

**Bachelor of Science (Research)
Material Science**

Final Year Project Report

**The influence of Convection on the
Evolution of Microstructure during
Solidification**

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Acknowledgements

Synopsis

Solidification is one of the major processing routes among the various materials processing techniques, where in there is phase transformation from a liquid to one or more solid-phases. The process gives rise to a variety of microstructures, arising out of dendritic, eutectic, peritectic and monotectic reactions. It is well known that properties of a material can be linked to its structure at the microscopic scale. Thus, in order to manufacture a cast product with desired properties, it is essential to determine the influence of the processing parameters on microstructural evolution. Here, modelling of complex microstructural evolution allows for the determination of the useful process \rightarrow structure and parameter \rightarrow structure correlations.

Being a first order phase transition, the solidification reaction involves a transfer of heat/mass across the interface between the solid and the liquid phases concomitant with diffusion in the bulk liquid and the solid phases. Any modelling technique describing this phenomenon will require, that in addition to the global boundary conditions controlling the heat/mass exchange, it self-consistently is able to integrate the transport processes at the moving interface between the evolving phases with those in the bulk.

Classically, such problems were treated with *Sharp-interface* methods, wherein, separate transport equations are solved in the respective bulk phases, while the appropriate (*Stefan-boundary*) conditions are imposed at the interface nodes which therefore need to be marked after each time iteration. The method becomes cumbersome in case of complex morphological evolution which is commonly encountered in most solidification reactions, where explicit interface tracking becomes computationally expensive and resolving sharp curvatures necessitates the use of finer meshes.

In this context, we resort to the *phase field method* - a state-of-the art technique, developed in the past three decades which obviates the need for tracking of the moving interfaces. Succinctly, the method describes evolution equations for order parameters varying smoothly between the various phases, thereby representing phase evolution. The various boundary conditions at the moving interface are self-consistently described in the transport equations describing solutal/heat/momentum transport, which are defined globally. The transport equations are coupled with the evolution equations for the order parameter. This class of methods has been applied to a variety of phase transformation reactions.

Presently, we have used the phase field method to investigate the influence of

convection on dendritic growth. There is a large practical interest in dendritic solidification as an overwhelming majority of metallic systems solidify in this manner. This is also an interesting problem in terms of mechanisms of pattern selection in a non-equilibrium systems and engenders a lot of theoretical interest.

The fundamental problem being addressed here is the selection of the scaling and growth velocities of the dendrites during solidification. The problem is well understood in the purely diffusive regime. However, in practical set-ups the dendrite growth is unavoidably influenced by buoyancy driven flows due to gravity. The effect is due to large scale transport of mass or heat by the fluid currents. This effect of fluid flow on the dendrite tip selection, however, has not yet been fully understood and is of great practical import.

In the present study we have endeavoured to throw some light this aspect of solidification. We have set up a phase field model based on the fundamental equations for solidification while incorporating flow in the fluid phase. We thus seek to directly simulate dendritic growth in a convective regime and gauge the effect of flow on the evolving morphology.

An introduction to Dendritic Solidification

Dendritic solidification is ubiquitous in material science and is of great practical and theoretical importance. The dendritic morphology can be characterised by growth of primary arms along well defined crystallographic directions with each primary arm also giving rise to self-similar secondary and tertiary arms. This self-similarity between the patterns at various length scales is a key characteristic of fractals and is observed commonly in nature. Dendritic microstructures result as a destabilisation of a planar or spherical growth front. A critical component giving rise to dendritic growth, where the primary growth occurs along well defined growth directions, is the surface energy anisotropy.

Physically, an understanding of dendrite morphology and growth would entail the knowledge of the scaling laws pertaining to - dendrite tip radius, inter dendritic and secondary arm spacings and how they change as a function of the processing conditions, alloy composition and properties. The requirement for understanding the scale and variation of these parameters lies in the fact that macroscopic parameters such as tensile strength, fracture toughness etc. are often determined by the scale of the microstructure.

There have been significant advances in understanding the physics of dendrite growth. The role of crystalline anisotropy in the selection of the dendrite tip radius and velocity is well explained by the microsolubility theory [1]. The theory has been validated by sophisticated microgravity experiments as well as phase field simulations in both two and three dimensions, that focussed on the purely diffusive regime.

On earth, however, dendritic growth is almost unavoidably influenced by buoyancy driven mass and energy advection in the melt phase. Melt flow introduces a new length scale in the problem, viz. the length of the convecting rolls, and breaks the symmetry of transport depending on the direction of the vector of gravity relative to the direction of growth. Self-Organising pattern formation in solidification begins to compete with the self-organisation of the convection patterns. On the one hand, convective transport of solute significantly alters the growth conditions. On the other hand, the magnitude of convection critically depends on solute gradients due to growth and on the friction of the convecting liquid melt against the growing solid-liquid interface[3].

