

MTS Model Theory in Small Volume Electronic Materials

R. Saptono* and C-U. Kim**

*Universitas Indonesia, Kampus UI Depok, Depok, JABAR 16424, Indonesia, rahmat.saptono@ui.ac.id

**University of Texas, Arlington, TX 76019, USA, choongun@uta.edu

ABSTRACT

Efforts have been made to study and model the temperature-dependence of flow stress. Mechanical Threshold Stress (MTS) model is one of the most admired physical mechanistic models. The model was developed on the basis of dislocation-mediated, thermally-activated, obstacle-controlled local process of dislocation overcoming obstacle. The present paper deals with the development and application of MTS model theory in small volume electronic materials. Small volume term corresponds to a volume characterized by a high surface to volume ratio including foil and thin film either with or without confinement. First, foundations of the MTS model theory are revisited. Then, specific issues in small volume materials are elaborated. Discussion is limited within the kinetic regime of thermal activation. It is shown that geometrical and micro-structural constraints are mixed up in small volume material, modifying the local mechanism of interest. As mechanical behaviour is averaged only over a small section volume, reduced geometry and increased surface to volume ratio have become more significant and articulated. More efforts must be made to reinterpret the model already established in bulk materials to accurately describe the behaviour in small volume material and extract plausible parameters related to the responsible mechanism.

Keywords: mts, model, electronics

1 INTRODUCTION

Plastic response of materials subjected to mechanical loading under various strain rates and temperatures has been extensively studied for many reasons and purposes. Attempts have been made to achieve ultimate benefits out of physical experimentation. Modeling is one of the best ways to gain understanding of this multifaceted behavior. A model idealizes the complex phenomena. It simplifies the overall while capturing the ground rules[1].

At best, a material model reveals the essential physics of the problem. A so-called physical material model represents accurately the relationship between the variables of interest, produces new data, and searches optimum. Such an ideal model encloses physically-sound material parameters. These parameters should, if possible, be applicable for a wide range of conditions and materials, kept least, and simply measured[2]. It does not, however, necessarily imply that there exists a completely general material response equation over a broad range of

experimental argument. If there is any, it might be too complicated for any practical use. A more productive approach, according to Kocks[2] is to specifically deal with materials classes, variables regimes, and behavior characteristic, for which one can specify a 'universal' constitutive law.

The earliest physical model in bulk material was developed from the study of the flow stress variation of structural steel with strain rate[3]. Equivalence effects of strain rates and temperature on the relation of stress-strain was assumed. These coupling effects were represented by a single parameter, later known as Zener Parameter. Rate characteristic was involved in this parameter through a factor associated with the heats of activation. One of the major drawbacks of this model is the fact that it does not explain clearly the responsible mechanism at microscopic level.

On the other hand, Zerilli-Armstrong (Z-A) and Mechanical Threshold Stress (MTS) models had enclosed physical mechanisms. These two so-called physical-mechanistic models were developed on the basis of dislocation plasticity theories along with first principles. Dislocation plasticity theory generally assumes that plastic deformation is mostly dictated by a local glide or climb of dislocations. Thermodynamically, there is an activated state in between two equilibrium states. Random thermal vibrations of atoms may occasionally meet the amount of energy required to reach such a state and hence defining a driving force for the motion. More specifically, the two models assumed that the inelastic flow is completely controlled by a thermally-activated process of dislocation lines overcoming local obstacles. Velocity of microscopic dislocation motions can be related to the macroscopic strain rate by flow kinetics.

Kinetics of flow is normally dictated by the slowest event. In most cases, the kinetics is much likely dependent on the waiting time at the local barrier rather than the excess time used up by the dislocations to move from one to another. A thermally-activated process will occasionally provide pre-requisite energy to release a specific dislocation at a particular time. Statistically, the frequency of the attempts and the probability of success will govern the kinetics. It is generally accepted that the probability of a system changing its equilibrium position from one state to another state with energy larger by ΔG is governed by the Boltzmann statistics.

At absolute zero temperature, the amount of energy required to release pinned dislocations at local obstacles (the obstacle energy) must be entirely supplied by

mechanical energy alone. At higher temperature, part of them can be provided by thermal activation energy. Gibbs free energy of activation, therefore, is assumed to be dependent on the applied stress (associated with mechanical energy) along with the geometry dislocations (associated with the obstacle energy or the threshold energy). More specifically, the activation energy is dependent on the ratio of effective and threshold stress.

Zerilli and Armstrong[4] used a physical dimension, i.e. the volume of activation or the mean value of it, to characterize the thermal activation process. It was revealed that a particular structure has its own dislocation characteristics by which the behavior is mostly governed. Two constitutive equations had been developed, to specify the thermal activation behavior of BCC and FCC materials. Based on the fact that the flow stress of FCC materials are less sensitive to the temperature change compared to that of BCC materials, it was assumed that Peierls potential barriers associated with a single dislocation motion played a significant role. It is reasonable to assume that in closed-packed structure dislocation can easily glide without considerable Peierls resistance. Temperature-dependence of flow stresses in these types of materials will be then very much dependent on the quality and quantity of the local obstacles.

