

3

CASTING OF LIGHT ALLOYS

3.1 INTRODUCTION

The casting of metals is an art as well as a science. To be able to repeatably produce sound castings, each new casting configuration needs a combination of appropriate casting design software and the accumulated experience of knowledgeable foundry staff. This is because casting is a complicated combination of factors and parameters many of which cannot be independently varied. These include the complexity of the component to be cast, the mold materials, melt preheat and pouring temperatures, melt handling, runner and gating design, feeding capacity of the risers, alloy chemistry and shrinkage properties, melt treatment such as degassing and the addition of refiners and modifiers, and the quality and quantity of added recycled alloy. The process of solidification and development of the semisolid microstructure also impacts on the alloy's fluidity and the pathways for accessing the feed liquid and, therefore, on the formation of casting defects such as porosity and hot tearing.

Over the last two decades, the capability and accuracy of casting design packages for computer modeling and simulation of solidification processes has improved significantly. Applied initially to predicting heat transfer and freezing patterns, these software packages are now being used to model most aspects of the casting process, including the computer-aided design of molds, evolution of microstructures, and the estimation of thermal stresses generated during solidification. Software packages are also available to assist with gating and feeder design, and to describe fluid flow during mold filling. It is therefore possible to assess the performance of a casting process relative to the geometry of an engineered cast component before the first melt is poured. During this period, the direct observation of the actual solidification process has also become possible using techniques such as synchrotron X-ray tomography and 2D radiography that are providing new insights into the actual nucleation and growth of phases as the microstructure is developed. Nevertheless, unexpected outbreaks

of defects can still occur. Such occurrences are often blamed on the supplier of the alloy ingots being melted, or to small changes in casting parameters arising from deviations in process control, even though they may be within acceptable limits.

Both aluminium and magnesium alloys can be cast into components by a large range of technologies, the most common of which are sand casting, permanent mold (gravity die) casting, and cold- and hot-chamber pressure die casting. In 2014, about 16 million tonnes of aluminium were converted into cast products compared with 190,000 tonnes of magnesium. Only a small quantity of titanium is cast into components due to the high melting point and reactivity of titanium alloys. Control of the atmosphere or vacuum requires very expensive melting and casting equipment.

This chapter begins with a description of the solidification process from the molten state to the final as-cast solidified microstructure in the context of our current understanding of the factors affecting nucleation and growth of the phases that form during solidification. Next, factors that contribute to the ability of an alloy and casting process to produce sound products (referred to as castability) are considered followed by descriptions of the range of casting methods available for manufacturing cast light metal components. Details of the commercial aluminium casting alloys are presented in [Chapter 5](#), Cast Aluminium Alloys, and commercial magnesium alloys are presented in [Chapter 6](#), Magnesium Alloys. As the amount of titanium castings produced is so small, this chapter will focus on the casting of aluminium and magnesium alloys.

3.2 SOLIDIFICATION OF LIGHT ALLOYS

Most commercial light alloys are hypoeutectics (i.e., those alloys of composition less than the eutectic composition) with microstructures consisting of primary phase grains of aluminium ([Fig. 3.1](#)) or magnesium ([Fig. 3.2](#)) containing alloying elements in solid solution, which are surrounded by a eutectic phase. Ternary and quaternary eutectics may also form depending on the number of elements in the alloy. The morphology of the primary phase can vary from globular to rosette-like or dendritic grains depending on the composition, grain size, and casting conditions. Grain morphology also controls the distribution of the eutectic regions. As a better combination of strength and toughness is usually obtained by refining the microstructure, the grain size of the primary phase can be reduced by increasing the cooling rate or by adding a grain-refining master alloy. In Al–Si casting alloys, reducing the grain size has less effect on yield strength than refining the secondary dendrite arm spacing with which there is a Hall–Petch relationship. However, a reduced grain size usually improves castability and promotes a finer distribution of intermetallic particles and porosity which improves toughness and ductility. Finer grain size also has the advantage of reducing homogenization times when heat treating large wrought alloy billets.

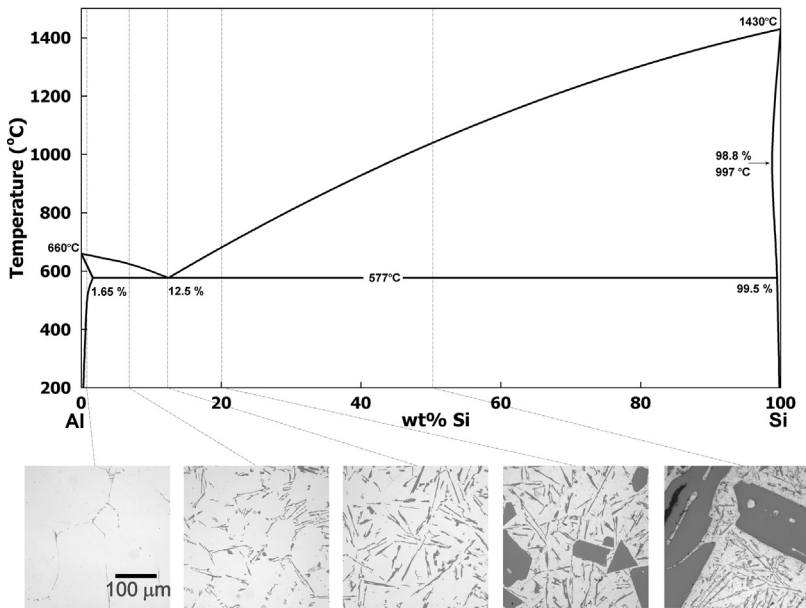


Figure 3.1 The aluminium–silicon phase diagram with associated microstructures showing the change in Si morphology for hypoeutectic (<12.5%), eutectic, and hypereutectic (>12.5%) silicon compositions. Courtesy S. McDonald, The University of Queensland.

