

6.1 Introduction

6.1.1 Structures of Cast Materials

Metals are crystalline materials, i. e. they have regular structures. The crystals have various forms and sizes and in addition the crystal structure varies from one metallic material to another.

The crystal structure determines most of the material properties of the metal. It is thus of greatest importance to study metal structures in detail and try to identify the relation between structure and material properties. The way of formation determines the structure of a metal.

If one wants to produce metallic materials with certain specific properties, it is necessary to understand the relation between the way of formation and different process parameters in order to find an efficient way of controlling the structure and the properties.

Metal- and alloy components are often produced by aid of casting. The material properties of ingots and other castings are to a great extent determined by the way of solidification and by the choice of parameters, i. e. casting method, casting temperature and rate of solidification or in practice by the cooling process.

The cooling rate determines the coarseness of the metal structure. By combining heat conduction equations with relations, which describe coarseness of structure, it is possible to analyse the formation of the structure mathematically for different casting processes. The analysis results in predictions of the material properties of the ingot or casting. Inversely the analysis can be used for control of the casting process and for reduction of the casting errors to a minimum.

In this chapter we will discuss nucleation, appearance and formation of casting structures and the relation between the structure formation in various casting processes and the corresponding solidification processes. We will also discuss the possibility of controlling casting processes in order to give the material the desired material properties.

6.2 Structure Formation in Cast Materials

6.2.1 Nucleation

A necessary condition for formation of crystals in a melt is that it is *undercooled*, i. e. has a temperature which is lower than the melting point T_M respectively the liquidus temperature T_L .

Attraction forces are acting between the atoms in a melt. They are no stronger than that they permit the atoms to move freely relative to each other. The binding energy is low.

In a solid crystalline phase the forces between the atoms are stronger than the ones in a liquid phase. The atoms are arranged in a crystal lattice and can only make small deviations from their equilibrium positions in the form of vibrations. The binding energy of the lattice atoms is higher than the one between liquid atoms.

All solidification starts by formation of so called *nuclei* at various positions in the melt. The crystals grow from these nuclei. There is a distinction between *homogeneous* and *heterogeneous* nucleation. Heterogeneous nucleation implies that foreign particles in the melt or at the surface serve as nuclei for subsequent growth of the solid phase. The process occurs spontaneously but may, if special measures are taken, be controlled and contribute to the desired structure of the solidified material.

Homogeneous Nucleation

The kinetic motion in a melt results in incessant collisions between its atoms. The particles, so-called *embryos*, which consist of several atoms, arise spontaneously and at random. Between the embryos and the melt there is a continuous exchange of atoms. In many cases the embryos are dissolved and disappear.

If the kinetic motion is not too violent – the temperature is lower than the melting point – the attraction between the atoms may sometimes be so strong that some embryos "survive" in the melt. An embryo, which achieves a certain critical minimum size, continues to grow as a crystal.

Embryos, which are bigger than the critical size, required for continuous growth, are called *nuclei*. Embryos, which successively increase their sizes, form nuclei by random additions of atoms from the melt. As can be concluded by the statements above an embryo has a higher free energy than the melt. This excess free energy $-\Delta G_i$ can be written

$$-\Delta G_i = \frac{V \cdot (-\Delta G_m)}{V_m} + \sigma A \quad (1)$$

where

$-\Delta G_i$ = free energy required to form an embryo with the volume V and the area A

$-\Delta G_m$ = change in free energy per kmole at transformation of melt to solid phase

V_m = molar volume = M/ρ (m^3/kmole)

σ = surface energy per unit area of the embryo.

For a spherical nucleus we have

$$V = \frac{4\pi r^3}{3} = \frac{4\pi r^2 \cdot r}{3} = \frac{A \cdot r}{3}$$

The surface energy is caused by the surface tension forces, which work in the interface between the embryo and the melt. They try to keep the embryo together and minimise its area.

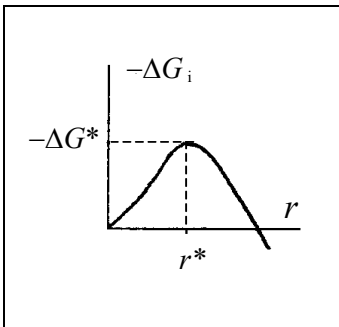


Figure 1.

Free energy, required to form an embryo, as a function of the its radius.

By common maximum-minimum calculations on the function $(-\Delta G_i)$ (the curve in figure 1) we obtain both the critical size r^* of a nucleus, capable of growing, and an expression for the energy $(-\Delta G^*)$, which is required to form such a nucleus. The calculations are not performed here but the final result is

$$-\Delta G^* = \frac{16\pi}{3} \cdot \frac{\sigma^3 V_m^2}{(-\Delta G_m)^2} \quad (2)$$

where

$-\Delta G^*$ = activation energy for formation of a nucleus of the critical size r^* .

An embryo has to grow at least to the critical size r^* to form a nucleus, capable of further growth.

If we can find an expression for $(-\Delta G_m)$ in terms of heat of fusion and temperature or in terms of composition differences we may understand the formation of new crystals.