A Phase Field Model for Solidification

3.1 The Phase Field Method

In the phase field method the system properties are expressed using a set of phasefield variables that vary continuously within a bound across all the phases in the system. The thermodynamics and transport properties of the system are coupled with or expressed in terms of the dynamics of these phasefield variables.

The time evolution of the phase field variables is given by a set of coupled partial differential equations, one equation for each variable. The equations are derived according to the principles of non-equilibrium thermodynamics. They are so chosen that the free energy of the system decreases with time and mass is conserved for all components. Likewise transport properties within the bulk as well as the interface are incorporated to be consistent with the dynamics of these phasefield variables. Numerical solutions of these PDE's yields the temporal evolution of the phasefield variables, which is a representation of the morphological evolution of the microstructure in the system.

The application of the phase field method starts with the creation of a functional which includes the material properties involving both the surface properties of the interfaces and the thermodynamic energies of the bulk phases in the system. A variational derivative of this functional with respect to any of the changing phasefield variables, gives us the driving force for the change.

As a first step to constructing a phase field model for solidification, we need a phase field variable that shall vary smoothly over liquid and solid phases. A suitable candidate is order parameter: ϕ . Physically, it corresponds to the probability amplitude of finding an atom at a particular location in the lattice and goes from unity in the solid phase to zero in the liquid phase where there is complete breakdown of order.

3.2 The Allen-Cahn Equation

Formally, the functional can be written in the following manner,

$$\mathcal{F} = \int_{-\infty}^{\infty} \left(f_o(\eta) + \kappa (\nabla \eta)^2 \right) dx. \quad (3.1)$$

where f is our constructed free-energy functional $f(\phi, \nabla\phi) = f_o(\phi) + \kappa(\nabla\phi)^2$.

If we were, to compute the time-derivative of the change in the free-energy functional, it can be written as,

$$\frac{\delta\mathcal{F}}{\delta t} = \int_{-\infty}^{\infty} \frac{\delta\mathcal{F}}{\delta\phi} \frac{\partial\phi}{\partial t} dx, \quad (3.2)$$

Given that ϕ is a non conserved variable, the simplest way to ensure that the functional is minimised in time, would be,

$$\frac{\partial\phi}{\partial t} = -M \frac{\delta\mathcal{F}}{\delta\phi}, \quad (3.3)$$

such that the time derivative of the free-energy functional now reads,

$$\frac{\delta\mathcal{F}}{\delta t} = - \int_{-\infty}^{\infty} M \left(\frac{\delta\mathcal{F}}{\delta\phi} \right)^2 dx. \quad (3.4)$$

Clearly, here, we are minimizing the energy. However, the rate of change is zero, when the functional is at an extremum given by $\frac{\delta\mathcal{F}}{\delta\phi} = 0$.

The dynamical evolution is then expressed as,

$$\frac{\partial\eta}{\partial t} = -M \frac{\delta F}{\delta\eta}. \quad (3.5)$$

The operator for variational derivative is,

$$\frac{\delta\mathcal{F}}{\delta c} = \left(\frac{\partial}{\partial c} - \nabla \cdot \frac{\partial}{\partial \nabla} + \nabla^2 \cdot \frac{\partial}{\partial \nabla^2} \dots \right) f \quad (3.6)$$

Applying this, we derive,

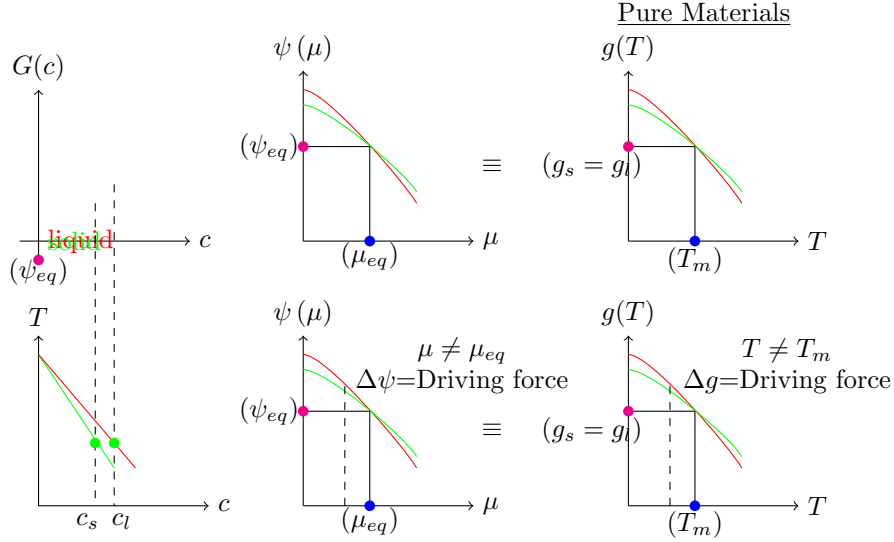
$$\frac{\partial\eta}{\partial t} = -M \left(\frac{\partial f_o}{\partial\eta} - 2\kappa \nabla^2 \eta \right). \quad (3.7)$$