A series of phase transformations occur during the solidification of a casting from the liquid state. In simple binary alloys such as Al–Si or Mg–Al (Figs. 3.1 and 3.2), the primary α -phase nucleates first and, as the temperature falls, the α -phase will grow until the eutectic temperature is reached at which point the eutectic nucleates consuming the remaining liquid. In ternary alloys such as the Al–Si–Cu alloys, a ternary eutectic will also form before solidification is complete. As practical casting conditions are nonequilibrium, the solidus line may be suppressed extending the two-phase region to lower alloy contents. This is particularly important in Mg–Al alloys (Fig. 3.2A) where the two-phase region is extended to about 2 wt% Al down from an equilibrium solubility limit of 12.7 wt% Al. The commercial 3, 6, and 9 wt% Al alloys fall within this nonequilibrium range of compositions. Fig. 3.2B shows representative high-pressure die cast microstructures of AM60 and AZ91. The white areas are β -Mg₁₇Al₁₂ and the lighter gray areas adjacent to the β -Mg₁₇Al₁₂ are high Al content (“eutectic”) α -Mg. The dark gray areas surrounded by the eutectic α -Mg are primary α -Mg dendrites.

Before discussing solidification in more detail, there are concepts that are critical to understanding the development of microstructure. The most important are thermal undercooling and the concept of constitutional undercooling (also referred to as constitutional supercooling) which play important roles in nucleation and grain formation of both the primary and eutectic phases.

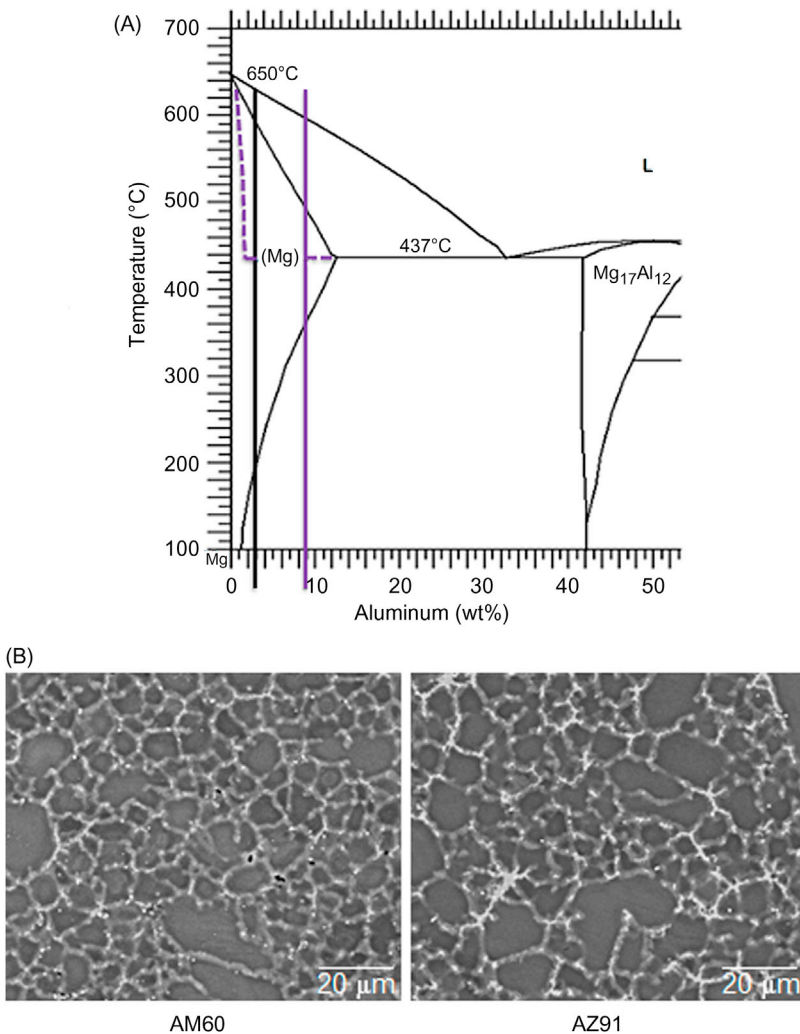


Figure 3.2 (A) The magnesium-rich end of the Mg–Al phase diagram with a dashed line indicating the suppression of the solidus line during nonequilibrium cooling. The vertical lines represent the range of commercial Mg–Al alloys (AZ31, AZ61, and AZ91). (B) Micrographs of pressure die cast AM60 and AZ91. From Zhu, SM et al.: *Metall. Mater. Trans. A*, 46, 3543, 2015.

Thermal undercooling promotes nucleation particularly near the mold wall of the casting where the cooling rate is higher. In this instance the thermally undercooled region near the mold wall will trigger nucleation of grains known as chill crystals or wall crystals, as illustrated in Fig. 3.3. The relative importance of thermal and constitutional undercooling on nucleation is described later.