For *pure* metals there is a simple relation between the molar free energy of the melt and the molar heat of fusion, which involves the temperature of the melt and the melting point temperature

$$-\Delta G_m = \frac{(T_M - T)}{T_M} \cdot (-\Delta H_m^{\text{fusion}}) \quad (3)$$

where

$-\Delta G_m$ = change in free energy per kmole at transformation of melt to solid phase

$-\Delta H_m^{\text{fusion}}$ = molar heat of fusion of the alloy

T_M = melting point temperature of the pure metal

T = temperature of the melt

$T_M - T$ = undercooling.

Equation (3) is *not* valid for alloys as they have a solidification interval instead of a well-defined melting point. The phase diagrams of alloys have to be involved. The simple phase diagram in figure 2 describes the relation between undercooling and supersaturation. Supersaturation is a simpler variable to describe $(-\Delta G_m)$ for alloys than undercooling. This matter will be discussed in next section.

Equations (2) and (3) do not give any information concerning *when* or *at what temperature* the nucleation occurs. However, by use of Boltzmann's statistical mechanics in combination with laws, which describe the possibility of the nuclei to grow and not shrink and become embryos again, it is possible to derive

- the number of nuclei per unit volume
- the number of nuclei formed per unit volume and unit time
- the average activation energy, required to form a nucleus of the critical size.

By such calculations it is found that a large number of nuclei per unit volume are formed at a certain critical temperature T^* . The average activation energy of forming a nucleus at this temperature is calculated to be

$$-\Delta G^* = 60 kT^* \quad (4)$$

where k is Boltzmann's constant.

The critical temperature T^* , when a large number of nuclei are formed, is found by combining the equations (2) and (4) with equation (3), applied at the critical temperature (T is replaced by

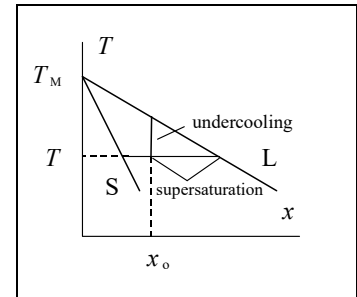


Figure 2.

Phase diagram of an alloy.

T^*). Elimination of the molar free transformation energy ($-\Delta G_m$) gives

$$60 kT^* = \frac{16\pi}{3} \cdot \frac{\sigma^3 V_m^2}{\left(\frac{(T_M - T^*)}{T_M} \cdot (-\Delta H_m^{\text{fusion}}) \right)^2} \quad (5)$$

The solution of this equation gives the critical temperature T^* .

Below the critical temperature very few nuclei per unit volume and unit time are formed. *Above* and *at* this temperature a great number of nuclei per unit time are formed. The critical temperature is also called the *nucleation temperature*.

The undercooling $T_M - T$ is the driving force of solidification. The lower the temperature T of a pure metal melt is, the faster will the nucleation process be.

Heterogeneous Nucleation in Presence of Small Amounts of Foreign Elements

If the nucleation temperature T^* is calculated by aid of equation (5) it is found that the value is *much* lower than the melting point temperature. A very large undercooling is required to give nucleation in an absolutely pure metal melt. Such large undercoolings are quite unrealistic at normal casting and solidification processes.

In reality nucleation occurs at considerably higher temperatures, i. e. at much lower undercoolings than the calculations indicate. The explanation of this is that even "pure" metals contain small amounts of foreign elements, which influence the nucleation temperature strongly.

The nucleation occurs on foreign particles crystals, so-called *heterogeneities*, which are precipitated in the melt. When crystals are formed on these small heterogeneities, some of their surface energy supplies the required formation energy of the new crystal. The last term on the right-hand side in equation (1) will then decrease. By aid of the analogous calculations as for pure metals it is found that the critical temperature T^* in the presence of homogeneities becomes much higher than for pure metals.

Many times these small foreign particles nucleate homogeneously. The metal melt, including the foreign element, which is difficult to dissolve, constitute a two-phase system with a phase diagram like the one in figure 3.

At the nucleation temperature T^* small crystals precipitate with a composition x^s which is approximately equal to 100 % of element B. These small crystals grow somewhat at the continued cooling.

For elements with very low solubility in the melt the relation between the change of the molar free energy of the melt at nucleation (solidification) is

$$-\Delta G_m = RT^* \ln \frac{x_o^L}{x_{eq}^L} \quad (6)$$

where

x_{eq}^L = equilibrium concentration (mole fraction) of the foreign element in the melt at temperature T^*

x_o^L = original concentration (mole fraction) of the foreign element in the melt.

In this case the critical temperature T^* is found by combining the equations (2), (4) and (6). Elimination of $(-\Delta G_m)$ gives

$$60 kT^* = \frac{16\pi}{3} \cdot \frac{\sigma^3 V_m^2}{\left(RT^* \ln \frac{x_o^L}{x_{eq}^L} \right)^2} \quad (7)$$

The solution of this equation gives the critical temperature T^* . It is found to be much higher than that of pure metal melts and corresponds to a low undercooling. It is low enough to permit easy formation of new crystals at normal casting and solidification processes.