This is the set-of equations describing Allen-Cahn dynamics and is applicable to a large class of materials transformations, for instance evolution of grain-boundaries, order-disorder transitions, solidification etc..

3.3 Modelling Alloy Solidification

We have in the previous section derived the Allen-Cahn equation, where in, if the two phases have the same free energy, a stationary interface with a defined width is created. However, if we were to add a term to the energy such that it tilts one of the energy levels with respect to the other, the phenomenological equation of motion, has so been constructed that the evolution of ϕ will be in a direction such that the phase with the higher energy will be consumed in favour of the phase with the lower energy.

This then is an apt recipe for modelling solidification. Formally this tilting, or rather the departure from equilibrium between the two phases, can be expressed as the *driving force* for phase transformation. In the case of solidification of binary isomorphous alloy, at a given under cooling, the driving force arises due to deviation of the diffusion potential from the equilibrium and results in solute rejection at the interface and expansion of the solidification front.



Assuming conditions of local thermodynamic equilibrium(LTE), the driving force for phase transformation in alloys is the difference of the grand-potentials of the solid and the liquid phases (or the difference of the chemical potentials of any of the chemical entities depending on whether we start from a Helmholtz free-energy density or a Gibbs-free energy density in the functional).

In the present description, we adopt a Helmholtz free energy density which then results in the driving force for solidification being described as $\Psi_l - \Psi_s$. This is analogous to the case of pure material solidification where the driving force is given by $g_l - g_s$. However, while the state-variable in the case of pure material solidification is the temperature T , it becomes the diffusion-potential μ in the case of binary alloy solidification. This analogy can be well appreciated from the diagram in Fig 3.3.

At equilibrium, the common-tangent construction gives the equilibrium compositions of the solid and liquid phases, which can be read from the equilibrium phase diagram. This same equilibrium can also be represented as the intersection of the grand-potential densities $\Psi(\mu)$ expressed as a function of the diffusion-potentials μ , where $\Psi_s(\mu_{eq}) = \Psi_l(\mu_{eq})$, in the same manner as $g_s(T_m) = g_l(T_m)$ for the case of pure materials. Away from equilibrium $\mu \neq \mu_{eq}$, $\Psi_l \neq \Psi_s$, and we have a driving force for phase transition given by the difference of the grand-potential densities.

Therefore, using this above analogy we can write down the driving force for phase transition for alloys purely in terms of the diffusion potential μ by performing a Taylor series expansion around equilibrium, until the first order as follows,

$$\Psi_s(\mu) = \Psi_s(\mu_{eq}) + \frac{\partial \Psi_s}{\partial \mu}(\mu - \mu_{eq}) \quad (3.8)$$

$$\Psi_l(\mu) = \Psi_l(\mu_{eq}) + \frac{\partial \Psi_l}{\partial \mu}(\mu - \mu_{eq}). \quad (3.9)$$

From the fig. 3.3 it is clear that we can express the grand-potential as $\Psi(\mu) = g(c(\mu)) - \mu c(\mu)$, which is the essentially the *Legendre transform* of g . Therefore, we can derive,

$$\frac{\partial \Psi}{\partial \mu} = \frac{\partial g}{\partial c} \frac{\partial c}{\partial \mu} - \mu \frac{\partial c}{\partial \mu} - c \quad (3.10)$$

$$= -c, \quad (3.11)$$

where we have used $\frac{\partial g}{\partial c} = \mu$. Clearly, then we can write the driving force for solidification $\Psi_l - \Psi_s$ as,

$$\Psi_l - \Psi_s = (c_s^{eq} - c_l^{eq})(\mu - \mu_{eq}), \quad (3.12)$$

where we have used $\Psi_l(\mu_{eq}) = \Psi_s(\mu_{eq})$ and $\frac{\partial \Psi_l}{\partial \mu} = -c_l^{eq}$ and $\frac{\partial \Psi_s}{\partial \mu} = -c_s^{eq}$. This relation can be transmitted to the evolution equation for the phase-field variable ϕ as,

$$\frac{\partial \phi}{\partial t} = -M \left(\frac{\partial f_o}{\partial \phi} - 2\kappa \nabla^2 \phi \right) - \underbrace{M(c_l^{eq} - c_s^{eq})(\mu - \mu_{eq})}_{\text{driving force}} \frac{\partial h(\phi)}{\partial \phi}. \quad (3.13)$$