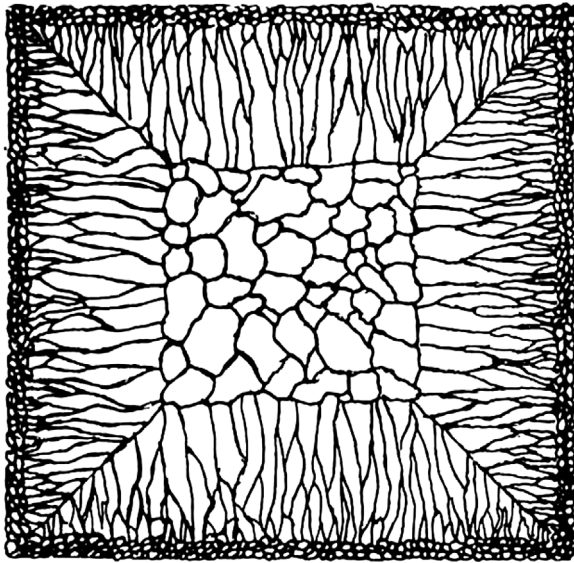


Figure 3.3 Schematic cast grain structure illustrating the transition from columnar to equiaxed grains. The finer grain structure adjacent to the mold walls is formed by chill crystals generated by the higher cooling rate at and/or near the mold walls. From Flemings, MC: *Solidification Processing*, McGraw-Hill, New York, NY, USA, 1974.

Fig. 3.4 is a representation of the development of constitutional undercooling in front of a growing equiaxed grain where the solute is rejected producing a concentration gradient within the diffusion field surrounding the grain (Fig. 3.4A). This concentration gradient can be converted to the gradient of the equilibrium liquidus temperature T_E as shown in Fig. 3.4B. Also, shown in Fig. 3.4B is the actual temperature, T_A . In the case of equiaxed solidification, the actual temperature gradient at the solid–liquid interface is slightly negative as latent heat is released through the liquid (Fig. 3.4C). The difference between T_E and T_A is the amount of constitutional undercooling generated by the grain's growth at that point in time. If there is a positive temperature gradient typical of directional solidification (Fig. 3.4B), then the amount of constitutional undercooling will decrease ahead of the interface eventually becoming zero when T_A is greater than T_E . For equiaxed solidification, constitutional undercooling is the dominant driving force for nucleation and, because nucleation is occurring within a constitutionally undercooled zone, the new grains are protected from remelting. Note that the maximum value of constitutional undercooling (Fig. 3.4C) is not at the solid–liquid interface but at the end of the solute diffusion field which is indicated by the vertical marker on the x -axis of Fig. 3.4.

Another important concept is the growth restriction factor designated by Q . Q is calculated from phase diagram parameters by $Q = C_0 m_1 (k-1)$, where C_0 is the alloy composition, m_1 is the slope of the liquidus line, and k is the partition

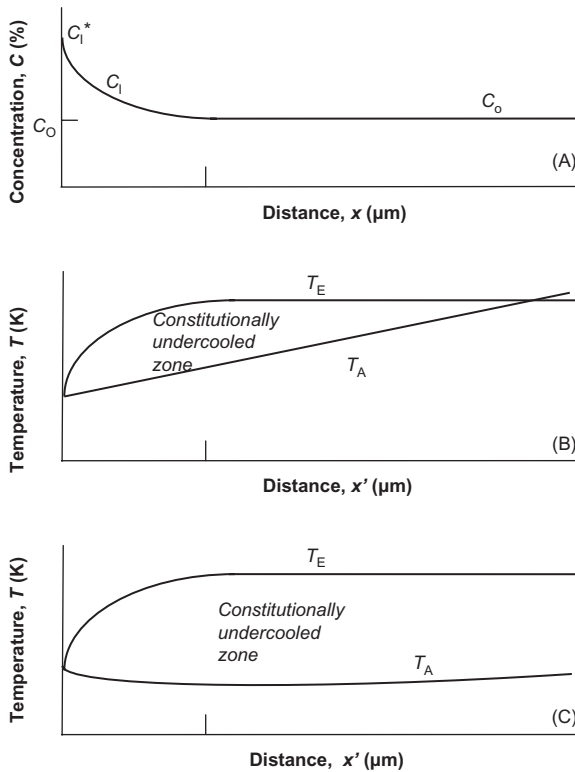


Figure 3.4 (A) A representation of the solute diffusion field in front of the growing solid–liquid interface. (B) The values of composition along the diffusion field in (A) converted to values of their equilibrium liquidus temperature. The constitutionally undercooled zone is the region between T_E and the actual temperature, T_A , due to heat extraction by the mold wall, and is typical of directional solidification. (C) As for (B) except in this case T_A is typical of equiaxed solidification where the gradient is negative at the solid–liquid interface due to heat flow from the hotter grains into the melt and then becoming positive due to heat flow out of the bulk melt into the colder mold walls.

coefficient equal to $C_s(\text{solidus composition})/C_l(\text{liquidus composition})$. The term “growth restriction” is used to describe the decrease in the rate of interface growth as the composition of an alloy, and thus Q , increases and also when the type of solute is a more strongly segregating element (i.e., a higher Q value). It was determined that Q also represents the rate at which constitutional undercooling develops. This has important consequences for interface stability and nucleation. The higher the value of Q , the sooner the solid–liquid interface will break down from a planar to a dendritic morphology, and the sooner the amount of constitutional undercooling reaches the nucleation undercooling of a particle in the melt such that another nucleation event will occur.