6.2.2 Inoculation

Calculations by use of equation (7) show that the required concentration of foreign elements in the melt for formation of heterogeneities is very low. This fact is used at so-called inoculation of metals to start the solidification process. Small

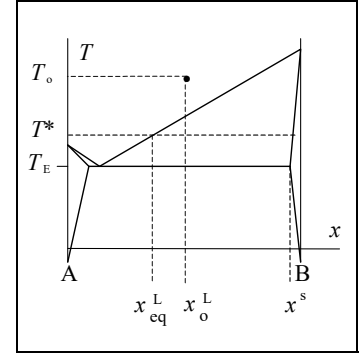


Figure 3.

Phase diagram of a metal with a low concentration of a foreign element B, which is difficult to dissolve.

amounts of elements are added to the melt and small crystals are precipitated by homogenous nucleation. These crystals constitute the heterogeneities on which new crystals nucleate. The mechanism is called *heterogeneous nucleation*.

The properties of cast materials are often improved by increase of the number of formed crystals. Thus it is of great interest to increase the number of crystals during the solidification in technical processes.

Inoculation is often used at casting of aluminium to get a great number of small crystals, which prevent formation of bad texture (bad mechanical strength in certain crystal directions) at machining after casting. Inoculation is also used at casting of iron to reduce the risk of white solidification (page 27).

In addition, in many casting processes an increase of the number of crystals occurs by so called *crystal multiplication*, when a single crystal splits up into two or more new crystals. This phenomenon is treated in section 6.3.3 on page 13.

At nucleation on heterogeneities in the melt it is difficult to define a specific nucleation temperature. Instead it has to be based on experimental observations. In many cases a relation between the number of nucleated small crystals per unit volume N and the undercooling $\Delta T = T_L - T$, which is the difference between the liquidus temperature T_L and the temperature T of the melt, has been found to be

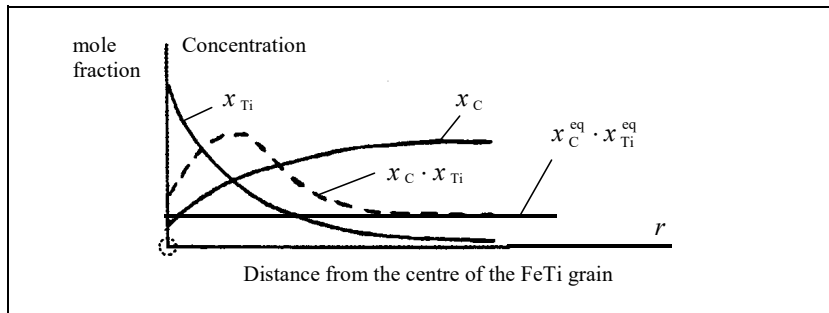
$$N = A \cdot (T_L - T)^B = A \cdot (\Delta T)^B \quad (8)$$

A and B are constants, which are determined experimentally.

Example 1.

At inoculation of metals an alloying element is added to the melt. An example on this is addition of FeTi to a steel melt with about 2.3 at-% C. When FeTi is added TiC is precipitated.

When FeTi is added to the melt Ti-atoms will diffuse into the melt at the same time as C-atoms diffuse inwards towards the FeTi-grains. The processes lead to the concentration profile, which is given in the figure. The solubility product of C and Ti concentrations at equilibrium is constant. This product is illustrated in the figure by a straight line.

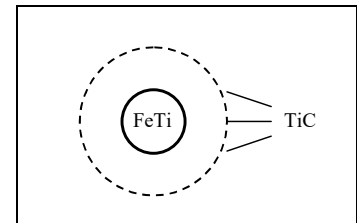


The solubility product of TiC at the given temperature 1500 °C is $x_{\text{Ti}}^{\text{eq}} \cdot x_{\text{C}}^{\text{eq}} = 5.2 \cdot 10^{-6}$ (mole fraction)². Other data have to be taken from tables.

- Calculate the concentration of FeTi (at-%), which has to be added to give an inoculation effect.
- The product of the Ti- and C-concentrations is illustrated in the figure by the dotted curve. It varies strongly with the distance from the grain. Why is it not constant and equal to the product $x_{\text{Ti}}^{\text{eq}} \cdot x_{\text{C}}^{\text{eq}}$?

Solution:

The precipitation is illustrated schematically in the margin figure. Each grain of FeTi gives many dispersed TiC particles, which serve as heterogeneities in the melt. Thus FeTi serves as an effective inoculation agent. Besides, TiC lowers the solubility of C in the melt.



Equation (4) on page 5 is applied in combination with equations (6) on page 7. If we assume that $T^* = T$ we get:

$$-\Delta G^* = 60 kT^* = \frac{16\pi}{3} \cdot \frac{\sigma^3 V_m^2}{(-\Delta G_m)^2} \quad (1')$$

$$\text{where} \quad -\Delta G_m = RT^* \ln \frac{x_{\text{Ti}} \cdot x_{\text{C}}}{x_{\text{Ti}}^{\text{eq}} \cdot x_{\text{C}}^{\text{eq}}} \quad (2')$$

as both C and Ti are alloying elements. We insert the expression (2') into equation (1') and solve x_{Ti} :

$$x_{\text{Ti}} = \frac{x_{\text{Ti}}^{\text{eq}} \cdot x_{\text{C}}^{\text{eq}}}{x_{\text{C}}} \cdot e^{\frac{1}{RT^*} \sqrt{\frac{16\pi \sigma^3 V_m^2}{3 \cdot 60 kT^*}}} = \frac{x_{\text{Ti}}^{\text{eq}} \cdot x_{\text{C}}^{\text{eq}}}{x_{\text{C}}} \cdot e^{0.648} \quad (3')$$