The evolution equation of the coupled variable μ can be derived using mass conservation. Using the interpolation polynomial $h(\phi) = \phi^2(3 - 2\phi)$, which has the property $h(\phi) + h(1 - \phi) = 1$, the local composition can be written as,

$$c = c_s h(\phi) + c_l(1 - h(\phi)), \quad (3.14)$$

More rigorously, the above relation derives from the interpolation of the grand-potential densities $\Psi(\mu) = \Psi_s(\mu)h(\phi) + \Psi_l(\mu)(1 - h(\phi))$, and taking the derivative with respect to the diffusion potential leads us to the relation, $\frac{\partial \Psi}{\partial \mu} = \frac{\partial \Psi_s}{\partial \mu}h(\phi) + \frac{\partial \Psi_l}{\partial \mu}(1 - h(\phi))$.

Substituting the above relation between the composition and the diffusion-potential into the mass-conservation equation we have,

$$\frac{dc}{dt} = \nabla \cdot (M \nabla \mu) \quad (3.15)$$

$$\frac{dc}{dt} = \frac{\partial c}{\partial t} + \frac{\partial c}{\partial x} \frac{\partial x}{\partial t} + \frac{\partial c}{\partial y} \frac{\partial y}{\partial t} + \frac{\partial c}{\partial z} \frac{\partial z}{\partial t} = \frac{\partial c}{\partial t} + \vec{v} \cdot \nabla c \quad (3.16)$$

$$\frac{\partial c}{\partial t} = \nabla \cdot (M \nabla \mu) - \vec{v} \cdot \nabla c \quad (3.17)$$

$$\left(\frac{\partial c_s}{\partial \mu} h(\phi) + \frac{\partial c_l}{\partial \mu} (1 - h(\phi)) \right) \frac{\partial \mu}{\partial t} = \nabla \cdot (M \nabla \mu) - (c_s(\mu) - c_l(\mu)) \frac{\partial h(\phi)}{\partial t} - \vec{v} \cdot \nabla c \quad (3.18)$$

Here, the first term on the right hand side represents the diffusive flux, while the second term is the source term for phase transformation. The third, velocity coupled, term represents mass advection and accounts for influence of melt flow on the morphological evolution.

Thus, we are able to derive the evolution equations for solidification in alloys by starting with a phenomenological model for non conservative dynamics, based on an ansatz, and thereafter deriving the driving force in terms of the relevant state variables and then coupling it with the appropriate conservation equations.

The Equations

4.1 The phasefield equations

We have derived the evolution equation of the phasefield variable to be,

$$\frac{\partial \phi}{\partial t} = -M \left(\frac{\partial f_o}{\partial \phi} - 2\kappa \nabla^2 \phi \right) - \underbrace{M (c_l^{eq} - c_s^{eq}) (\mu - \mu_{eq})}_{driving\ force} \frac{\partial h(\phi)}{\partial \phi}. \quad (4.1)$$

which is coupled with the evolution equation for diffusion potential μ as

$$\left(\frac{\partial c_s}{\partial \mu} h(\phi) + \frac{\partial c_l}{\partial \mu} (1 - h(\phi)) \right) \frac{\partial \mu}{\partial t} = \nabla \cdot (M \nabla \mu) - (c_s(\mu) - c_l(\mu)) \frac{\partial h(\phi)}{\partial t} - \vec{v} \cdot \nabla c \quad (4.2)$$

For the sake of simplicity, in the present model we choose the following relations of c^s and c^l ,

$$c^l(\mu) = \mu \quad (4.3)$$

$$c^s(\mu) = kc^l = k\mu \quad (4.4)$$

where k is the partition coefficient.

