

# THE EFFECTS OF A STRESS-DEPENDENT MOBILITY ON INTERFACIAL STABILITY

P.W. VOORHEES

*Department of Materials Science and Engineering Northwestern University, Evanston, IL 60093*

M.J. AZIZ

*Division of Engineering and Applied Sciences, Harvard University, Cambridge MA, 02138*

A linear morphological stability analysis of a planar interface in interface reaction-controlled growth is presented. We allow the mobility of the interface to be a function of the elastic stress, and find that the stress dependent mobility can be either stabilizing or destabilizing. We find that a stress dependent mobility can be the dominant factor leading to instability at small stresses. The predictions are compared to the recent experiments by Barvosa-Carter et al. In agreement with their conclusions, we find that the experiments were performed in a parameter range where the stress dependent mobility is the dominant cause of instability. The prediction of the critical wavenumber for and amplification rate of the instability is consistent with the experimental results.

## 1 Introduction

The stability of interfaces in elastically stressed solids has been the subject of great interest recently. A reason for this is the connection between the stability of a surface of a thin film that is deposited on a substrate and the subsequent evolution of the film morphology. It is clear that an elastic-stress driven morphological instability results in a planar film undergoing a transition to islands for a review, see [1]. Controlling this instability is thus central to the production of novel electronic and optoelectronic devices because the properties of such devices depend upon the morphology of the film. For example, under some circumstances nanoscale islands called “quantum dots” form which can exhibit unique electronic properties as a result of carrier confinement within the small volume of these dots. However, devices with well characterized properties require that the size of the dots be relatively uniform.

To date, much of the work has been focused on the interfacial instability that is driven by the elastic energy in the system. This instability, identified by Asaro and Tiller (AT) [2] and many others [3, 4, 5, 6], follows from the dependence of the chemical potential on the elastic energy density. Consider a solid under uniaxial stress in contact with a fluid. Although the stress in the unperturbed or basic state is uniform, this is not the case when the interface is nonplanar. When the surface is perturbed the stress is concentrated at the troughs, in analogy with the stress concentration of a crack, and relieved at the peaks. Because the chemical potential is linearly proportional to the elastic energy density, the chemical potential of an atom is higher at the trough than at the peak and atoms thus move from the trough to the peak tending to make the peaks higher and the troughs deeper. Opposing this, of course, is surface energy that would tend to move atoms in the

opposite direction and lead to a planar surface. The stress concentration in the trough appears in both compression and tension, and thus the AT instability is independent of the sign of the applied stress.

Given the basic mechanism, the development of the elastic energy driven, or AT, instability must also depend on the mode of mass transport. Thus, the elastic energy driven instability has been examined in the context of bulk diffusion [2], surface diffusion [4, 5, 6] and evaporation-condensation [4, 7]. The evaporation-condensation model is, essentially, a completely kinetically controlled interface in which the rate-limiting mass transport process is atom motion across an interface between a solid and another solid, a vapor, or even a liquid. In this case the velocity of the interface is linearly proportional to the difference between the chemical potential at the surface and that in the parent phase. The type of mass transport process does not alter the critical wavenumber of the instability, it alters only the dependence of the growth rate of the instability on wavenumber.

The elastic energy driven instability in alloys has also been examined. In an alloy the chemical potential couples to both the stress and the composition. Moreover, as stresses can result from gradients in composition in the solid, the composition and stress are also coupled through the continuum mechanics. As before, however, the evolution of the surface occurs by gradients in the chemical potential for surface diffusion driven mass transport [8, 9, 10, 11], or by the difference between the chemical potential and that in a fluid for evaporation and condensation mass transport [12]. These compositionally generated stresses can act to either enhance or prevent the usual AT instability [9, 11, 12]. In addition, compositionally generated stresses can cause an instability even in the absence of an applied stress [10, 11, 12, 13].