Inserting the given data gives $x_{\text{Ti}} = 4.36 \cdot 10^{-4}$ mole fraction.

at-% = mole fract

Data:

-
 $R = 8.314 \cdot 10^3 \text{ J/k}$
 $T = 1500 + 273 =$
 $V_m = 8 \cdot 10^{-3} \text{ m}^3/\text{k}$
 $\sigma = 0.5 \text{ J/m}^2$
 $k = 1.38 \cdot 10^{-23} \text{ J/}$
 $x_{\text{C}} = 2.3 \text{ at-\%}$
 $x_{\text{Ti}}^{\text{eq}} \cdot x_{\text{C}}^{\text{eq}} = 5.2 \cdot$
(mole

-

Answer:

a) One has to add such an amount of FeTi that the initial Ti-concentration in the melt will be 0.044 at-%.

b) A few crystals of TiC are primarily nucleated at the surface of the added FeTi-particles and stay there. C- and Ti-atoms diffuse inwards respectively outwards. The difference between the dotted curve and the equilibrium line represents the supersaturation of TiC. It is seen from the figure that no supersaturated zone is initially present close to the FeTi particles, because the supersaturation there is too low for nucleation of TiC.

The diffusion of the atoms results in an increase of the concentrations of C and Ti around the FeTi particle with time and the product $x_{\text{C}} \cdot x_{\text{Ti}}$ increases and exceeds the solubility product. Thus a driving force is developed and a spherical shell of TiC crystals around the FeTi particle is formed.

The answer of the question is that the diffusion shows that the system is not at equilibrium. Equilibrium is achieved at large distances from the FeTi grains.

6.3 Dendrite Structure and Dendrite

Growth

Because structure is extremely important for the material properties of metals structure studies has been the object of the scientists' interest for a very long time.

Dendrite Structure

The most apparent features of the structure of ingots are the sizes and shapes of the crystals. Already in the 18th century the French scientist Grignon accomplished a detailed study. He found that needle-shaped crystals with branchings were formed when cast iron melts solidify. A great number of modern investigations have been done to find out the influence of various factors on the casting structure.

Today the needle-shaped crystals, which were detected in the 18th century, are called *dendrites*. Most technically interesting alloys solidify by a primary precipitation of dendrites.

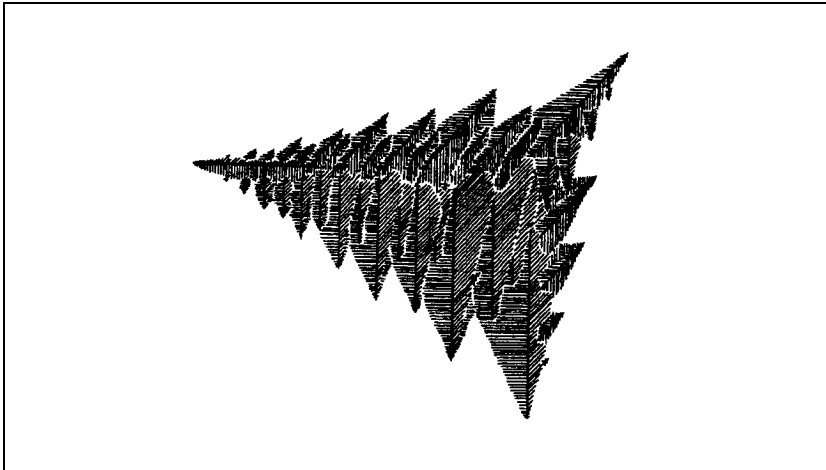


Figure 4.
Schematic sketch of a dendrite crystal aggregate with primary and secondary dendrite arms. From the secondary arms tertiary dendrite arms grow, which is apparent from the figure.

Figure 4 shows an explanatory sketch of a dendrite. Such a crystal aggregate is formed because its growth in certain specific crystal directions is favoured. From a crystal nucleus a dendrite tip grows, which forms the *main branch* or *primary arm*.

Immediately behind this tip *lateral arms* or *secondary dendrite arms* are formed. For metals with cubic structures they are

normally situated in two mutually perpendicular planes. The lateral arms are perpendicular to each other and the main branch. The lateral arms are successively formed behind the growing tip and get different lengths for this reason. The oldest ones are longest.

6.3.1 Relation between Dendrite Arm Distance and Growth Rate

It has been found that *the distance between the secondary dendrite arms* is not constant but *increases with the distance from the cooled surface* at directed solidification and at growth towards the centre in an ingot. Both experimentally and theoretically it has been shown that the following relation is valid between the distance λ for secondary or primary dendrite arms and the growth rate v :

