, Zielinski deposited 1.2-\(\mu\)m thick Cu films (coefficient of thermal expansion \(16.6 \times 10^{-6} \, ^\circ\text{C}^{-1}\)) on Cu, Si (\(3.0 \times 10^{-6} \, ^\circ\text{C}^{-1}\)), or Al (\(25.3 \times 10^{-6} \, ^\circ\text{C}^{-1}\)) substrates (55, 126). Layers of Si\(_3\)N\(_4\) and W separated the Cu from the substrates to eliminate heterogeneous interface effects. Abnormal growth of giant (100)-oriented grains was seen in the Cu/Si and Cu/Al cases, which went into compression and tension, respectively, during annealing. Only normal growth was observed for Cu/Cu, where the thermal strain was zero. Separate X-ray measurements of (111) and (100) grain populations showed a maximum strain energy density difference of \(68 \times 10^3 \, \text{J m}^{-3}\) between (100) and (111) grains in a Cu film on Si in which abnormal grain growth was observed (123). An opposing driving force from interface and surface energy differences was calculated to be approximately \(-20 \times 10^3 \, \text{J m}^{-3}\), so the strain energy density difference was sufficient to cause the preferential (100) growth (Figure 4).

Carel et al examined the growth of polycrystalline Ag on single crystal (001) Ni grown on (001) MgO, both by experiment (124) and by simulation (125). Calculations showed that interface energy favored a (111) Ag texture, with a (111)Ag||(001)Ni alignment. On the other hand, strain energy differences favored a (001) Ag orientation because (111) grains are much stiffer in-plane than (100) grains. The experimental results show that it is possible to develop a (001) Ag orientation when the film is highly strained, when the film is sufficiently thick to
reduce the surface and interface driving forces, and when strain relief by plastic flow is suppressed. The simulation results support a strain energy driving force for growth of the (001) orientation and indicate that the primary growth mode can actually change during an anneal as the grain size increases. The simulation calculates a strain energy density that is a decreasing function of grain size after yielding (combined with a $\sigma_y$ that is grain size dependent, according to the Thompson model) and assumes an initial grain size on the order of the film thickness. This results in a relationship for the boundary between the interface and strain energy-dominated growth regions.

MODULUS AND HARDNESS ENHANCEMENT IN MULTILAYERS  As films become thinner, the effects of surfaces and interfaces begin to dominate the observed properties (3). Compositionally modulated films (CMFs), comprised of periodically alternating layers of different materials, have been known to exhibit unusual optical, electrical, magnetic, and mechanical properties. Because the elastic behavior of a material is not
structure sensitive, it is not expected that CMFs should exhibit elastic behavior other than that predicted by rule-of-mixtures calculations. Surprisingly, though, several reports of a greatly enhanced elastic modulus (the “supermodulus effect”) have appeared in the literature over the past fifteen years. While this generated considerable excitement, it now appears that the reported enhancement, sometimes of more than 100%, was due to artifacts of the mechanical testing techniques used (36, 43, 127). A recent analysis demonstrates that compressive stresses in the films could lead to supermodulus-like nonlinear behavior as a result of buckling and uncertainty in film height during bulge testing (127). However, measurements of certain CMF systems using Brillouin scattering (which is not susceptible to the uncertainties present in the early mechanical measurements) and improved mechanical measurements (using techniques such as those described in the first section of this review), suggest that small anomalous changes in elastic modulus are possible (128, 129). In contrast to the supermodulus effect, slightly low elastic moduli are often observed for small bilayer periods (the repeat distance of two layers) in a number of CMF systems.