All of this work, however, assumes that the relevant transport coefficient, the surface diffusion coefficient for surface diffusion limited interface motion, or the interface mobility for interface reaction-limited interface motion, is independent of elastic stress. Atomistic simulations have shown that the surface diffusion coefficient can be linearly proportional to the elastic stress on the surface. However, recent experiments by Barvosa-Carter et al (BC) have shown that a stress-dependent interfacial mobility alone can generate an interfacial instability when the transformation is interface-limited [14]. The mechanism for the kinetic instability follows from the observation that the change in the mobility with stress in certain cases can be linearly proportional to the stress. Consider a basic state consisting of a moving planar interface under uniform uniaxial stress. Assume that the stress compressive, i.e. *negative*, and that the mobility increases with increasingly positive stress. Then at the troughs, where the stress is the more negative than that of the planar state, the mobility is smaller than that of the planar interface. Conversely at the peaks, where the stress is larger than that of the basic state, the mobility is larger than that of the planar interface. Thus the troughs grow slower and peaks grow faster than the moving planar interface. In this case the stress-dependent mobility is destabilizing.

Moreover, experiments have shown that the amorphous to crystalline (a-c) transformation in Si is an ideal system in which to explore this kinetic instability. The stress dependence of the interfacial mobility has been measured [15]. The

mobility,  $M$ , was found to depend upon stress,  $\sigma$ , as

$$M(\sigma_{ij}) = \bar{M} \exp(V_{ij}^* \sigma_{ij} / kT) \quad (1)$$

where  $V_{ij}^*$  is the activation strain [15],  $\bar{M}$  is the mobility of the interface in the absence of stress,  $k$  is Boltzmann's constant and  $T$  is temperature. For the Si(001) a-c interface,  $V_{11}^* = V_{22}^* = 0.14\Omega$ , and  $V_{33}^* = -0.35\Omega$ , where  $\Omega$  is the atomic volume, and the off-diagonal elements are zero. The transformation is completely interface reaction-limited, and the other relevant materials parameters have been measured. It is possible to carefully control the morphology of the initial interface using lithography. Finally, because,  $V_{11}^* > 0$  a compressive uniaxial stress  $\sigma_{11} < 0$ , should result in an interfacial instability. Such an instability was observed by BC.

BC analyzed the instability using a fully nonlinear numerical approach. This approach has the advantage of being able to follow the instability into the strongly nonlinear, large-amplitude, regime and confirmed that the instability was, in fact, responsible for the nonplanar morphologies observed in the experiments. However, because the analysis was numerical, it was difficult to establish the relationship between the various materials and processing parameters that control the instability. Grinstein, Tu and Tersoff [16] examined analytically the effects of a stress dependent mobility in the context of thin film oxidation. The motion of the interface in this case is controlled by both interface mobility and long range diffusion. Given the complexity of the problem, however, they did not focus on the reaction-limited case, and their treatment of long-range diffusion is controversial [18, 19]. Yu and Suo have recently considered the reaction-limited case as well, focusing on the two-dimensional patterns that might be formed on the interface [17].

We present a linear stability analysis of a planar interface where the motion of the interface is interface reaction-limited. We allow for a stress dependent interfacial mobility and examine its effect on the stability of the interface. Using the materials and processing parameters of the BC experiments, we make a comparison between theory and experiment with no adjustable parameters. However, we assume that the amorphous phase into which the crystal grows is stress-free. Extrapolation of the known [20] viscous blow behavior of a-Si indicates that although there is a significant amount of relaxation in the BC experiments, it does not go to completion in the experimental time scales. We also assume  $M$  to be independent of crystallographic orientation. Although the mobility of the (001) interface is known to be 20 times faster on (001) than on (111),  $M$  at (001) is an extremum due to crystal symmetry. Nevertheless, we shall neglect this orientation dependence. The effects of growth kinetic anisotropy and flow in the a-Si layer on the instability are currently under investigation.

## 2 Theory

The crystal grows into the amorphous phase at a constant velocity  $v$  in the positive  $x_3$ -direction where  $x_3$  is parallel with the [001] direction. All quantities except stress and activation strain are assumed to be isotropic.