$$v_{\text{growth}} \cdot \lambda^2 = \text{const} \quad (9)$$

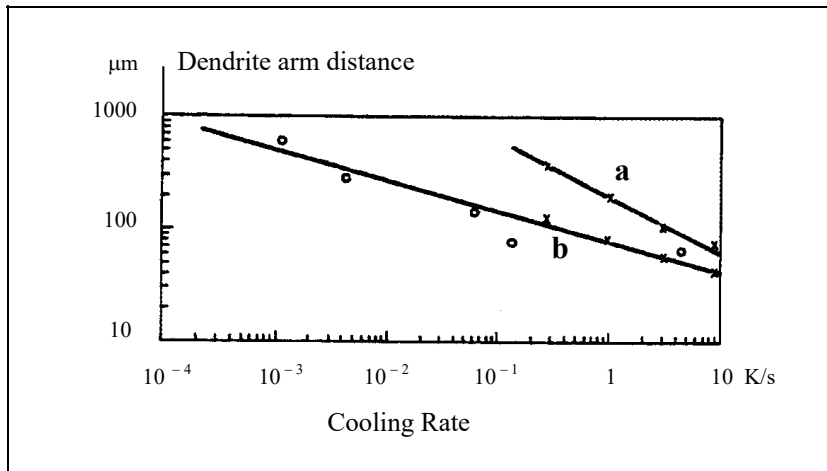
The dendrite arm distance also depends on a number of other factors, among them the composition of the alloy and its phase diagram. The influence of these factors affects the value of the constant in equation (9).

In most experimental investigations the authors have chosen to report relations between the dendrite arm distance and some other experimental variable, for example the total solidification time θ . In this case the relation will be

$$\lambda = K \cdot \theta^n \quad (10)$$

where K and n are two constants.

The constant n has a value between $1/3$ and $1/2$ for different types of steel alloys. Figure 5 shows the result of an investigation of a low-carbon alloy where the primary and secondary dendrite arm distances have been measured as functions of the cooling rate of the melt by microscope studies.



When a network of dendrite arms has been formed there is remaining melt left between the arms. This melt solidifies at the continued temperature decrease by precipitation of solid phase on the dendrite arms. The solid phase and the melt of the alloy have different compositions. During the solidification process the melt will successively be enriched of the alloying element. The last solidified melt will therefore have a higher concentration of the alloying element than the one, which has solidified during an earlier stage. The phenomenon is called *microsegregation* and will be treated in chapter 7.

6.3.2 Relation between Growth Rate and Undercooling

In section 6.2.1 we pointed out that a condition for formation of nuclei is that the melt has a temperature which is lower than its melting point, i. e. the melt is *undercooled*. A necessary condition for growth of a dendrite tip is that the melt is undercooled. Another necessary condition is that the solidification heat is transported away from the solidification front. If this condition is not fulfilled the solidification process will stop. The heat flux from the solidification front determines the growth rate.

The higher the growth rate is the larger will the undercooling be. In most cases the following simple relation between the growth rate of the dendrite tip and the undercooling of the melt is valid

$$v_{\text{growth}} = \mu \cdot (T_L - T)^n \quad (11)$$

where

T_L = liquidus temperature of the melt

T = temperature of the melt at the solidification front.

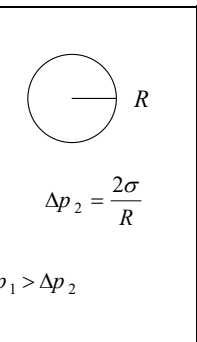
μ and n are constants. The value of n is usually between 1 and 2.

6.3.3 Crystal Multiplication

In the past we have treated inoculation as a method to facilitate the formation of crystals. The second method is so-called *crystal multiplication*. It means that parts of the dendrite skeleton are carried out into the melt and serve as nuclei for new crystals. Different mechanisms for spontaneous crystal multiplication have been suggested. One of them is based on the theory that *fragments are torn off in a purely mechanical way from the growing dendrite tips*, for example by influence of the natural convection in the melt. Examples of such a process are the broom-distributed cementite in white cast iron (page 27), the corresponding phenomenon for silicon in silumin (page 20-21) and feather crystal growth in aluminium.

This mechanical mechanism of crystal multiplication can be used on purpose in many ways. It has been found that many materials can be forced to fine-grain solidification by supersonics treatment. In order to get full effect one has to combine the supersonics treatment with violent convection to make sure that the torn fragments are carried out into the melt in an effective way. This method has been successful for both aluminium alloys and stainless steel.

A rotating magnet field has also proved to be an effective way to cause a fine-grain structure in magnetic metal melts. The electromagnetic field is applied in a circle around the ingot. By choosing a convenient frequency of the alternating current in the magnetic coils it is possible to force the melt to rotate with the magnetic field. To increase the fine-grain structure the direction of the magnetic field is changed regularly. At the forced rotation of the melt shearing forces arise on the dendrite arms, large enough to break them and then carry them into the melt. The method has been used successfully on both static and continuous casting.



The second mechanism of crystal multiplication is based on the principle of *melting off dendrite arms*. Papapetrou introduced this method already in the 1930th. He found that certain dendrite arms are melted off under the influence of the surface tension.

The reason for this is that surface tension causes an excess pressure of the melt inside the interface, which is reciprocally proportional to the radius of curvature. The overpressure causes a melting point decrease, which is proportional to the pressure.

The excess pressure, caused by the surface tension of a small droplet, is larger than the pressure caused by a big droplet.

The secondary dendrite arms are always thinnest at the root, where the radius of curvature is smallest, and the excess pressure maximum. Thus the dendrite arms preferably melt off at the roots. If this happens the released dendrite fragments might be distributed into the melt by convection and serve as nuclei for new crystals.