Then, the expression for driving force becomes

$$\Psi_l - \Psi_s = (c_s^{eq} - c_l^{eq}) (\mu - \mu_{eq}) \Rightarrow \Psi_l - \Psi_s = (k - 1)\mu (\mu - \mu_{eq}) \quad (4.5)$$

The evolution equation for ϕ , thus suitably modified, can be written as,

$$\frac{\partial \phi}{\partial t} = -M \left(\frac{\partial f_o}{\partial \phi} - 2\kappa \nabla^2 \phi \right) - \underbrace{M (k - 1)\mu (\mu - \mu_{eq})}_{driving\ force} \frac{\partial h(\phi)}{\partial \phi}. \quad (4.6)$$

Also, we can consider χ such that,

$$\chi = \left(\frac{\partial c^s}{\partial \mu} h(\phi) + \frac{\partial c^l}{\partial \mu} (1 - h(\phi)) \right) \quad (4.7)$$

$$\Rightarrow \chi = 1 + (k - 1) h(\phi) \quad (4.8)$$

Then, the evolution equation for μ becomes,

$$\chi \left(\frac{\partial \mu}{\partial t} \right) = \nabla \cdot (M \nabla \mu) - (c^s(\mu) - c^l(\mu)) \frac{\partial h(\phi)}{\partial t} - \vec{v} \cdot \nabla c \quad (4.9)$$

$$\chi \left(\frac{\partial \mu}{\partial t} \right) = \nabla \cdot (M \nabla \mu) - (k-1) \mu \frac{\partial h(\phi)}{\partial t} - \vec{v} \cdot \nabla c \quad (4.10)$$

For the sake of reiteration, $\phi = 1 \implies$ solid and $\phi = 0 \implies$ liquid
Also, we have chosen the interpolation function $h(\phi)$ such that,

$$h(\phi) = \phi^2 (3 - 2\phi) \quad (4.11)$$

$$\frac{\partial h(\phi)}{\partial \phi} = 6\phi (1 - \phi) \quad (4.12)$$

We chose f_o to be the classic double well potential, given as,

$$f_o = 9\phi^2 (1 - \phi)^2 \quad (4.13)$$

We also factor in some constants in order to non-dimensionalise the equations: τ as relaxation time and ϵ as relaxation length and γ which corresponds to surface energy. Incorporating all this, the equation for evolution of the phase field variables can finally be written down as,

$$\tau \epsilon \frac{\partial \phi}{\partial t} = \gamma \nabla^2 \phi - \frac{\gamma}{\epsilon} 18\phi(1 - \phi)(1 - 2\phi) + (k-1) \mu (\mu - \mu_{eq}) (6\phi)(1 - \phi) \quad (4.14)$$

$$\chi \frac{\partial \mu}{\partial t} = M \nabla^2 \mu - 6(k-1) \mu \phi (1 - \phi) \frac{\partial \phi}{\partial t} \quad (4.15)$$

4.2 Incorporating Anisotropy

Our free energy functional is of the form,

$$\mathcal{F} = \int_{-\infty}^{\infty} (\gamma \epsilon |\nabla \phi|^2 + \frac{\gamma}{\epsilon} 9\phi^2(1 - \phi)^2 + \dots) dx$$

We define an inter facial energy term with a small cubic anisotropy to the order of a few percent.

$$a_c = \gamma_o (1 - \delta_{\alpha\beta} \cos(4\theta))$$

where $\delta_{\alpha\beta}$ is the strength of anisotropy.

Accordingly, the functional gets modified to be,

$$\mathcal{F} = \int_{-\infty}^{\infty} (\gamma a_c^2(\theta) |\nabla \phi|^2 + \frac{\gamma}{\epsilon} 9\phi^2(1 - \phi)^2 + \dots) dx$$

The interface energy needs to be a function of interface normal. For this, we determine the following relation between interface normal \vec{n} and θ which is the angle of orientation.

$$\theta = \tan^{-1} \left(\frac{n_x}{n_y} \right), \cos \theta = n_x, \sin \theta = n_y, \hat{n} = n_x \hat{i} + n_y \hat{j}$$

$$n_x^2 + n_y^2 = 1$$

$$\hat{n} = \frac{\nabla\phi}{|\nabla\phi|}, n_x = \frac{\left(\frac{\partial\phi}{\partial x}\right)}{|\nabla\phi|}, n_y = \frac{\left(\frac{\partial\phi}{\partial y}\right)}{|\nabla\phi|}$$

We need to be able to fully describe the anisotropy in terms of gradients in the ϕ field. To achieve this we resorted to the following treatment,