3.2.1 Grain formation

Nucleation of the primary phase Nucleation on heterogeneous sites, such as impurity or oxide particles and inoculant particles provided by master alloys, is most relevant to commercial casting. At present it is generally accepted that the potency of a particle is related to the degree of lattice matching or the solid–liquid interfacial surface energy between the substrate and the newly formed phase, and particle size where a larger size requires less undercooling to trigger nucleation and thus is a particle with a higher potency. The proportion of large particles that become active nucleants depends on the size distribution of the particles. For example, in the case of AlTiB master alloys added to aluminium alloys, only a small proportion (<2%) of particles are sufficiently potent to trigger nucleation. While the scientific understanding of the development of grain morphology and eutectic solidification are well described by theory, and can be simulated by numerical models, nucleation remains an area of considerable research and debate. New analytical techniques that allow investigation at the atomic level, along with new modeling approaches, are revealing new insights into atomic ordering at the substrate interface and it is expected that, over the next decade, new theories of nucleation will be developed. These developments will not be covered in this book and the following will discuss nucleation in the context of casting processes. Grain refinement will be discussed separately later in this chapter.

Depending on the casting conditions, nucleation in an alloy melt may initially occur on or near the walls of the mold triggered by thermal undercooling due to the melt being cooler than in the bulk of the melt, or may occur uniformly throughout the melt. The process of nucleation is affected by the casting conditions in several ways:

1. When the temperature gradient into the melt is strongly positive such as in directional solidification, the first “chill” crystals to form near the walls evolve into aligned columnar grains as shown in Fig. 3.3. In this case, the columnar grains continue to grow toward the center of the casting with no further nucleation occurring.
2. As the temperature gradient across the casting decreases due to thermal equilibration of the melt and heating of the mold, a columnar to equiaxed transition may occur as illustrated in Fig. 3.3, because constitutional undercooling becomes the dominant driving force for nucleation.
3. The aim of grain refinement or processes that lower the temperature gradient in the melt, as illustrated in Fig. 3.4C, is to promote equiaxed solidification and eliminate columnar growth altogether. Once equiaxed grains form, the temperature gradient from the solid–liquid interface of each grain into the liquid becomes slightly negative. This results in an increase in the amount of constitutional undercooling, thus, enhancing the likelihood of further nucleation and the survival of grains as they are transported into the melt.

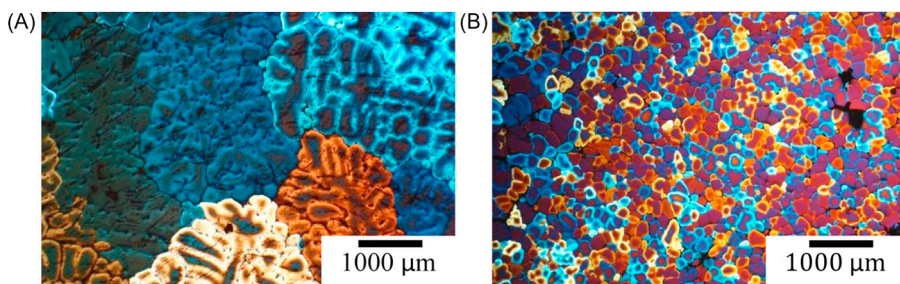


Figure 3.5 The microstructure of anodized samples from the central part of the Al-2Cu alloy ingot samples: (A) without ultrasonic treatment and (B) with ultrasonic treatment. Courtesy G. Wang, The University of Queensland.

4. An alternative mechanism occurs when the temperature gradient remains positive and solidification begins while the melt is turbulent due to, for example, pouring of melt into the die cavity. In this case, the grains nucleated on or near the mold walls are remelted by the hot metal being poured on top. Toward the end of filling the die, the wall crystals formed near the top of the walls of the casting survive and are carried into the bulk of the melt by convection. The macrostructure formed in this case is similar to case (2) above.

When nucleation readily occurs, the distance between grains becomes very small limiting the amount of grain growth. All grains begin with near spherical growth morphology and then the growth interface quickly transitions to globular (or cellular) and finally to a dendritic morphology as the grain continues to grow. If growth is interrupted by neighboring grains, the final as-cast morphology can be globular rather than dendritic which would be obtained when the growth distance is large. Fig. 3.5 shows that the macrostructure of an ingot of Al-2wt% Cu alloy has large dendritic grains when solidified in a normal casting environment. However, if ultrasonic treatment is applied, many more nucleation events occur resulting in a near spherical morphology. (These morphological changes can also occur for eutectic cells/grains that form during eutectic solidification, especially for alloys containing ternary elements.)

Nucleation of grains occurs on particles naturally present in the base alloy (e.g., impurity intermetallic particles, oxide films) or on deliberately added inoculant particles via the addition of master alloys. Recent synchrotron studies on Al-Si and Al-Cu alloys have shown that nucleation in grain refined alloys occurs in waves of nucleation events as the alloy cools with a low, positive temperature gradient across the melt, and each wave forms an approximately linear array of new grains (Fig. 3.6). Once these arrays of grains form, no further nucleation occurs between the initially nucleated grains. This implies that the grains are nucleating near a specific temperature (i.e., the nucleation temperature of the largest refiner particles) when the amount of constitutional undercooling reaches the nucleation undercooling. The distance between the arrays

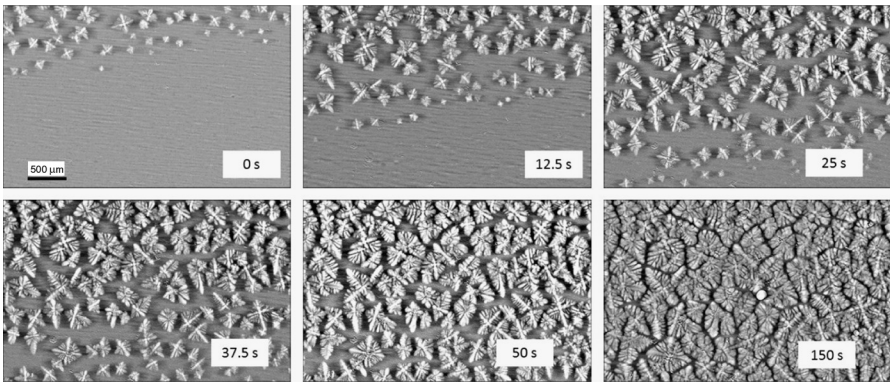


Figure 3.6 A sequence of images from real-time synchrotron X-ray observation of the solidification of an Al–15 wt% Cu alloy. From Prasad, A et al. *Mater. Sci. Eng.* 84, 012014, 2015.

of grains is affected by the formation of a nucleation-free zone (NFZ) by the diffusion field in front of the growing grains within the arrays ([Section 3.3.1](#)).