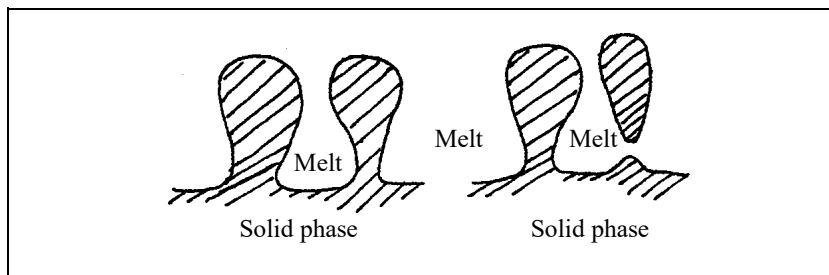


Figure 7.
Melting off of dendrite arms.

In practice the melting off method can be realised in a very direct way by increasing the temperature during the solidification. This can be done by the convection in the melt, which might direct a flow of hot material to a growing crystal. The simplest way to cause a desired temperature increase is to add a small amount of hot melt during the solidification in progress.

6.4 Eutectic Structure and Eutectic Growth

In section 6.3 we have discussed dendrite solidification. In the last part of the 20th century eutectic alloys are getting used as composite material. Two of the most frequently used cast alloys,

silumin and cast iron, are eutectic. It is therefore important to understand the solidification process and its influence on the structure of this type of alloys, too, to make it possible to give them good material properties.

A eutectic alloy is a binary alloy with a composition, which corresponds to the eutectic point in the phase diagram of the binary system.

Eutectic Reactions. Eutectic Alloys

Figure 8 shows a simplified version of the phase diagram of a binary alloy.

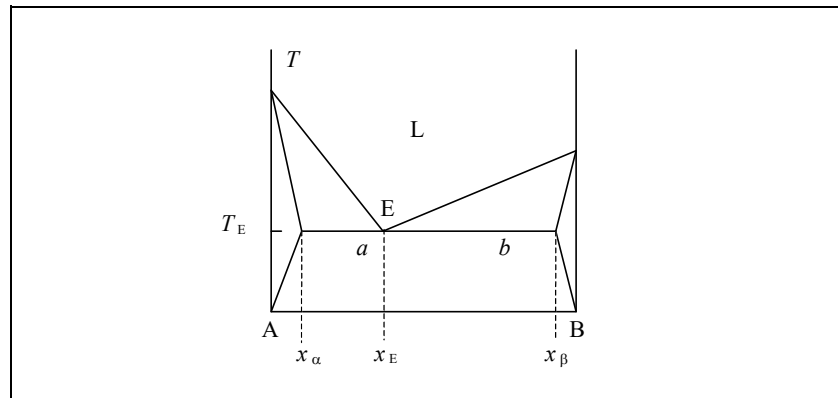
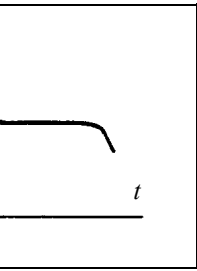


Figure 8.
Schematic phase diagram of a
binary alloy.

If the temperature decreases in a melt with eutectic composition, solid phase is not precipitated until the temperature has gone down to the eutectic temperature T_E . The alloy solidifies by precipitation of two solid phases with the compositions x_α and x_β at the constant proportions

$$\frac{N_A}{N_B} = \frac{b}{a} \quad (12)$$



These constant proportions correspond to a constant chemical composition of the solid and the solidification process is called a *eutectic reaction* for this reason. During the whole solidification process the temperature remains constant and close to T_E . When all melt has solidified the temperature continues to sink.

No microsegregation is present in eutectic alloys.

Figure 9 shows a typical temperature-time curve at eutectic solidification.

Normal Eutectic Structure

The structures of eutectic alloys show many different morphologies. The main types are described in table 1 on next page.

Several of the morphologies may occur within one and the same alloy, depending on the growth mechanism of the individual phases. It is primarily the conditions at the interface between the phases, which determine the structure. The growth of the phases is thus the basis of classification.

Table 1. Eutectic structures.

Designation	Description
Lamella eutectic structure	The two solid phases are tied in separate planar layers.
Rod eutectic structure	One of the phases is precipitated as rods and is surrounded by the other phase.
Spiral eutectic structure	One of the phases is precipitated as spirals and is surrounded by the other phase.
Flake-like eutectic structure	One of the phases is precipitated as plates, separated from each other and surrounded by the other phase.
Nodular eutectic structure	One of the phases is precipitated as spherical particles and is surrounded

	by the other phase.
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At a *degenerated eutectic reaction* one of the two phases grows more rapidly than the other phase.

At a *normal eutectic reaction* both phases grow in close co-operation with each other and with the *same* growth rate. This is the definition of normal eutectic structure. We will restrict the further discussion to this case.

The microstructure of a eutectic alloy has many different appearances. Figures 10 a-b and figures 11 a-c give some examples. The matter will be discussed more in detail later.