In contrast to elastic properties, plastic properties such as hardness are expected to depend strongly on bilayer period, layer intermixing, coherency strains, and other interface-moderated factors (129). This brings up the possibility of extremely hard materials useful for their abrasion and wear characteristics. Recent work strongly indicates that the primary cause of hardness enhancement in many CMF systems is due to a modulus difference between the layers (130–132). When a CMF has a spatially varying elastic modulus, dislocations in the layers with the lower modulus are repelled from the nearest interface due to a force from an image dislocation in the adjoining higher modulus layer. This concept was introduced early in the study of CMFs (133–135), but recently introduced models include the effects of multiple interfaces and image dislocations in distant layers (136). These modifications have resulted in reasonable predictions of hardness for small bilayer periods in certain multilayer systems. However, it appears that other strengthening mechanisms, such as coherency strains that develop between layers with mismatched lattice parameters, may play a significant role in certain multilayer systems and cannot be summarily neglected (130, 136). When the bilayer period of a CMF is large, dislocations may be able to move within a layer, allowing plastic flow at stresses below those required to move a dislocation past an interface. This may account for the decrease in hardness with increasing bilayer period usually observed. This has typically been modeled using the phenom-
enological Hall-Petch relation (130, 134, 137), but other, more mechanistic explanations have also been offered (129).

An interesting variation of the use of CMFs for hard industrial coatings is a recent report of extremely hard crystalline carbon nitride/TiN multilayers (138). In the late 1980s, calculations of theoretical bulk moduli for covalent solids indicated that a hypothetical compound, $\beta$-C$_3$N$_4$, could possess a hardness greater than that of diamond (139). Unfortunately, efforts to produce this compound have typically resulted in amorphous carbon nitride films (a-CN$_x$) that exhibit unspectacular hardnesses. Controversial reports of production of small quantities of $\beta$-C$_3$N$_4$ have been made, but no modulus/hardness data were reported. Assuming that $\beta$-C$_3$N$_4$ is metastable, Li et al postulated that it might be possible to grow the desired phase by magnetron sputtering onto a lattice-matched substrate (138). TiN, with a theoretical lattice mismatch of $\sim$7%, was used to encourage the growth and stability of the desired carbon nitride phase. Because the lattice match was not perfect, only small thicknesses of $\beta$-C$_3$N$_4$ were produced before the developing strain energy caused a phase change. By creating multilayers with small bilayer periods, however, it appears that crystalline $\beta$-C$_3$N$_4$ layers were formed between the TiN layers. Nanoindentation tests of this multilayer material demonstrated indentation hardnesses of 45–55 GPa, approaching the hardness of diamond (60–100 GPa). The industrial demand for easily fabricated hard coatings is significant, so work in this area will continue with different lattice-match materials for which the lattice mismatch is even smaller than for TiN. If, as the authors believe, the strengthening is due to the $\beta$-C$_3$N$_4$ layers, and not to a CMF hardness enhancement effect, then the ability to grow thicker carbon nitride layers should increase the hardness of the coating.

Because the potential applications for compositionally modulated films are many, and the prospects for unique and useful properties are good, multilayer films are currently receiving significant attention. The reader is referred to several recent reviews for further information on this growing field (129, 140, 141).

**Patterned Films**

The mechanical behavior of Al alloy IC interconnects has received significant attention because the stress state of the lines directly affects their reliability. As in many continuous films, stress develops primarily as a result of the difference in coefficient of thermal expansion between the Al alloy lines and the Si substrate (49). The circumstances are complicated by the lithographic patterning of the Al alloy film into arrays of lines, often less than a micrometer in thickness and width but
hundreds of micrometers in length. Further complicating the stress state is the presence of a dielectric layer encapsulating the lines. The dielectric often has a thermal expansion coefficient close to that of the Si substrate and is mechanically stiff. This leads to a complicated triaxial stress state developed in the metal during temperature changes typical of IC processing technologies (142).

Simply patterning a film lithographically has direct consequences for the stress state, even before passivation deposition (72, 74, 142–145). Al alloy films on Si substrates typically exhibit a (111) texture normal to the film plane and random crystallite orientation in the film plane. As a result, continuous films sustain an equal biaxial stress state. When the film is patterned into long lines, the narrow transverse dimension is relieved of much of the stress. The amount of relief depends on the aspect ratio of the lines, with wider lines approaching the behavior of continuous films. Unpassivated lines typically exhibit stresses of magnitude similar to that of continuous films along the length of the lines, where the constraint is maintained. Even this stress is reduced, however, because of the reduction in the contribution of the Poisson contraction from the transverse stress. This has been demonstrated by several authors using X-ray measurements to determine the orientational components of the stress tensor (72, 74, 142, 143). The stress normal to the substrate remains at zero.