These new primary grains grow to be of spherical or globular morphology when the grain size is small (e.g., $<100\mu\text{m}$) or dendritic for larger grain sizes until they impinge on each other. [Fig. 3.6](#) shows typical dendritic images of an Al–15 wt%Cu alloy.

Grain growth and morphology For a directionally solidified alloy, the growth morphology of the solid–liquid interface can be changed from a planar growth front to a dendritic morphology by independently manipulating the rate of growth of the interface, V , and the temperature gradient, G , as illustrated by [Fig. 3.7](#). These morphological transitions are also affected by the cooling rate $\dot{T}(=GV)$, where higher cooling rates produce finer microstructures. The transitions represented in [Fig. 3.7](#) are affected by the alloy composition. For example, an alloy with a low solute content subjected to a steeper temperature gradient will promote a planar interface while the opposite promotes a dendritic morphology eventually allowing nucleation of equiaxed grains ahead of the growing interface. As the alloy content increases, the concentration gradient in the melt ahead of the solid–liquid interface becomes steeper promoting earlier formation of perturbations on the initially planar interface. These perturbations become stable generating cellular growth and at higher compositions, and lower temperature gradients, secondary and tertiary perturbations can occur on the primary and secondary arms, respectively, producing highly dendritic grain morphologies.

In the case of equiaxed solidification of grains, G and V cannot be controlled in the same way as for directional solidification. G is not constant over time from the start of solidification nor throughout the casting, and depends on the location in the melt as well as the complexity of heat extraction pathways.

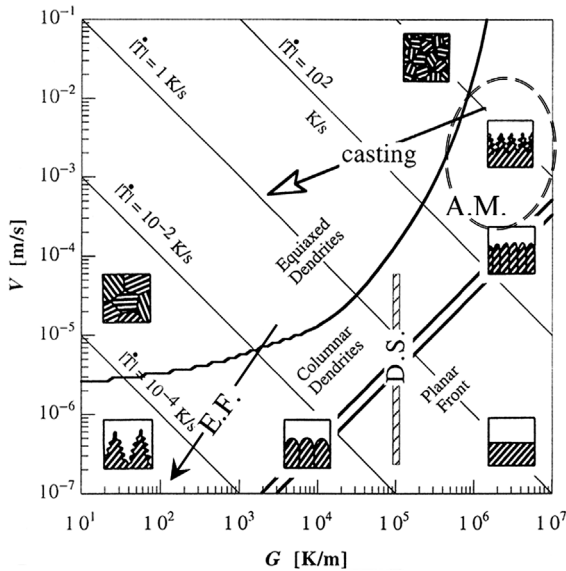


Figure 3.7 Schematic illustrating the change in grain morphology for a particular alloy composition due to changes in the interface growth velocity, V , and temperature gradient, G . The figure also shows the effect of cooling rate $\dot{T} (=GV)$ on the morphology which becomes finer at higher cooling rates. The base figure is from Kurz, W and Fisher, D.J.: *Fundamentals of Solidification*. Trans Tech Publications, 1998, modified by the addition of data for the A.M. region derived from Collins, P.C. et al.: *Annu. Rev. Mater. Res.*, 46, 63, 2016 and Marshall, G.J. et al.: *JOM*, 68(3), 778, 2016, and the data for region E. F. is from a recent study, courtesy of G.Wang, The University of Queensland.

The latter affects the local cooling rate which can vary significantly from place to place depending on the complexity of casting design. Therefore, G and/or V will also vary. Near the walls of the mold, G can be initially high and as the mold heats up G decreases. To complicate matters, the local temperature gradient at the grain–liquid interface will be slightly negative (Fig. 3.4C). However, the general relationships shown in Fig. 3.7 can provide a guide to the likely changes in the morphology of equiaxed grains at the location of interest during solidification of an alloy. Alloy composition also affects morphological changes. For example, for low alloy compositions, a decrease in G and/or an increase in V may change the grain morphology from spherical to globular/rosette (i.e., the equiaxed forms of planar and cellular solidification fronts, respectively, shown in Fig. 3.7) and then to dendritic. As the alloy compositions increase, the morphology will transition from globular to rosette to dendritic more rapidly.

The final morphology of a grain is also affected by the growth of the surrounding grains. If the growing grains are close together growth may be impeded before there is a transition from, for example, globular to dendritic.

This occurs because solute accumulates between the grains reducing V to very low levels which in turn promotes spherical or globular interfacial growth as predicted by Fig. 3.7. Thus, very fine grain structures can be almost spherical or globular while larger grain sizes of the same alloy will be fully dendritic. A simple example where a change in G during solidification causes a columnar to equiaxed transition is illustrated in Fig. 3.3 and by the “casting” arrow in Fig. 3.7, where the temperature gradient becomes lower with distance from the mold wall eventually promoting nucleation of equiaxed grains ahead of the columnar grains’ interface.