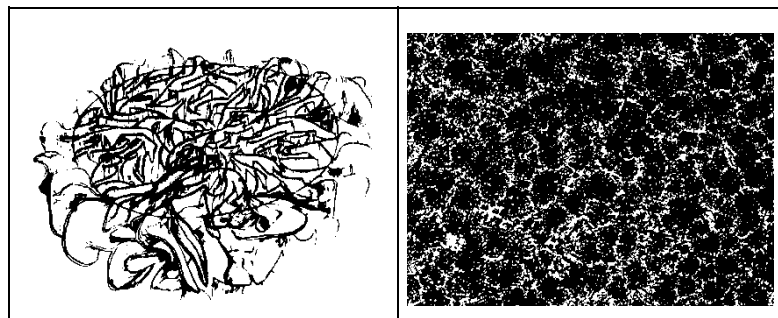
During several casting processes the growth occurs radially from centra when the two phases are formed (figure 10 a). These centra are called *eutectic cells* or just *cells*. They grow radially until they meet and fill the whole molten volume.

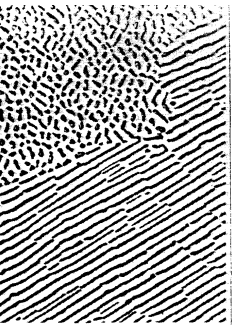
Figure 10 a.

Graphite crystal, which grows radially from a centre, forming a cell.

Figure 10 b.

Eutectic macrostructure of a large number of cells of the type shown in figure 10 a.





rod eutectic structure.

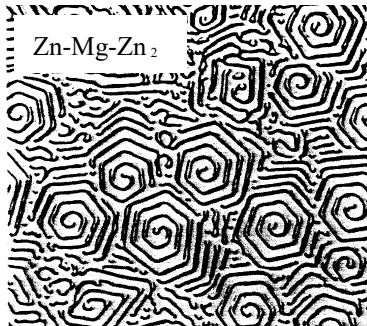


Figure 11 b.
Spiral eutectic structure.
Zn-Mg-Zn₂.



Figure 11 c.
Disk-shaped Si plates in a matrix
of Al.

Lamella Eutectic Structure

A lamella structure is formed when both precipitated phases grow side by side as figure 12 a shows.

During the growth of an α -lamella the B-atoms will continuously concentrate in the melt in front of the solidification front of the α -lamella. In the same way the A-atoms concentrate in front of the solidification front of the β -lamella. This leads to diffusion of A- and B-atoms as is shown in figure 12 b.

It can be shown that that the three-phase equilibrium along the lines, where α -phase, β -phase and melt meet, results in curved surfaces instead of planes. The formation of lamella eutectic structure is associated with the surface tension conditions of the α - and β -phases.

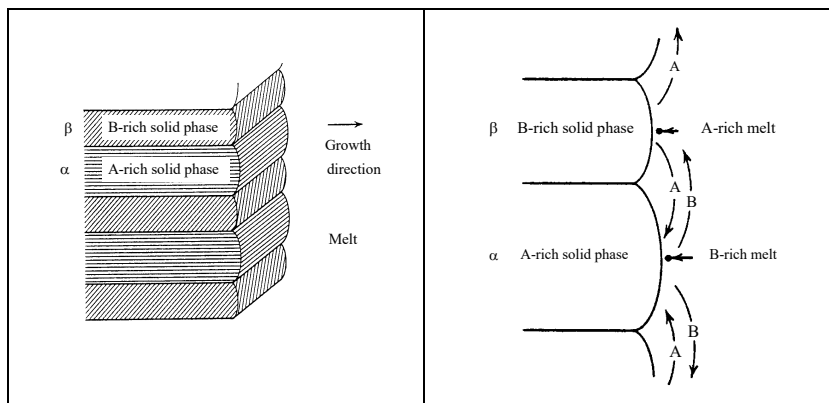


Figure 12 a.

Figure 12 b.

6.4.1 Relation between Lamella Distance and Growth Rate

The distance between two neighbouring lamellas is called λ . The relation between the lamella distance λ and the growth rate v_{growth} of the eutectic colony (pile of lamellas) is in most cases described by the equation

$$v_{\text{growth}} \cdot \lambda^2 = \text{const} \quad (13)$$

in analogy with the relation which is valid for dendritic growth.

The structure will be coarse (large λ) when the growth is slow and fine (small λ) at rapid growth. The finer the structure is the better will the mechanical properties of the material be.

6.4.2 Relation between Growth Rate and Undercooling

For eutectic growth there is a relation between the growth rate v_{growth} and the undercooling corresponding to the one, valid for dendritic growth:

$$v_{\text{growth}} = \mu \cdot (T_E - T)^n \quad (14)$$

where

T_E = eutectic temperature of the melt

T = temperature of the melt at the solidification front.

μ and n are constants.

As concrete examples of eutectic alloys we will discuss silumin and cast iron in this section. Both have great technical importance and are used to a great extent at the production of castings.

6.4.3 Eutectic Structures of Silumin and Cast Iron

Eutectic Structures of Silumin

Aluminium-silicon alloys are widely used for producing various commercial products in foundries, for example engine blocks. Particularly the eutectic Al-Si alloy *silumin* is of great technical importance.

Figure 13 on next page shows the phase diagram of the Al-Si system. It contains a eutectic point at 12.6 w/o Si. The eutectic temperature is 577 °C. The Al-phase solves maximum 1.65 w/o Si while the solubility of aluminium in silicon is very low and can be neglected.