MTS model, quite differently, assumed flow stress at absolute zero as a primary internal state parameter, linking it to the internal structure in both response and structure evolution equations. A scaling factor, a function of temperature and strain rate containing Arrhenius rate equation, was introduced to link directly the flow stress in the thermally-activated region to the reference stress.

Foundation for the MTS model had been established in the mid-1970s[5,6]. Kocks et.al.[7-10] had developed a constitutive equation for metal plasticity based on the use of MTS as the internal state variable. The equations count the slip kinetics at both constant and evolving structure.

Nemat-Nasser et.al. had elaborated MTS constitutive equations for FCC poly-crystals (OFHC Cu)[11,12] as well as BCC poly-crystals (Ta)[13,14]. Such a physical-based model was particularly developed based upon the established concept of dislocation kinetics along with systematic empirical observations from experimental data. Scope of the models is mainly in the range of temperature and strain rate when transport phenomena are not dominant and plastic deformation occurs basically by the motion of dislocations[15]. The model assumed that resistance to the movement dictates plastic flow. The flow stress equations could be generally expressed as the summation of (1) thermal, (2) a-thermal, and (3) viscous drag components. Specifically, Nemat-Nasser[15] introduced strain as the load parameter dictating the change of dislocation density in athermal region. This parameter can be simplified by assuming a simple power law representation and averaging the value of reference shear modulus. In viscous region, a phenomenological constitutive equation with a material constant measured directly at very high strain rate and high

temperature, were introduced. A single parameter representing an effective damping to dislocation motions was able to be specified. Nemat-Nasser[15] also proposed three-dimensional model based on the same principles.

More applicative MTS models for important structural materials have also been developed. Important dedicated physical-based models were developed, for instance, by Goto et.al.[16], Oussouaddi et.al.[17], and Banerjee[18] for Steel, as well as Banerjee and Bhawalkar[19] for Aluminum.

Holmedal[20] generalized MTS model to accommodate applications in continuum and crystal plasticity. An elastic-plastic formulation had been developed to be implemented in a tensile/compression test involving tools modulus. The modification allowed the model to be fully used in continuum and polycrystal applications. Mathematical expression for the thermal stress component was formulated in two equations.

Lately, Cai et.al.[21] modified and combined the power law relations of Z-A and MTS model to predict the stress to deform plastically as a function of thermally-activated stress and temperature over a large range of strain rate and temperature. The new constitutive equation can also concurrently depict the change of activation volume with the thermally-activated stress and temperature.

Not only has the MTS model been established in bulk materials, it has also been adopted to explain the temperature-dependence of flow stress in confined materials. Most of the models have been developed to explain the plastic behavior of thin film metals on substrates subjected to cyclic thermal loading. This is the prototypical-phenomenon of thin film metal interconnect.

Flinn et.al.[22,23], Volkert et.al.[24], Witvrouw et.al.[25,26], Proost et.al.[27], and Gruber et.al.[28], Kobrinisky and Thompson[29], made efforts to explain the temperature-dependence of stress phenomena in thin film on substrate by applying theories already established in bulk metals.

The present paper deals with the development and application of MTS model theory in small volume electronic materials. Firstly, the common grounds of the MTS model theory will be revisited. Secondly, specific issues in small volume material will be elaborated. Our discussion is limited in the region below the dynamic viscous region. Small volume label in the present paper represents a volume which is characterized by a high ratio of surface to volume including foil and thin film, either with or without confinement should be produced within the dimensions.

2 FOUNDATION OF THE MODEL

2.1 Mechanical Threshold Stress

Central of the MTS theory is rest at the use of mechanical threshold stress as a primary internal state parameter. To get the sound picture of mechanical

threshold stress, we begin with a familiar term established in stress strain diagram (figure 1).

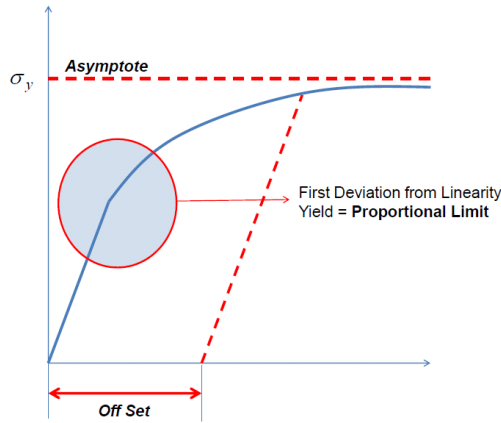


Figure 1. Yield in a stress-strain diagram.