Expand $\cos 4\theta$ as,

$$\cos 4\theta + i \sin 4\theta = (\cos\theta + i \sin\theta)^2$$

on collecting the real terms we get

$$\begin{aligned} \cos 4\theta &= \cos^4\theta + \sin^4\theta - 4C_2\cos^2\theta\sin^2\theta \\ &= n_x^4 + n_y^4 - 6n_x^2n_y^2 \end{aligned}$$

$$\text{using: } n_x^2 + n_y^2 = 1$$

$$\Rightarrow \cos 4\theta = 4(n_x^4 + n_y^4) - 3$$

$$\text{Now, } a_c = \gamma_o (1 - \delta_{\alpha\beta} (4(n_x^4 + n_y^4) - 3))$$

$$a_c = \gamma_o \left(1 - \delta_{\alpha\beta} \left(4 \left(\frac{\frac{\partial\phi^4}{\partial x} + \frac{\partial\phi^4}{\partial y}}{|\nabla\phi|^4} \right) - 3 \right) \right)$$

$$\text{call } \frac{\partial\phi}{\partial x} \rightarrow \phi_x \text{ and } \frac{\partial\phi}{\partial y} \rightarrow \phi_y$$

$$a_c = \gamma_o \left(1 - \delta_{\alpha\beta} \left(4 \left(\frac{\phi_x^4 + \phi_y^4}{(\phi_x^2 + \phi_y^2)^2} \right) - 3 \right) \right)$$

The variational derivative operator expands as,

$$\frac{\delta}{\delta\phi} = \left(\frac{\partial}{\partial\phi} - \nabla\phi \cdot \frac{\partial}{\partial\nabla\phi} \right)$$

Then on incorporating anisotropy, the gradient energy term in the evolution equation gets modified as,

$$\frac{\delta}{\delta\phi} (a_c^2 |\nabla\phi|^2) = \frac{\partial}{\partial\phi} (a_c^2 |\nabla\phi|^2) - \nabla\phi \cdot \frac{\partial}{\partial\nabla\phi} (a_c^2 |\nabla\phi|^2)$$

As both a_c and $|\nabla\phi|$ are a function of ϕ_x and ϕ_y only, the $\frac{\partial}{\partial\phi}$ term goes to zero.

Also, in Cartesian coordinates $\frac{\partial}{\partial\nabla\phi}$ can be written as,

$$\frac{\partial}{\partial\phi_x} \hat{i} + \frac{\partial}{\partial\phi_y} \hat{j}$$

As such, on including anisotropy the time evolution equation for the phase field variable becomes:

$$\tau\epsilon\frac{\partial\phi}{\partial t} = \gamma\epsilon\nabla \cdot \begin{pmatrix} \frac{\partial}{\partial\phi_x} (a_c|\nabla\phi|)^2 \\ \frac{\partial}{\partial\phi_y} (a_c|\nabla\phi|)^2 \end{pmatrix} - \frac{\gamma}{\epsilon} 18\phi(1-\phi)(1-2\phi) + (k-1)\mu(\mu - \mu_{eq}) 6\phi(1-\phi) \quad (4.16)$$

The individual components of the vector $\begin{pmatrix} \frac{\partial}{\partial\phi_x} (a_c|\nabla\phi|)^2 \\ \frac{\partial}{\partial\phi_y} (a_c|\nabla\phi|)^2 \end{pmatrix}$

when fully expanded look as follows,

$$\frac{\partial}{\partial\phi_x} (a_c|\nabla\phi|)^2 = 2a_c\phi_x \left(\gamma_o - \frac{16\gamma_o\delta(\phi_x^2\phi_y^2 - \phi_y^4) + 4\gamma_o\delta(\phi_x^4 + \phi_y^4)}{(\phi_x^2 + \phi_y^2)^2} + 3\delta \right) \quad (4.17)$$

$$\frac{\partial}{\partial\phi_y} (a_c|\nabla\phi|)^2 = 2a_c\phi_y \left(\gamma_o - \frac{16\gamma_o\delta(\phi_y^2\phi_x^2 - \phi_x^4) + 4\gamma_o\delta(\phi_y^4 + \phi_x^4)}{(\phi_y^2 + \phi_x^2)^2} + 3\delta \right) \quad (4.18)$$

$$\text{where, } a_c = \gamma_o \left(1 - \delta \left(4 \left(\frac{\phi_x^4 + \phi_y^4}{(\phi_x^2 + \phi_y^2)^2} \right) \right) \right)$$

4.3 Fluid Flow

The navier stokes equation for continuity and momentum transport are as follows,

$$\nabla \cdot \vec{v} = 0 \quad (4.19)$$

$$\frac{\partial \vec{u}}{\partial t} = -\nabla P + \nu \nabla^2 (\vec{u}^2) - \vec{u} \cdot \nabla (\vec{u}) \quad (4.20)$$

For coupling with the phasefield model we utilised a description where in the diffuse interface region is viewed as a rigid porous medium. Here, the usual no-slip condition at the sharp solid- liquid interface is enforced through a varying interfacial force term $(1 - \phi)$ in the diffuse interface region. The mass and momentum conservation equations can then be written, respectively as,

$$\nabla \cdot [(1 - \phi)\vec{v}] = 0 \quad (4.21)$$

$$\frac{\partial \vec{v}(1 - \phi)}{\partial t} = -(1 - \phi)\nabla P + \nu \nabla^2 (\vec{v}(1 - \phi)) - \vec{v} \cdot \nabla (\vec{v}(1 - \phi)) \quad (4.22)$$