Figure 14 a shows a typical microstructure of a eutectic Al-Si alloy, which has been formed at a relatively low cooling rate. The structure consists of relatively coarse plates of Si imbedded in a matrix of Al-phase.

It can be seen in figure 14 a that these plates often have a broom-like shape. The reason for this is that the disk- and flake-shaped Si-crystals easily break at the solidification front, which results in crystal multiplication. The broken crystal fragments often turn somewhat before they grow. The result is that the silicon plates get divergent directions and the broom structure is formed.

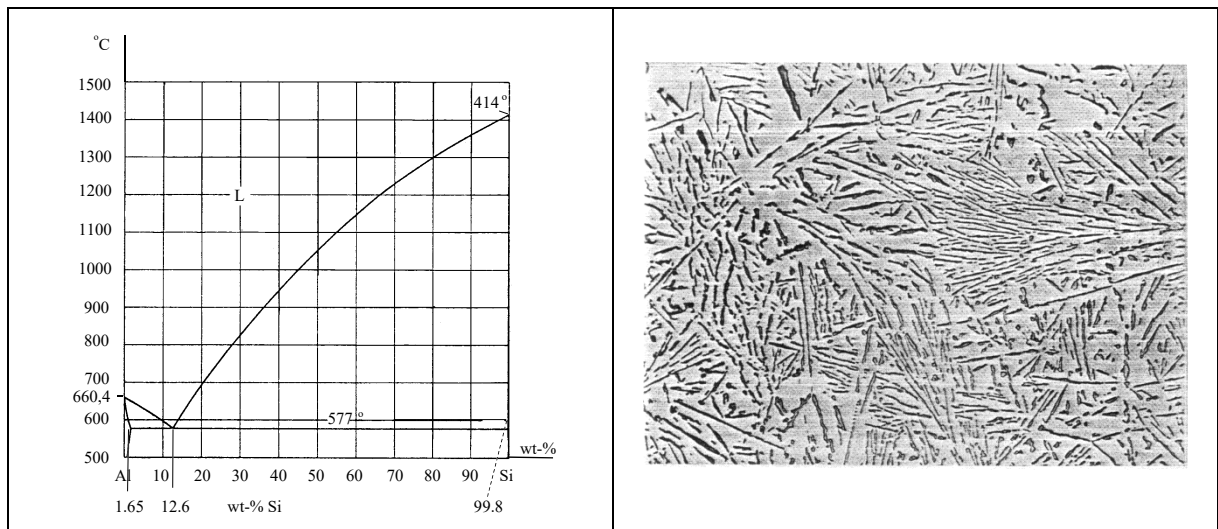


Figure 13.
Phase diagram of the system Al-Si.

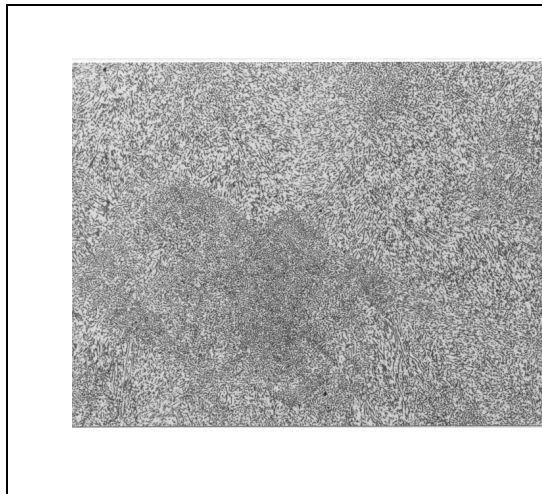


Figure 14 b.
Structure of modified eutectic silumin.

Figure 14 a.
Structure of unmodified eutectic silumin.

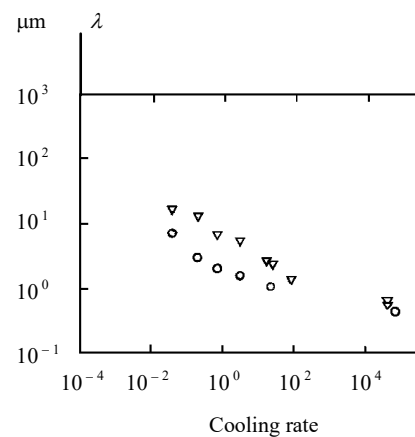


Figure 15.
The distances between the Si plates as a function of the cooling rate for :
▽ unmodified silumin
○ silumin, modified with 250 ppm Sr.

The American metallurgist Pacz discovered in 1920 that addition of small amounts of sodium (a couple of hundredths per cent) to a silumin melt before solidification changed the flaky, plate-like and branched microstructure of normal silumin into a much finer, more regular and fibrous microstructure (figure 14 b). The product is called *sodium-modified silumin*. If the solidification is rapid (quench modification) the structure of the alloy will be even finer, i. e. the lamella distances become smaller than those without inoculation. Modified silumin has better mechanical properties, ductility and hardness than plain silumin.

One complication with sodium modification of silumin is that the added sodium fades away rapidly, due to volatilisation and oxidation, and it is difficult to control the sodium content in the melt. For this reason strontium is used as an alternative modifier with almost the same improvement of the mechanical properties of the alloy that can be achieved by use of sodium.

The distance between the Si plates depends on the cooling rate and the extent of modification. The higher the cooling rate is the