It is well-recognized that yield is a point where linear elastic deformation departs from its proportional relationship. Physically, a yield point or yield surface in three dimensional state of stress separates a region of static mechanical equilibrium from a region outside it where the initial equilibrium is no longer exist (figure 2)

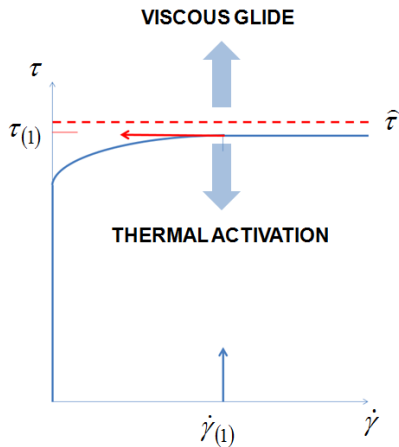


Figure 2. Mechanical threshold stress.

A similar concept does exist in micro-plasticity involving thermodynamic equilibrium (figure 2). Practically, it is defined as a flow stress in stress-temperature diagram extrapolated to absolute zero. Firstly, it can be first understood as a flow stress in the absent of thermal activation process. More precisely, it is a boundary stress that separates a kinetic region where dislocations are rest at their equilibrium position (thermal activation region) from another region where dislocations cannot find any equilibrium positions (dynamic equilibrium). This concept involves thermodynamics equilibrium in addition to statics equilibrium. In the thermal activation region, metal flow in

discontinuous fashion. This is related to the probability that thermally-activated process cause dislocation to move from its initial equilibrium position to another state of equilibrium. In the dynamic region, metal will flow spontaneously in the continuous fashion. Significantly, the threshold stress corresponds exactly to the asymptote at which plasticity become more general macroscopic[2]. It simply means that yield surface can be identified by the MTS locus. In short, MTS is a yield stress in a broader spectrum. Not only does it define regions of macroscopically and microscopically equilibrium but also distinguish two different kinetics regimes.

Thermodynamically, the threshold can be related to the Helmholtz energy to overcome the obstacle (obstacle energy). From the basic thermodynamic principle (1), the free energy of activation can be scaled by using the ratio of applied stress (effective stress), τ^* to the threshold stress, $\hat{\tau}$ (obstacle stress)(2). This relationship is illustrated in the figure 3.

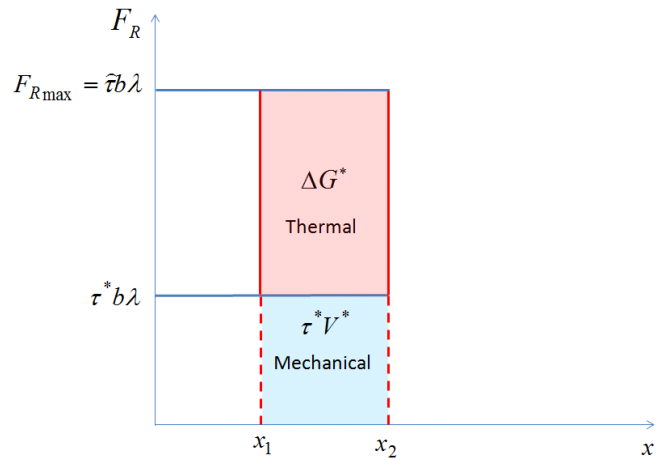


Figure 3. Obstacle energy represented by the sum of thermal and mechanical energy

$$\Delta F = \Delta G^* + \tau^* V^* \quad (1)$$

$$\Delta G^* \propto \Delta F \left(1 - \frac{\hat{\tau}}{\tau^*} \right) \quad (2)$$

The two important concepts above implies that in the range of experimental argument where the viscous or inertial flow is neither existent nor significant, flow kinetics can always be related to the threshold stress, more specifically to the ratio of effective and threshold stresses.

2.2 Constitutive Model Equation

Basic structure of the MTS constitutive model equation was underlain on the idea of Seeger who initially and explicitly split flow stress equation in crystalline solid into athermal and thermal components[30,31]. The idea was in

agreement with the facts that a critical resolved shear stress vs. temperature diagram of crystalline materials exhibited the region of decreasing stress with the increase of temperature and the plateau region (figure 3). It was assumed that the athermal component was not directly affected by temperature except through the temperature-dependence of shear modulus. The basic equation was formulated as (3)[31]:

$$\tau = \tau_{\mu(T)} + \tau^*(T, \dot{\gamma}) \quad (3)$$

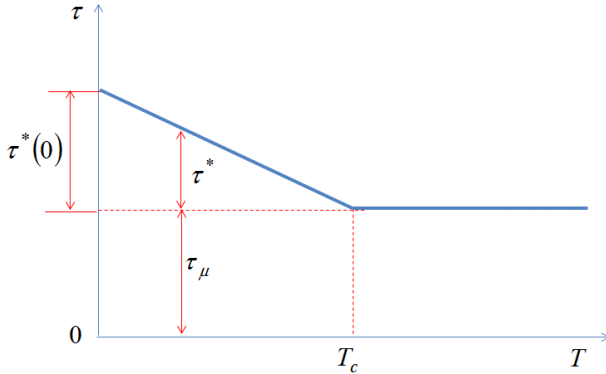


Figure 4. Athermal and thermal stress (after [30,31])

Division of athermal and thermal component was made particularly based on the way a dislocation line overcoming obstacles, whether it is aided with thermal energy or not. Two types of barriers for dislocation motion was assumed based upon the wavelength of its stress field. Long-range barrier is associated with those whose stress field is in the order of greater than 10 atomic diameter and short-range barrier is related to those less than 10 atomic diameter[31]. It is generally assumed that thermal process is effective only for the first type of barrier.