The motion of the a-c interface is kinetically limited, thus in the limit of small

driving forces,  $v$  is given by [14, 15]

$$v = -M(\boldsymbol{\sigma}) [\Delta G_v + W_{pV} + W(\boldsymbol{\epsilon}) + \gamma\kappa] \quad (2)$$

where  $M(\boldsymbol{\sigma})$  is the stress-dependent interface mobility,  $\boldsymbol{\sigma}$  is the stress in the small strain approximation, the chemical driving force for crystallization  $\Delta G_v^0$  is the Gibbs free energy change ( $\Delta G_v^0 < 0$ ) per unit volume of growing phase upon crystallization of stress-free material at a planar interface,  $W_{pV}$  is the  $pV$  work done on the system by external forces per unit volume crystallized,  $W(\boldsymbol{\epsilon})$  is the elastic strain energy,  $\boldsymbol{\epsilon}$  is the small strain tensor,  $\gamma$  is the interfacial energy,  $\kappa$  is the mean curvature of the interface (reckoned positive for a spherical crystal). The elastic strain energy is assumed to be negligible in the parent phase and we will neglect  $W_{pV}$ . Since the transformation is isothermal and occurs in a pure material, the only field equation that must be solved is that for the elastic stress.

The stress field is found by requiring that the crystal is in elastic equilibrium,

$$\sigma_{ij,j} = 0 \quad (3)$$

where summation of repeated indices from 1 to 3 is assumed. The stress is linearly related to the strain,

$$\sigma_{ij} = \frac{E}{1+\nu} \left( \frac{\nu}{1-2\nu} \epsilon_{kk} \delta_{ij} + \epsilon_{ij} \right) \quad (4)$$

where  $E$  is Young's modulus and  $\nu$  is Poisson's ratio. The strain is related to gradients in the displacement  $u_j$ ,

$$\epsilon_{ij} = (u_{i,j} + u_{j,i}) / 2 \quad (5)$$

Using Eq. (5) and Eq. (4) in Eq. (3) yields Navier's equations for the displacement field in the crystal,

$$(1 - 2\nu) u_{i,kk} + u_{k,ki} = 0 \quad (6)$$

The a-phase is assumed to be stress-free, and thus along the a-c interface,

$$\sigma_{ij} n_j = 0 \quad (7)$$

where  $n_j$  is the normal to the interface pointing into the a-phase. Finally, the elastic energy density is given by,

$$W = \sigma_{ij} \epsilon_{ij} / 2 \quad (8)$$

In the unperturbed, or basic state, the interface is planar and moving at a constant velocity into the amorphous phase. The applied stress in the crystal is biaxial and of magnitude  $\sigma_{11}^0 = \sigma_{22}^0 = \sigma_a$ . The stress-free boundary condition, Eq. (7), implies  $\sigma_{i3} = 0$  throughout.

To examine the linear stability of the moving planar interface located at  $x_3 = 0$ , we perturb the height of the interface in the  $x_3$ -direction,  $h = h(x_1, x_2)$  and all

other quantities in normal modes,

$$\begin{pmatrix} h \\ v \\ u_i \\ M \\ W \\ \kappa \end{pmatrix} = \begin{pmatrix} 0 \\ v^0 \\ u_i^0 \\ M^0 \\ W^0 \\ 0 \end{pmatrix} + \delta \begin{pmatrix} \hat{h} \\ v^1 \\ u_i^1 \\ M^1 \\ W^1 \\ \kappa^1 \end{pmatrix} \Phi \quad (9)$$

where  $\Phi = \exp(i\mathbf{q} \cdot \mathbf{x} + rt)$ ,  $\mathbf{q}$  and  $\mathbf{x}$  are the wavevector and position vector, respectively, in the plane of the interface,  $\delta \ll 1$  and  $t$  is time. If  $r > 0$  the perturbation grows and if  $r < 0$  the perturbation decays. If  $r$  is imaginary then an oscillatory instability is possible.