The four areas marked in Fig. 3.7 are related to different solidification processes as follows:

- The region D.S. represents the range of directional solidification conditions for producing a perfect homogenous crystal at the bottom of the region to single crystals with an internal dendritically cored structure at the top of the region.
- The direction of the “casting” arrow marked on the figure indicates the reduction in the cooling rate, and G and V , during solidification of the melt in a die cavity as the mold becomes hotter, which leads to a transition from a columnar to equiaxed grain structure as illustrated in Fig. 3.3.
- The area marked A.M. is for the conditions present during additive manufacturing. During additive manufacturing, both G and V are very high resulting in columnar dendrites.
- The arrow denoted as E.F. is for solidification when external fields such as ultrasonic and oscillating magnetic fields are applied. When external fields are applied to the melt, strong convection induced by acoustic streaming generates a very low value of G throughout the melt. In addition, the large number of grains swept into the melt after nucleation (Fig. 3.5 and see Section 3.6.3) causes V to become much slower due to solute accumulation between the grains thus producing near spherical or globular grains. The E.F. arrow is an estimate derived from thermal modeling and recent measurements undertaken during ultrasonic treatment of an Al–2wt% Cu alloy. Although the direction of the arrow indicates a change from equiaxed to columnar growth, once the equiaxed morphology is established, it remains to the end of the casting process because each grain is surrounded by many other equiaxed grains. As mentioned earlier, Fig. 3.7 has limited applicability to equiaxed solidification because the local temperature gradient at the interface is negative, a condition which is not taken into account by the figure. The calculated temperature gradients used here are the long range positive gradients due to heat being extracted by the mold walls. Thus, caution should be used when applying Fig. 3.7 to equiaxed solidification.

As the grains approach the point where they impinge on each other, the viscosity of the semisolid mush increases until mechanical interlocking of dendrites causes the movement of grains to cease. When the dendrites impinge

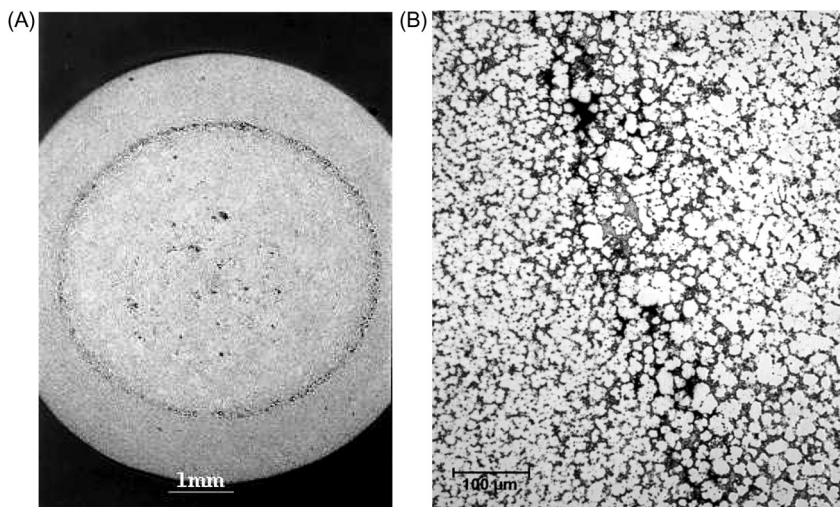


Figure 3.8 (A) A defect band containing segregation of eutectic and porosity in a circular section of a high-pressure die cast AM60. The flow direction is into the page. Note the very fine grain size on the outside of the defect band compared with the coarser microstructure inside. Higher magnification in (B) with the region outside the band on the left-hand side of the micrograph. From Dahle, AK et al.: *J. Light Metals*, 1 (1), 61, 2001.

on each other, a dendritic network is formed that exhibits a measureable shear strength which is dependent on the degree of mechanical interlocking between the grains. However, at this stage of solidification, the network of grains does not have any strength when placed under a tensile load. Magnesium high-pressure die castings (HPDC) can exhibit shear bands parallel to the flow direction of the molten metal during die filling (Fig. 3.8). These bands can be filled with porosity or tears. These shear bands are not necessarily a problem and may even improve fluidity, but the formation of cracks and interlinked porosity can lead to poor pressure tightness and reduced mechanical performance.

After grain impingement, the rate of dendritic growth decreases as the dendrite arms continue to thicken due to solidification and coarsening because of the driving force to reduce the surface area of the solid-liquid interface while the temperature drops to the eutectic temperature. During this stage, bridging across the dendrite tips of adjacent grains can occur. These bridges may tear under the tensile load caused by shrinkage. The eutectic then solidifies and, if feed liquid is insufficient, shrinkage can manifest as porosity or, when the dendritic network is under a tensile stress, hot tearing between the grains.