Athermal component of the flow stress was particularly elaborated in Conrad[31]. Temperature-dependence of elastic modulus, through which the athermal is in reality implicitly dependent on the temperature had been studied quite extensively. Comprehensive review on temperature-dependence of elastic constant, for example, was found in Varshni[32].

Of the heart of MTS model is the thermal part of the equation, in which MTS is used as a primary internal state parameter. The basic relation can be expressed in (4).

$$\frac{\tau^*}{\dot{\gamma}} = s(T, \gamma) \quad (4)$$

The scale factor, $s(T, \gamma)$, can be derived from the first principles of thermodynamics and Boltzmann statistics. Inserting (4) into (3), the basic flow stress constitutive equation can be expressed as (5).

$$\tau = \tau_{\mu(T)} + s(T, \gamma) \dot{\gamma} \quad (5)$$

3 MODEL THEORY IN SMALL VOLUME ELECTRONIC MATERIALS

3.1 Stress Phenomena in Small Volume Electronic Materials

Although the functional characteristic of electronic materials is not to carry mechanical loading, the importance of stress-related properties cannot be overlooked. In electronic packaging material stress is generally produced by an interaction of a small volume material with its surrounding. The interaction stress may be generated by the difference of CTE, lattice orientation as well as structure evolution and bonding modification at the interface/surface. In a particular situation, the stress causes excessive deformation, produce void and hillocks, or create cracks and delamination, which may result in a structural failure.

In small dimensional structure, this problem is much intensified since a tiny discontinuity or excess is sufficient to produce a functional failure. Failure of a sub functional unit can result in the breakdown of the entire system. We learned from these causes and effects that stress-related phenomena are very important in term of reliability.

In thin film metal interconnects, a plastic response of polycrystalline thin film metal to loading can be identified as a root phenomenon. Characterizing when a polycrystalline thin film metal starts to flow under different range of arguments, therefore, is not only interesting and challenging but also important in terms of design, fabrication, use, and performance.

Efforts have been made to study and model the temperature-dependence of flow stress in small volume materials. Figure 6 present cases in point of experimental data collected and selected from available published sources. It can be generally learned from the plots a discrete space of interest in small volume materials. Difference of average behavior between bulk and small volume material within the same range of experimental arguments are presented in figure 7 in the form of isostress and isostrain deformation contours.

Although the difference can be intuitively recognized, efforts are demanded to understand and capture this unique phenomena in a more quantitative way. One of the key challenges in understanding and describing such phenomena, along with multiple mechanisms, intertwined process, and altering structure, is a complex interaction of dimensional and micro-structural constraints.