Using Eq. (9) in Eq. (2) we obtain in the basic state,  $\delta = 0$ ,

$$v^0 = -M^0 (\Delta G_v^0 + W^0) \quad (10)$$

The perturbed velocity is

$$V^1 = -(\Delta G_v^0 + W^0) M^1 - M^0 W^1 - M^0 \gamma \kappa^1 \quad (11)$$

We now need to find  $M^0, M^1, W^0, W^1, V^1$  and  $\kappa^1$ .

The elastic strain energy density in the basic state follows from the stresses given above,

$$W^0 = \sigma_a^2 / Y \quad (12)$$

where  $Y = E/(1-\nu)$ . The strain energy density in the perturbed state is determined by solving Eq. (6) with the boundary condition Eq. (7) and the condition that the  $u_i^1(x_3) \rightarrow 0$  as  $x_3 \rightarrow -\infty$  [21]. Using these displacements in Eq. (5) and substituting into Eq. (4) and using Eq. (8) yields the magnitude of the perturbed elastic strain energy density along the interface [12],

$$W^1 = 2\hat{h}(1+\nu)q\sigma_a^2/Y \quad (13)$$

where  $q = |\mathbf{q}|$ . In addition, the magnitude of the trace of the stress is

$$\sigma_{kk}^1 = -2\hat{h}(1+\nu)q\sigma_a. \quad (14)$$

The mobility of the interface in the basic and perturbed states,  $M^0$  and  $M^1$  respectively, can be found by using Eq. (1)

$$M(\boldsymbol{\sigma}) = \bar{M} \left( 1 + \frac{\sigma_{ij} V_{ij}^*}{kT} \right) \quad (15)$$

where we have assumed that  $\sigma_{ij} V_{ij}^*/kT \ll 1$ , Thus,

$$M^0 = \bar{M} \left( 1 + \frac{\sigma_{ij}^0 V_{ij}^*}{kT} \right) \quad (16)$$

$$M^1 = \frac{\sigma_{ij}^1 V_{ij}^*}{kT} \quad (17)$$

where  $\sigma_{ij}^0$  and  $\sigma_{ij}^1$  are the stresses in the basic and perturbed states, respectively.

To determine  $\sigma_{ij}^1 V_{ij}^*$ , we first rewrite it as,

$$\sigma_{ij}^1 V_{ij}^* = \sigma_{kk}^1 V_{11}^* + (V_{33}^* - V_{11}^*) \sigma_{33}^1 \quad (18)$$

because for an (001) surface of a cubic crystal,  $V_{11}^* = V_{22}^*$  [15]. From Eq. (7) we find  $\sigma_{33}^1 = 0$ . Using this and Eq. (14) in Eq. (18) yields

$$\sigma_{ij}^1 V_{ij}^* = -2\hat{h}(1 + \nu)V_{11}^* \sigma_a q. \quad (19)$$

Using the stress in the system with a planar interface,

$$\sigma_{ij}^0 V_{ij}^* = 2V_{11}^* \sigma_a. \quad (20)$$

Using these stresses in Eq. (16) and Eq. (17) yields

$$M^0 = \bar{M} \left( 1 + \frac{2V_{11}^* \sigma_a}{kT} \right) \quad (21)$$

and

$$M^1 = -\bar{M} 2\hat{h}(1 + \nu)V_{11}^* \sigma_a q / kT \quad (22)$$

Thus, if  $V_{11}^* > 0$ , a compressive stress decreases the mobility of the planar interface. For compressive stresses the perturbed mobility  $M^1$  is in phase with the interface shape. Therefore the mobility is the highest at the peaks and the lowest at the troughs.

The remaining terms in Eq. (11) are functions only of the shape of the interface. Using the definition of the curvature in cartesian coordinates, the form of the perturbed interface, and by taking the derivative with respect to time we obtain,

$$\begin{aligned} \kappa^1 &= \hat{h}q^2 \\ v^1 &= \hat{h}r \end{aligned} \quad (23)$$

Using Eq. (23), Eq. (21), Eq. (22), Eq. (12)-Eq. (14) in Eq. (11) yields the dispersion relation,

$$\begin{aligned} \frac{r}{\bar{M}} &= [(1 + \nu) \alpha \Delta G_v \sigma_a + 2\beta \sigma_a^2 + 3\alpha \beta \sigma_a^3] q \\ &\quad - (1 + \alpha \sigma_a) \gamma q^2 \end{aligned} \quad (24)$$

where  $\alpha = 2V_{11}^*/kT$  and  $\beta = (1 + \nu)/Y$ . The dispersion relation is linear in  $r$  and thus instability cannot be oscillatory.