Results and Discussion

5.1 One Dimensional Solidification Profile

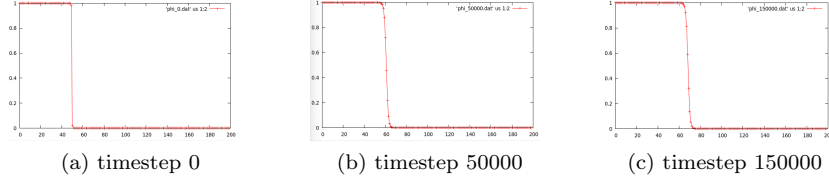


Figure 5.1: time evolution of the phase field variable ϕ

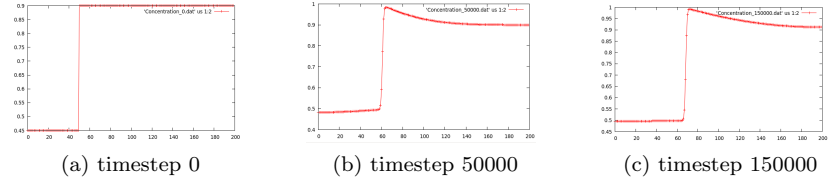


Figure 5.2: time evolution of concentration

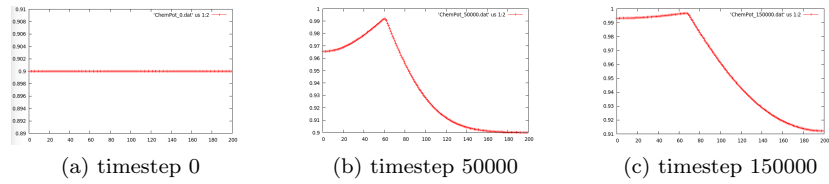


Figure 5.3: time evolution of chemical potential μ

5.2 Isotropic Solidification in 2D - only diffusion

Gibbs Thompson Effect

Our functional has the form of

$$\mathcal{F} = \int_{-\infty}^{\infty} (\gamma \epsilon |\nabla \phi|^2 + \frac{\gamma}{\epsilon} 9 \phi^2 (1 - \phi)^2 + \dots) dx$$

By the equipartition of energy, we have for one dimension,

$$\begin{aligned} \gamma \epsilon \left(\frac{\partial \phi}{\partial x} \right)^2 &= \frac{\gamma}{\epsilon} 9 \phi^2 (1 - \phi)^2 \\ \Rightarrow \frac{\partial \phi}{\partial x} &= \frac{3}{\epsilon} \phi (1 - \phi) \end{aligned}$$

Also, we know that total surface energy σ is given by the interfacing the interfacial energy term over the entire domain,

$$\begin{aligned} \sigma &= \int_{-\infty}^{\infty} 2\gamma \epsilon \left(\frac{\partial \phi}{\partial x} \right)^2 \cdot dx \\ &= \int_0^1 2\gamma \epsilon \left(\frac{\partial \phi}{\partial x} \right) \cdot d\phi \\ &= 6\gamma \int_0^1 2\phi(1 - \phi) d\phi \\ &= 6\gamma \left[\frac{\phi^2}{2} - \frac{\phi^3}{6} \right]_0^1 \\ \Rightarrow \sigma &= \gamma \end{aligned}$$

We also know that at critical radius, surface energy equals the driving force for solidification- due to deviation of deviation from equilibrium of the Free energy,

$$\frac{\sigma}{r} = (k - 1)\mu (\mu - \mu_e q)$$

Thus, for a given radius we can calculate a critical μ_c for which the nuclei is stable. For $\mu > \mu_c$ nuclei should grow, otherwise the nuclei should shrink.

We used the above relation to calculate the critical μ_c for a radius of 50 units and found out that this Gibbs Thompson relation indeed holds for the present model.