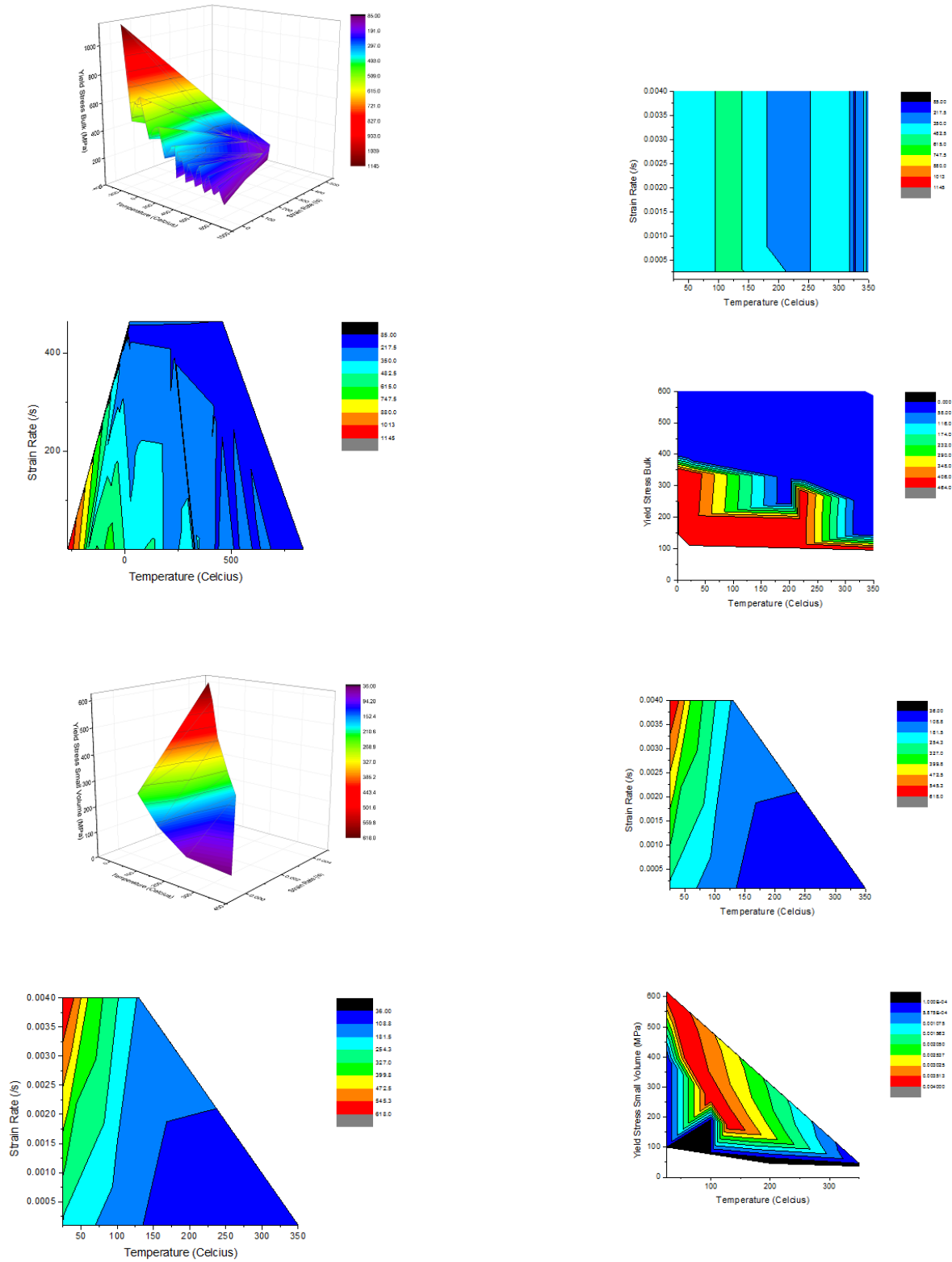


Figure 6. Typical temperature-dependence of flow stress in bulk Cu (top) and small volume Cu (bottom). Data from [33-39]

Figure 7. Isostress (top) and isostrain (bottom) deformation contours of bulk and small volume Cu in the same range of arguments. Data from [33-39]

3.2 Characteristics of Small Volume Materials

Dimensional Constraint

In general, the reduction of dimension will result in the constriction of slip plane along with modification of dislocation length, source, and motion. In thin film materials where the dimension is much constrained in transverse direction, thickness becomes a key size parameter.

Behavior of thin film is distinguished by threading dislocations whose length and characteristic are much dependent on the film thickness (figure 6). Bulk sources of dislocation may exist but not effective due to the loss of dislocation from the surface. Generation and multiplication of dislocations are unusable if not impossible. Surface nucleation of dislocation may become more important in such a condition of dislocation scarcity.

Dislocation movement in thin film is generally represented by dislocation channeling model (figure 6). Mean free path of dislocations is much affected by the geometry constraint. In the absence of other boundaries, dislocation mobility (the rate of dislocation motion and misfit dislocation formation) is dictated by the thickness and the density of available dislocations.

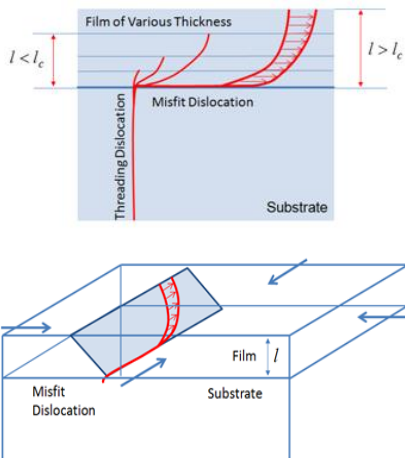


Figure 6: Dislocation threading and channelling (after [40-42])

Increased surface to volume ratio will normally enhance the sensitivity of surface/interface features. Minute cracks and coarse surface, for instance, become more effective stress raisers. In film on substrate this is even mixed up with the interaction stress. Effects of interface and adjacent material become more articulated. Emission, absorption, deposition, arrangement, and interaction of dislocation at the surface/interface may also modify the mechanism. Roles of dislocation sources, misfit dislocations, and obstacles at the interface become more important.

Microstructural Constraint

Microstructures of small volume materials are essentially the characteristics of their process history. A non homogeneous composite structure involving porous structure, densely packed fibrous grain, columnar grain, recrystallized grain, are usually produced by the sole process. Only small grained polycrystalline or nanocrystalline structures are isotropic. Some of them are finely patterned and highly textured. Epitaxy is usually single crystal, textured, columnar, and anisotropic. Important size parameters related to microstructural constraint involves grain size as well as the spacing and radius of obstacles. Effect of grain boundary is assumed to be similar to those in bulk but their level and quality may be atypical due to the dimensional constraint. In general, the effects may be discriminated by the ratio of grain sizes to thickness (d/h).