The wavenumber at which the growth rate of the instability is zero,  $q_c$ , is found by setting  $r = 0$ .

$$q_c = \frac{(1 + \nu) \alpha \Delta G_v^0 \sigma_a + 2\beta \sigma_a^2 + 3\alpha \beta \sigma_a^3}{(1 + \alpha \sigma_a) \gamma}. \quad (25)$$

For  $q < q_c$ ,  $r > 0$ ; for  $q > q_c$ ,  $r < 0$ , see Figure 1. The wavenumber with the maximum growth rate,  $q_m$ , is given by  $q_m = q_c/2$ . Thus  $q_c$  determines the length-scale of the instability.

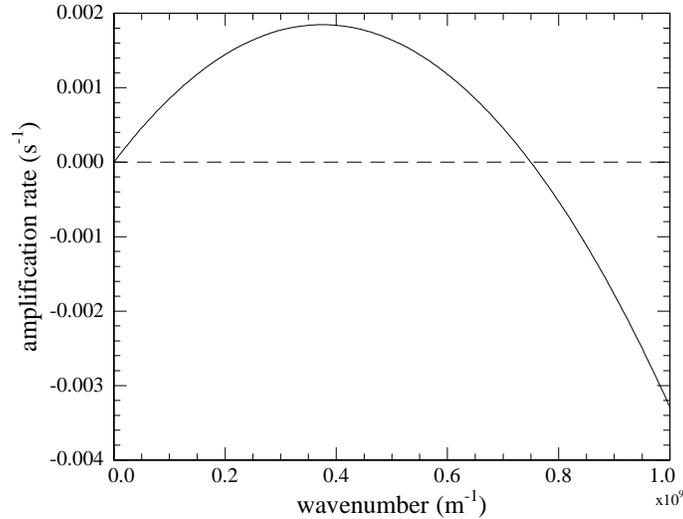


Figure 1. Amplification rate of the instability as a function of the wave number of the disturbance for the a-c Si transformation in the B-C experiments.

### 3 Discussion

The dispersion relation is quadratic in the wavenumber,  $q$ , with the terms related to the elastic stress being linear in  $q$  and the interfacial energy related terms going as  $q^2$ . Interfacial energy is always stabilizing, because by assumption  $\sigma_a V_{11}^*/kT \ll 1$ . Thus any instability is driven solely by the elastic stress. Note that there is no Mullins and Sekerka instability due to the absence of any long-ranged diffusion field. Of the terms that depend upon the applied stress, the term that is quadratic in the stress is the classical elastic energy driven Asaro-Tiller (AT) instability. In this reaction-limited case the AT instability is linear in  $q$  and not cubic as it is in the usual case where the instability develops by surface diffusion. The term that is linear in the applied stress can be either stabilizing or destabilizing depending on the sign of the product  $\alpha\sigma_a$ . If  $\alpha\sigma_a < 0$ , as in the a-c Si BC experiments, the stress dependence of the mobility itself can drive an interfacial instability. The term that is cubic in the applied stress is the AT instability modified by the stress dependent mobility.

Because  $\alpha\sigma_a < 0$  under compressive stresses for the Si(001) a-c interface, both the stress dependent mobility and the elastic energy promote instability. As shown in Eq. (24), both effects have the same scaling with the wavenumber and thus it can be difficult to determine which phenomenon is responsible for an experimentally observed interfacial instability. However, the elastic energy driven instability scales with stress in a different manner than the mobility driven instability. It is clear that for a sufficiently small applied stress, the stress-dependent mobility will dominate.