The addition of a stiff passivation layer partially replaces the constraint in the transverse direction and also adds a new constraint in the film thickness direction. Flinn & Chiang measured the average stress state of Al alloy lines under various passivation materials and demonstrated that the stress in the metal lines was independent of the stress state of the passivation materials (142). This finding was confirmed by Greenebaum et al (143). Hinode et al suggested that the metal stress is dependent on the passivation thickness and, therefore, the stiffness, reinforcing the role of the passivation as a mechanical constraint (146). The negligible effect of passivation stress state on line stress has also been shown through finite element modeling (147). In general, the stress in the length direction of the line greatly exceeds that in the width and normal directions (72, 142, 148). This is to be expected from the compliant constraint imposed by the overlying passivation layer.

The first measurements of stress as a function of temperature demonstrated that much of the behavior of passivated lines is elastic (72). The triaxial stress state found in metal lines encapsulated by a stiff dielectric has a large hydrostatic component. Unfortunately, a large hydrostatic stress is relieved by void growth that can lead to open circuits in lines with small dimensions. Further measurements of passiv-
ated lines have confirmed the earlier results and have demonstrated that the majority of relaxation by yielding is confined to the length direction of the Al alloy lines (148, 149). The strain in the normal direction may actually increase over time. This makes the total stress state relatively more hydrostatic, but differences in the individual stress components may remain.

The adoption of mechanically compliant dielectric materials, or mechanically anisotropic metals such as copper, will complicate the understanding of the behavior of passivated interconnects. Studies performed on such materials have shown some significant differences in the nature of the stress state in the metal lines (150, 151). Also, few studies of extremely small lines with large height/width aspect ratios are available, while such structures are becoming more common in industry. To date, the most reliable data have been from average measurements over many lines. Local stresses due to irregularities in topography, microstructure, or relaxation processes are still not well understood. The importance to industry is underscored, however, by the recent success of reliability models that depend on local stress variations as the primary consideration (152).

CONCLUSIONS

Development of thin film mechanical testing methods has resulted in several improvements in recent years, with a number of well-characterized tests for thin film mechanical properties, but challenges remain. Although nanoindentation is now a mature technique, more work needs to be done on understanding pile-up behavior. This will make nanoindentation more reliable for materials that demonstrate this behavior. The continuing development of techniques for examining the strain state of small volumes of material offers the possibility of understanding the links between local microstructures and film behavior. Many of these techniques, such as substrate micro-Raman spectroscopy and electron diffraction contrast imaging, will not be widely adopted until the reliability of the necessary simulation tools has been proven. It must still be demonstrated that microbeam X-ray techniques have the necessary accuracy and intensity to measure the typical strains in submicrometer patterned metals used in the integrated circuit industry. Nevertheless, the usefulness of such techniques should drive further investigation.

Despite the increasing study of the mechanical properties of numerous thin film systems, basic deformation processes are still not well understood over a range of temperatures and microstructures. Much of this comes from the difficulty of separating such factors as film thick-
ness and grain size, and stress state and temperature. Additional work will be needed to identify the relaxation mechanisms responsible for the behavior of thin film materials under different conditions because it is clear that direct application of bulk relaxation models is insufficient to describe thin film behavior.

ACKNOWLEDGMENTS

The authors thank WD Nix and JC Bravman for support and discussion throughout the development of this manuscript. Financial support from the Semiconductor Research Corporation (contract number 94-MJ-103), the National Science Foundation, Crystallume (contract number CRY 93-C-0199), and the Department of Energy (contract number DE-FG03-89ER45387) is gratefully acknowledged.

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