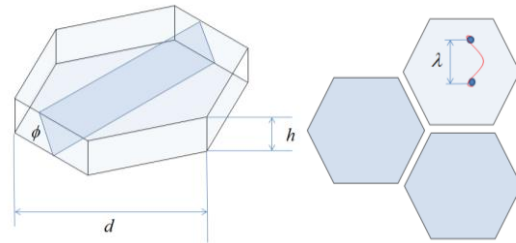


Figure 7. Constraint and size parameter (after [43,44])

Homogeneity

Apparent behaviour of bulk crystalline solid is generally accepted as an average of local responds over a large section of material. In small volume materials, quite differently, it is only made over a small segment. While similar basic mechanism is taking place at confined dimension, the averaged responds and hence the observed macroscopic behaviour, is more likely modified by geometry and micro-structural constraints. Control of high surface to volume ratio and the sensitivity of surface/interface conditions practically become more articulated. Moreover, these are complicated with a characteristic structure. Therefore, it is reasonable to assume that the apparent behaviour of small volume materials does not always follow the scaling law developed in bulk materials.

Observable Facts

Model of dislocations in small volume materials have been to some extent proven by in situ TEM observation. Model of dislocation channeling as well as additional formation and pre-existing network at near confined surface had been confirmed in single crystal thin film on a single crystalline substrate.

Threading dislocations had been observed in thin film. From the early study of Pb thin film on substrate, it was initially reported that the dislocation glide was confined in each grain by grain boundary and materials interfaces (surface oxide and substrate)[45,46].

One of the strong arguments of channeling was obtained from epitaxial Al on a single crystal substrate[47]. Figure 8 presented an advanced threading dislocation with a dislocation segment left in its path. Interaction between channeling dislocation and interface obstacles was associated with the stand-off of the dislocation from the interface.

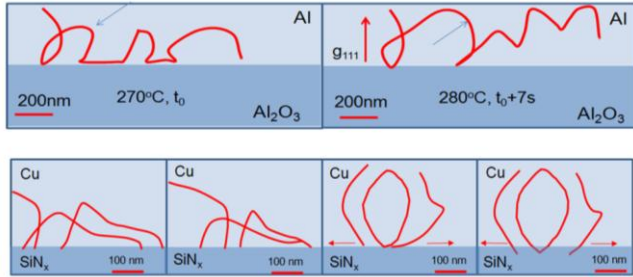


Figure 9. Cross sectional TEM image sketch of a 350nm thick epitaxial Al film deposited on a single crystal (0001) alpha-Al2O3 substrate during heating (upper side, after[48]). Dynamic cross sectional TEM images sketch a 450nm thick polycrystalline Cu on a Si substrate with a-SiNx diffusion barrier during heating (bottom side, after[47])

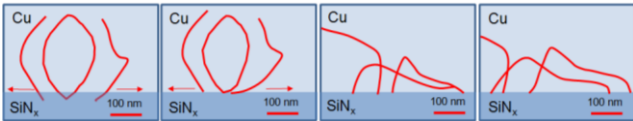


Figure 10. Dynamic cross sectional TEM images sketch a 450nm thick polycrystalline Cu on a Si substrate with a-SiNx diffusion barrier during heating illustrate interface as a sink rather than source of dislocation that hindered movement (after[47])

It was observed that dislocations, slipped on a plane inclined to the interface, had generated slip traces or dislocations lines at the interface[50]. The contrast of the slip traces in the TEM images, however, was particularly fading when they were irradiated below their threshold damage. The fading of the contrast in association with the strain field indicated the core spreading of the interfacial dislocation[50]. This spreading, which is schematically illustrated in figure 11 is a characteristic phenomenon at the interface of crystalline/amorphous, at which the substrate has no simple crystallographic relation with the film and hence can be considered as continuum[49].

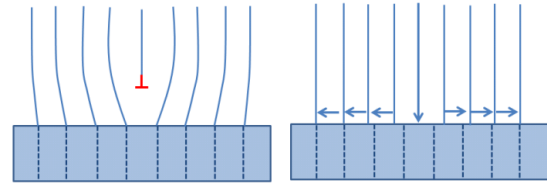


Figure 11. Sketch of dislocation core spreading (after [49])

It was believed that the metal film/amorphous interlayer interface pulled dislocations to the ground[44]. In general, the characteristics of the interface between film and substrate play very significant roles in determining dislocation glide in a thin film on substrate. When the interfacial bonding is sufficiently stiff to constrain dislocation segment from the escape, interfacial dislocation segments will be deposited. On the contrary, when the interfacial bonding is weak, interfacial dislocation segments hardly ever drag long segments of interfacial dislocations[52].