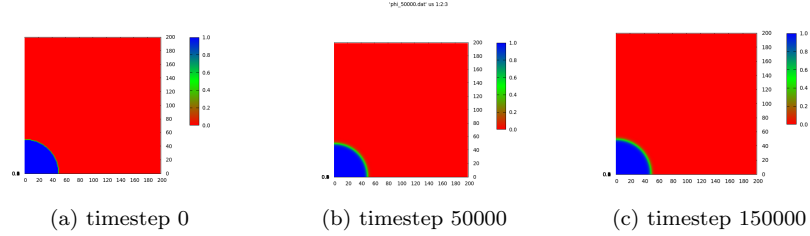


Figure 5.4: Nuclei at critical radius - not growing

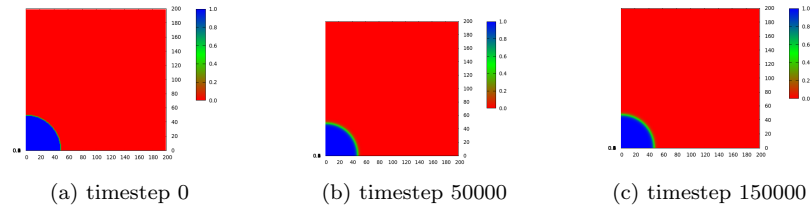


Figure 5.5: Nuclei Shrinking - μ less than critical

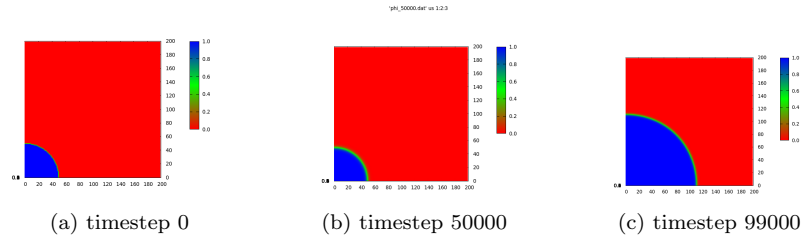


Figure 5.6: Nuclei growing - μ greater than critical

5.3 Dendritic Solidification

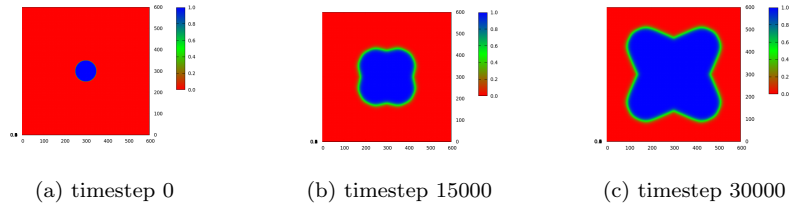


Figure 5.7: Dendritic solidification on introducing anisotropy

Discretisation

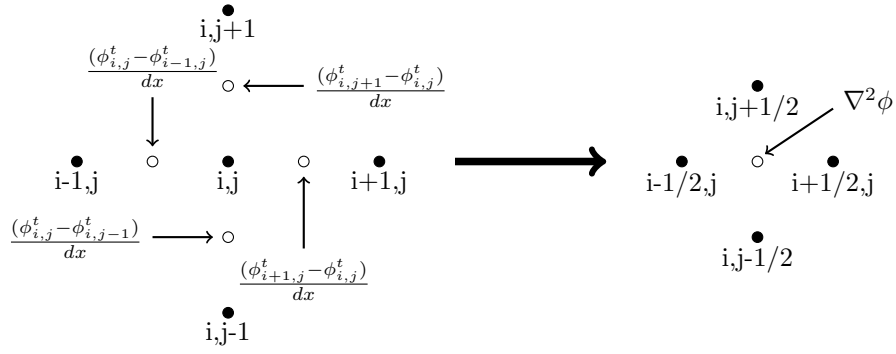


Fig 7: Discretization of the domain for calculating the Laplacian of ϕ .

Bibliography

- [1] J.S. Langer, in *Chance and Matter* edited by J. Souletie, J. Vannimenus and R Stora (North Holland, Amsterdam 1987), p. 629; D Kessler, J Koplik and H Levine, *Adv. Phys.* **37**, 255 (1988); E .A. Brenner and V.I. Mel'nikov, *ibid.* **40**, 53 (1991)
- [2] J. S. Langer, *Phys. Rev. Lett.*, Volume 44, Number 15, 14 April 1980
- [3] I. Steinbach, *Acta Materialia* **57** (2009) 2640 - 2645
- [4] Tong et al., *Phys. Rev. E*, vol. 61, No. 1, January 2000
- [5] Abhik Chaudhary and Britta Nestler, *Phys. Rev. E* **85**,021602 (2012)
- [6] Britta Nestler and Abhik Chaudhary, *Current Opinion in solid State and Materials Science* **15**(2011) 93-105
- [7] C. Beckermann et al., *Jour. of Comp. Phys.* **154**, 468-496 (1999)
- [8] W.J. Boettinger et al., *Annu. Rev. Mater. Res.* **2002**. **32**:163-94