3.2.2 Eutectic solidification

Eutectic solidification begins when the melt temperature decreases to the eutectic temperature. For Al-Si alloys, a eutectic of α -Al and Si is formed at 577°C (Fig. 3.1) and for Mg-Al alloys a eutectic of α -Mg and $\text{Mg}_{17}\text{Al}_{12}$

is formed at 437°C (Fig. 3.2A). Because the primary phase is one of the two eutectic phases, the eutectic would be expected to nucleate on the surface of the primary phase as the barrier to nucleation should be low. However, this is not always the case as nucleation may depend on the barrier to nucleation of the other eutectic phase. Al–Si alloys are an example of this situation where formation of the eutectic first requires nucleation of silicon. In commercial purity alloys, the nucleation of silicon occurs independently of the α -phase—liquid interface on aluminium phosphide (AlP) particles. Once silicon forms, the two phases grow together with a morphology that is dependent on the relative volume fraction of the phases and whether one or more phases are faceted. Silicon grows as faceted flakes as shown in Fig. 3.1, which reduces the ductility of the alloy. To overcome this problem, a modifying agent (e.g., strontium) is added that converts the flakes to a fibrous morphology. Section 5.2.1 discusses the use of modifiers to refine the eutectic silicon microstructure of these alloys in order to improve their mechanical properties. These properties are also affected by the volume fraction of eutectic. Since the brittle silicon is contained within the eutectic, a higher volume fraction of eutectic tends to make the eutectic regions continuous leading to a decrease in fracture toughness of the material. The nucleation of silicon independently of the primary α -Al phase results in the formation of eutectic grains (also known as eutectic cells) in much the same way that the primary α -Al grains are nucleated. Once formed these eutectic grains grow through the interdendritic spaces of the α -Al grains until solidification is complete.

The magnesium–aluminium eutectic is a divorced or semi-divorced eutectic because $\text{Mg}_{17}\text{Al}_{12}$ forms a layer on the edge of the α -Mg dendrite arms rather than as part of a two-phase structure with α -Mg (Fig. 3.2B). Adjacent to this layer is an aluminium-rich region that may decompose during further cooling via discontinuous precipitation of $\text{Mg}_{17}\text{Al}_{12}$ forming a lamellar-like morphology with the aluminium-rich magnesium. Grain refinement of magnesium–aluminium alloys is usually the result of rapid cooling during high pressure die casting (HPDC), where the fine grain size allows the $\text{Mg}_{17}\text{Al}_{12}$ layer to be more or less continuous. Although a continuous $\text{Mg}_{17}\text{Al}_{12}$ layer improves corrosion resistance, it can reduce ductility.

3.3 GRAIN REFINEMENT

3.3.1 Factors influencing grain refinement

Grain refinement of the primary phase is a very important process for improving the properties and castability of aluminium and magnesium alloys. Recent studies have clearly shown that both potent particles and solute with a strong segregating ability (i.e., high Q value) are needed to promote effective nucleation. Each new nucleation event requires some growth of the previous grain to generate enough constitutional undercooling to trigger nucleation on the next most potent particle present ahead of the solid–liquid interface. On cooling, the

most potent of these particles nucleate the primary phase grains producing a finer grain size. This interdependence between solute and particle potency has been described by the Interdependence model where it was shown that there are three lengths that together establish the as-cast grain size (Fig. 3.9). They are the length x_{cs} the previous grain must grow to generate sufficient constitutional undercooling to trigger the next nucleation event, the length x'_{dl} , of the diffusion field, from the solid–liquid interface to the location where this critical amount of constitutional undercooling is established (Fig. 3.4), and the length x_{sd} from this point to where a suitably potent particle is present. Fig. 3.9 illustrates how these three lengths change with the inverse of alloy composition converted to its growth restriction factor Q . It can be noted from Fig. 3.9 that as Q increases the first two lengths, x_{cs} and x'_{dl} , decrease resulting in a reduced grain size.

As mentioned earlier, a higher Q value generates constitutional undercooling faster causing nucleation to occur sooner resulting in a finer grain size. Table 3.1 lists the values of $m_i(k - 1)$ for the alloying elements added to aluminium, magnesium, and titanium. The calculation of Q for complex alloys can often be undertaken by adding together the Q values for each alloying element. However, for some complex alloys, where there is interaction between the alloying elements, it is necessary to use thermodynamic software to calculate Q .

The first two lengths, x_{cs} and x'_{dl} , represent a nucleation-free zone (NFZ) surrounding each grain (Fig. 3.9), which is the sum of the distance a grain must grow to trigger the next nucleation event plus the length of the diffusion field around each growing grain where the amount of constitutional undercooling increases with distance ahead of the grain–liquid interface. Thus, close to the interface, the amount of constitutional undercooling is much lower than in the bulk liquid meaning that nucleation is most likely to occur away from the

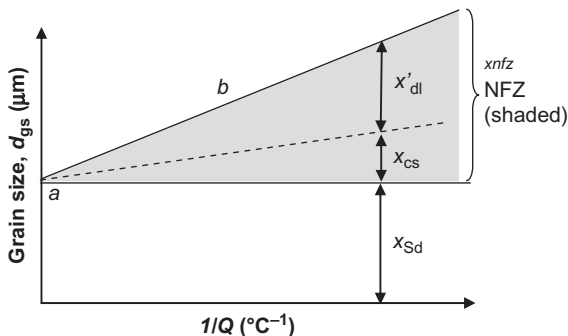


Figure 3.9 A simple representation illustrating that for each value of Q the grain size is the result of three components: x_{sd} is the average distance between activated particles assuming a constant inoculant particle density, and b is equal to the gradient of x_{cs} plus x'_{dl} over a unit of $1/Q$. The shaded area is the NFZ and represents the minimum grain size that can be achieved for each value of Q . From Easton, M et al.: *Curr. Opin. Solid State Mater. Sci.*, 20, 13, 2016.