The facts that dislocations behaved differently in polycrystalline had challenge the established assumptions and models of work hardening. One of the possible scenarios, as has been previously discussed, is to assume the “bulk-like” behavior of film, meaning that the flow stress is controlled by the pinning of internal dislocation whose density increase with the decrease of the thickness. On the other hand, it was also known that the number of events decreased with the decrease in thickness film[51].

One of the most significant evidence that supports the MTS theory is the difference of dislocations image observed at two temperature regimes. Dehm et.al.[47], for instance, had revealed the difference image of Cu grains at two significantly different temperature representing the discontinuous and continuous motion of dislocation line. Evidence of “Jerky” motion, observed at lower temperature had been related to the characteristic of dislocation mediated, thermally activated, obstacle controlled plasticity. Kobrinsky and Thompson[29] believed that such a phenomenon was a solid evidence for thermal activation process of dislocation overcoming obstacles. This solid argument was also supported by a characteristic volume of activation. Based on the study in Ag, it was believed that the assumption should work on other FCC thin film including Cu and Au.

Timeline

Efforts have been made to study and model the response of small volume electronic materials subjected mechanical loading at various temperature and experimental arguments. Most of the model was developed to study the behavior of thin film metal on substrate subjected to thermal cyclic loading. Some model, for instance in[46], are purely empirical, from which no physical mechanism can be interpreted. Discussion will be limited to physical mechanistic models related to the MTS model theory.

Development milestone of such models is depicted in figure 12.

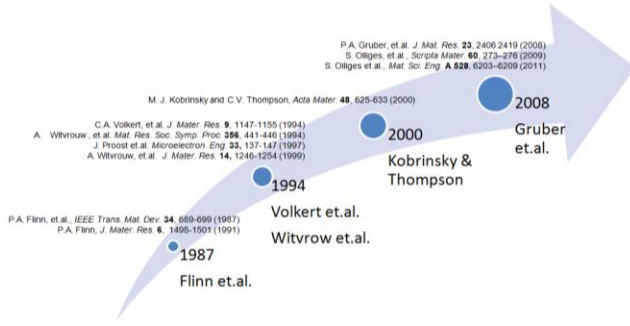


Figure 12. Time line of physical models in small volume electronic materials

Evaluation of the Model

Flinn et.al.[22] developed model to explain stress-temperature relationship in small volume Al attached to substrate. A simple critical stress equation for a given strain rate and temperature was developed by defining a hypothetical temperature of zero stress (6).

$$\sigma = \hat{\sigma} \cdot \left\{ 1 - \frac{T}{T_F} \right\} \quad (6)$$

$$T_F = \frac{-\Delta F}{k} \cdot \ln \left(\frac{\dot{\epsilon}_P}{\dot{\epsilon}_0} \right) \quad (7)$$

where:

$\dot{\epsilon}_P$ = plastic strain rate

$\dot{\epsilon}_0$ = characteristic constant

A general stress time relationship was formulated as (8)

$$\sigma = B \left\{ \ln \frac{B}{R} - \ln \left[t + \left(\frac{B}{R} \right) \cdot \exp \left(\frac{-\sigma_0}{B} \right) \right] \right\} \quad (8)$$

where:

$$B = \frac{kT\hat{\sigma}}{\Delta F} \quad (9)$$

$$R = \left[\frac{E}{(1-\nu)} \right] \cdot \dot{\epsilon}_0 \cdot \exp \left(\frac{-\Delta F}{kT} \right) \quad (10)$$

σ_0 = initial stress

Similar model was developed for Cu with modification of prior assumption[23].

$$\frac{T}{T_F} = \left(1 - \frac{\sigma}{\hat{\sigma}} \right) \frac{1}{\left(1 - C \ln \left(\frac{\sigma}{\hat{\sigma}} \right) \right)} \quad (11)$$

where:

$$C = \frac{n}{\ln \left(\frac{\dot{\epsilon}_P}{\dot{\epsilon}_0} \right)} \quad (12)$$

Explicit solution for $\sigma \left(\frac{T}{T_F} \right)$, however, was not available.

Equivalent equation for Al thin film on substrate was also formulated by Volkert et.al.[24]

$$\sigma = \frac{-kT\hat{\sigma}}{\Delta F} \ln \left[\exp \left(\frac{-\Delta F \sigma_0}{kT\hat{\sigma}} \right) + A' e^{\Delta F/kT} \frac{\Delta F}{kT\hat{\sigma}} t \right] \quad (13)$$

Similar relationship was also derived and applied by Proost and Witvrow et.al.[25-27] for Al-based alloy thin film:

$$\sigma = \frac{-kT}{a} \ln \left[\exp \left(\frac{-a\sigma_0}{kT} \right) + \frac{ac}{kT} t \right] \quad (14)$$

where:

$$a = \Delta F / (\sqrt{3}\hat{\sigma}) \quad (15)$$

$$c = \left[(\dot{\gamma}_0 / 2\sqrt{3}) (E/(1-\nu)) \right] \exp(-\Delta F/kT) \quad (16)$$

The three equations have been derived in much similar ways and also have comparable characteristics. Firstly, they were derived for thin film attached on the substrate. Secondly, the equations contain elasticity parameter of the metals (Young Modulus and Poisson Ratio) which reflects the assumption that elastic strain equal to total strain (purely elastic strain assumption). Thirdly, the equations are a function of time in addition to temperature. These equations are basically the rate equations of stress which characterize the stress relaxation of the metal with the change of temperature. The physical basis of the equations can be seen from the Helmholtz energy that represents the threshold energy to overcome obstacle without the application of mechanic stress. One of the disadvantages is that the equations do not represent clearly the athermal part of the stress. Moreover, they do not include dimensional and micro-structural constraints parameters.