When  $\alpha\sigma_a < 0$ , this will occur when

$$\frac{(1 + \nu) \alpha \Delta G_v}{2\beta\sigma_a + 3\alpha\beta\sigma_a^2} \gg 1. \quad (26)$$

Using the values for the a-c Si interface under the conditions employed in the BC experiments we find that LHS of the inequality is approximately 12, and thus the instability of the interface is predicted to be due to the kinetic instability.

Conversely, if  $\alpha\sigma_a > 0$ , then the stress-dependent mobility is stabilizing. The critical stress,  $\sigma_a^c$  below which the stress-dependent mobility will stabilize the AT instability is found by setting the first term in the brackets of Eq. (24) to zero,

$$\sigma_a^c = -1/(3\alpha) + [1 - 3\alpha^2(1 + \nu)\Delta G_v/\beta]^{1/2}/(3\alpha) \quad (27)$$

For the a-c Si system, this yields a stress of 9.6 GPa. All tensile stresses below this value result in an interface that is stable against the AT instability. Preliminary experimental results [22] confirm this prediction. Thus in the a-c Si system very large (perhaps impossibly large) tensile stresses are required to overcome the stabilization due to the stress-dependent mobility.

The dispersion curve for the a-c transformation in Si is shown in Figure 1, where we have used  $\sigma_a = -0.5GPA$  [14],  $Y = 100GPA$ ,  $\nu = 0.25$ ,  $T = 520C$ ,  $\bar{M} = 4.52 \times 10^{-20}m^2/Js$  [23],  $\Delta G_v = -6.27 \times 10^8 J/m^3$  [24] and  $\gamma = 0.4J/m^2$  [25, 26]. The length scale of the interfacial instability is provided by the critical wavenumber, Eq. (25), and can be also be read directly from Figure 1. The critical wavelength,  $\lambda_c = 2\pi/q_c$ , of the instability is 8.5 nm. Thus perturbations with wavelengths in excess of 8.5 nm will amplify and those with wavelengths less than 8.5 nm will decay. The critical wavenumber of the AT instability,  $q_c^{AT}$  is,

$$q_c^{AT} = 2\sigma_a^2\beta/\gamma \quad (28)$$

For the BC experiment, this yields  $\lambda_c^{AT} = 2000nm$ . It is thus clear that the length scale of the kinetically driven instability can be quite different from that driven solely by elastic strain energy. The wavelength of the kinetically driven instability with the maximum growth rate is 17 nm. Unfortunately, the BC experiments were unable to determine  $q_c$  because, for reasons of experimental practicality, they started not with a planar interface but rather with an interface with a large initial sinusoidal perturbation with a wavelength of 400 nm. They found that a perturbation of this wavelength is unstable, and thus is consistent with this prediction. The experimentally observed amplification rate of such a perturbation is  $6 \times 10^{-5}/sec$  whereas the predicted amplification rate is  $1.6 \times 10^{-4}/sec$ . This is very good agreement considering the uncertainties in the materials parameters.

It has been suggested that the stress dependent mobility may also affect the evolution of elastically stressed interfaces when growth is not fully reaction-limited [14]. To investigate this possibility, we consider an interface evolving in a fully surface diffusion-limited regime in which the atomic mobility for diffusion,  $M_D$ , is stress-dependent. In this case the velocity of the interface is related to a surface gradient in the chemical or diffusion potential at the surface,

$$v \propto \nabla_s \cdot M_D(\sigma) \nabla_s (\Delta G_v + W_{pV} + W(\epsilon) + \gamma\kappa) \quad (29)$$

where  $\nabla_s$  denotes the surface gradient operator. Consider the case of a planar surface where the basic state stress is constant. Perturbing about this state gives

$$v^0 + v^1 \Phi \propto -\nabla_s \cdot (M_D^0 + \delta M_D^1 \Phi) \nabla_s (\delta W^1 \Phi + \delta \gamma \kappa^1 \Phi) \quad (30)$$

Because the elastic energy density and chemical driving force in the basic state are constants, they are not present in Eq. (30) and thus the effects of the perturbed atomic mobility,  $M_D^1$ , are on the order  $\delta^2$ . Therefore the stress-dependent mobility, or a stress-dependent surface diffusion coefficient, will not affect the linear stability of a uniformly stressed planar surface in purely surface diffusion-limited growth. However, a stress-dependent atomic mobility will clearly affect the nonlinear, or large amplitude, evolution of the surface. In addition, a stress-dependent interfacial mobility will continue to affect the linear stability in a situation of mixed diffusion and interface reaction control.