Although they had been derived to describe stress relaxation of thin film on substrate, it was assumed that it could also be used to characterize film and predict stress. The fact that the relationship was able to describe the phenomena had confirmed to some extent that the assumed ruling mechanism was acceptable.

A quite difference form of model equation was presented by Kobrinsky and Thompson[29].

$$\sigma = \sigma_a + \left(\frac{C}{\Delta v^*} \right)^{\frac{M_T}{M_0}} \quad (16)$$

where:

Δv^* = volume of activation

C = parameter C

M_0 = biaxial elastic modulus for a (111) texture at 0 K

M_T = biaxial elastic modulus for a (111) texture at measurement temperature

The model had separated the athermal and thermal component of flow stress. It was also noticed that the volume of activation was explicitly expressed in the equation as a primary state parameter. Referring to Kobrinsky and Thompson[29], this parameter provides information about the geometry of plastic incident and obstacle spacing. With respect to the athermal part, Kobrinsky and Thompson[29] had revisited two model theories involving energy balance and work hardening model. It was believed that the interaction of moving threading dislocation with dislocations at the substrate/film interface was basically athermal. The amount of energy required to transfer the threading dislocation was largely reserved in the form of core and elastic energy. This energy would be spent at the interface as dislocation moving. Energy balance model, on the contrary ignored the interaction of the moving dislocation with others. The only aspect considered in the athermal component was temperature-dependent elastic modulus.

More recently, Gruber et.al.[28] had proposed a stress equation as a function of elastic modulus of the film and the experimental strain rate (17).

$$\sigma = \frac{\dot{\epsilon}_x}{C} - \left(\frac{\dot{\epsilon}_x}{C} - \sigma_0 \right) \exp \left(- \frac{EC\epsilon}{\dot{\epsilon}_x} \right) \quad (17)$$

where:

C = parameter to facilitate further calculation

$\dot{\epsilon}_x$ = experimental strain rate

One of the important benefits of the equation is that it can be described by a familiar figure of merit function from which key parameters can be extracted.

$$y = a + b \exp(-cx) \quad (18)$$

The a parameter is related to grain boundary width times the grain boundary diffusivity. Although the equation had involved both geometry and micro-structural constraint, it seemed to work in the regime governed by diffusional creep.

4 CONCLUDING REMARKS

MTS model has a solid physical foundation involving dislocation-mediated, thermally-activated obstacle-controlled local mechanisms. Structure of the model basically consists of athermal and thermal components of flow stress. Threshold stress associated with obstacle energy is used as a key state parameter. Physically, it is a limit stress, beyond which dislocations could not find any thermal mechanical state of equilibrium and no thermal activation needs to be existence. Thermal part of the equation, where the state parameter is attenuated by a factor containing Boltzmann probability function, represents the thermal activation process of dislocation overcoming obstacles. The model could be basically extended beyond the boundary by involving micro-structural evolution parameter but normally set under the assumption of constant microstructure within the range of arguments where diffusion is neither significant nor existence.

Application of MTS theory in small volume electronic can initially performed by assuming that similar local mechanisms exist in both bulk and small volume materials. There are some points, however, need to be addressed specifically. One of the most important issues is the fact that the observed mechanical behavior is a statistical average of local non-homogeneous responds of material. In addition to the in-homogeneity issue, geometrical and micro-structural constraints significantly influence the formation, movement and interaction of dislocation. The mechanism in small volume is much influenced by those intertwined constraints. Thickness and grain size are two important size parameters in addition to obstacle spacing and size. Dislocation mechanism model establish in small volume material involving threading dislocation, misfit dislocation, critical length, dislocation channeling, additional formation of dislocation and interaction with network of dislocation at the interface, piled up back stress and core spreading.

Constitutive model equations of stress have also established in small volume materials. As far as the authors are concerned, there is no single equation of stress that explicitly involves mechanical threshold as a state parameter and separates the athermal and thermal components of stress simultaneously as it is a common practice in bulk MTS model. Most of equations are developed based on the behavior of thin film attached on surface and presented in term of rate equations of stress in which temperature-dependence of yield stress are not presented explicitly. More efforts must be made to

reinterpret the model established in bulk materials in order to successfully describe the behavior at small volume and extract sound parameters that help to explain the responsible mechanism.

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