#### 4 Conclusions

The linear stability analysis for interface reaction-limited growth shows that when the mobility depends upon stress it can either lead to morphological instability or act to stabilize the planar interface. In addition, the more classical elastic energy driven AT-instability is also present. However, because the kinetically-driven instability scales linearly with the applied stress and the AT instability is quadratic in the applied stress, the kinetically-driven instability should dominate at low levels of stress. When the theory is applied to the a-c transformation in Si, we find that for the parameters used in the BC experiments the observed interfacial instability is a result of the stress dependent mobility. Finally the predicted critical wavelength and amplification rate of the kinetically driven instability is consistent with the experiments.

#### Acknowledgments

This research was supported by the National Science Foundation through DMR-9707073 (PWV) and DMR 98-13803 (MJA).

#### References

1. V.A. Shchukin and D. Bimberg, Rev. Modern Phys. **71**, 1125 (1999).
2. R.J. Asaro and W.A. Tiller, Metall. Trans. **3**, 1789 (1972).
3. M.A. Grinfeld, Dok. Akad. Nauk SSSR **290**, 1358 (1986).
4. D.J. Srolovitz, Acta Metall. **37**, 621 (1989).
5. J. Gao, Int. J. Solids Structures **28**, 703 (1991).
6. B.J. Spencer, P.W. Voorhees and S.H. Davis, Phys. Rev. Lett. **67**, 3696 (1991).
7. K.-S. Kim, J.A. Hurtado and H. Tan, Phys. Rev. Lett. **83**, 3872 (1999).
8. J.E. Guyer and P.W. Voorhees, Phys. Rev. Lett. **74**, 4031 (1995).
9. F. Leonard and R. Desai, Phys. Rev. B **56**, 4955 (1997).
10. V.G. Malyshekin and V.A. Shchukin, Semiconductors **27**, 1062 (1993).

11. B.J. Spencer, P.W. Voorhees, and J. Tersoff, in press.
12. J.E. Guyer and P.W. Voorhees Phys. Rev. B. **54**, 10710 (1996).
13. B.J. Spencer, P.W. Voorhees, S.H. Davis and G.B. McFadden, Acta Metall. Mater. **40**, 1599 (1992)
14. W. Baravosa-Carter and M.J. Aziz, Phys. Rev. Lett. **81**, 1445 (1998).
15. M.J. Aziz, P.C. Sabin, and G.-Q. Lu, Phys. Rev. B **44**, 9812 (1991).
16. G. Grinstein, Y. Tu and J. Tersoff, Phys. Rev. Lett. **81**, 2490 (1998).
17. H. H. Yu and Z. Suo, preprint.
18. V. P. Zhdanov, Phys. Rev. Lett. **83**, 656 (1999).
19. G. Grinstein, Y. Tu and J. Tersoff, Phys. Rev. Lett. **bf 83**, , (6)57 (1999).
20. A. Witvrouw and F. Spaepen, J. Appl. Phys. **74**, 7154 (1993).
21. B.J. Spencer, P.W. Voorhees and S.H. Davis, J. Appl. Phys. **73**, 4955 (1993).
22. J. F. Sage, W. Barvosa-Carter and M. J. Aziz (unpublished).
23. G.L. Olson and J.A. Roth, Mater. Sci. Rep. **3**, 1 (1988).
24. E. P. Donovan, F. Spaepen, D. Turnbull, J.M. Poate, and D.C. Jacobson, J. Appl. Phys. **57**, 1795 (1985).
25. N. Bernstein, M.J. Aziz, and E. Kaxiras, Physical Review B **58**, 4579 (1998).
26. C.M. Yang, Ph.D. thesis, California Institute of Technology, 1997.