Table 3.1 The values of $m_i(k-1)$ for the alloying elements added to aluminium, magnesium, and titanium for calculating the value of Q (i.e. $C_o m_i(k-1)$)

Al alloys	$m_i(k-1)$	Mg alloys	$m_i(k-1)$	Ti alloys	$m_i(k-1)$
Ti	~220	Fe	52.56	Be	72
Zr	6.8	Zr	38.29	B	66
Si	5.9	Ca	11.94	Si	21.7
Cr	3.5	Si	9.25	Ni	14.3
Ni	3.3	Zn	5.31	O	10.2
Mg	3.0	Cu	5.28	Co	8.8
Fe	2.9	Al	4.32	Y	7.9
Cu	2.8	Sr	3.51	Sc	6.7
Mn	0.1	Ce	2.74	Cu	6.5
		Y	1.70	Al	~0
		Mn	0.15	V	~0

The elements in bold have very high values of Q and are known to be excellent grain refiners.

growing interface rather than near it. The size of NFZ can make a significant contribution to the final grain size as illustrated in [Fig. 3.9](#).

Although the Interdependence model was developed for near equilibrium casting conditions, the linear relationship between grain size and the inverse of Q is often observed for casting processes involving dynamic conditions such as HPDC or when external fields such as ultrasonic treatments are applied during solidification.

3.3.2 Grain refinement by refinement methods and inoculation with master alloys

For aluminium alloys, grain refinement is well established with commercially available Al–Ti–B and, to a lesser extent, Al–Ti–C master alloys. These master alloys add TiB₂ or TiC particles and titanium solute into the melt. Additions of master alloy as low as 0.005 %Ti will result in grain refinement. The most common additions for aluminium alloys are Al5Ti1B and Al3Ti1B. In the Al–Ti–B master alloy, 2.2% of the titanium is combined with boron in the very stable TiB₂ particles. The remaining titanium is present as Al₃Ti intermetallic particles and in solid solution in the aluminium matrix up to 0.15%Ti. On addition to the melt, the Al₃Ti particles quickly dissolve providing titanium in solution. The higher amount of titanium in solution provided by Al5Ti1B is suitable for lean alloys (e.g., many wrought alloys) as the high Q values provided by titanium promote good refinement. Al3Ti1B is sometimes used in casting alloys as these alloys are already rich in growth restricting elements such as silicon.

Currently there is no reliable, safe, and cost-effective refiner for Mg–Al alloys which limits their use in sand and low-pressure die casting processes. Consequently, most magnesium alloys are cast by HPDC which produces a

very fine grain size without the need for refiner addition. For Mg–Zn alloys, zirconium is an excellent refiner for alloys that do not contain aluminium or manganese, because the zirconium dissolves providing high Q values and the zirconium particles that remain undissolved are potent nucleants for magnesium.

There are no commercial grain refiners available for titanium alloys. The common Ti6Al4V alloy has the distinction of having no growth restriction because both aluminium and vanadium have Q values of approximately 0 (Table 3.1). This combined with the fact that many elements and intermetallic compounds are soluble in liquid titanium, it has been difficult to identify stable intermetallic particles having good nucleation potencies. Significant grain size reduction is possible by adding solutes such as boron or silicon which have reasonably high Q values.

Aluminium alloys The Al–Ti–B master alloys are the most commonly used master alloys for achieving a fine grain size in commercial aluminium alloy castings. The reasons for their superior refining performance have been the subject of research for the last 80 years. Analyzing refinement across the broad spectrum of alloys was confusing as the performance of the refiner was better for direct chill (DC) cast wrought alloys (see Chapter 4, Wrought Aluminium Alloys) than for alloys suitable for foundry casting. As a result, a number of theories were developed such as the duplex, peritectic hulk, hypernucleation, and phase diagram related theories some of which seemed valid for wrought alloys but not valid for foundry alloys. These theories were created to satisfy two assumptions. One was that TiB_2 particles are relatively poor nucleants for aluminium and the other postulated that the peritectic reaction



is critical for improving the potency of TiB_2 . Both these assumptions are still subjects of debate.

The first assumption is correct but only in relative terms. TiB_2 is a reasonably good nucleant of aluminium but not as good as Al_3Ti . This assumption arose from the fact that adding TiB_2 particles to pure aluminium did not produce fine equiaxed grains while master alloy addition did result in refinement. It is now known that in order to produce equiaxed grains, we need alloy additions that segregate during solidification (i.e., they have a reasonable value of Q). When Al_3Ti is added along with TiB_2 in a master alloy, the Al_3Ti dissolves generating very high values of Q . Therefore, a direct comparison between the two particles is not possible. Because Al_3Ti is not an equilibrium phase below 0.15 wt% Ti, a number of theories postulated that Al_3Ti must form a metastable layer on the surface of TiB_2 for it to become a potent nucleant. Recent research suggests that rather than metastable Al_3Ti forming directly on the

surface of TiB_2 particles when the amount of Ti is less than 0.15%Ti, the liquid atomic layers adjacent to surface of TiB_2 crystals develop a structure with a higher density of titanium atoms that promotes the nucleation and formation of α -aluminium.

Regarding the second assumption, peritectic solidification does not need the peritectic reaction (Eq. (3.1)) to occur. Rather α -Al simply nucleates on the TiB_2 particles (or the pre-peritectic Al_3Ti) and grows as the temperature decreases like any other solid solution alloy. Further, the calculation of Q based on a value of $m_i(1-k)$ of about 220 clearly shows why titanium solute provides exceptional refinement.

Some solutes interfere with (poison) the ability of TiB_2 to facilitate grain refinement. Zirconium is one example and it is considered that atomic substitution occurs on the surface of TiB_2 becoming $(\text{Ti}_{1-x}\text{Zr}_x)\text{B}_2$ which changes its lattice parameters reducing the nucleation potency of the particles. Silicon, which is present in large quantities in several casting alloys, may also lead to coarser grain sizes if added to alloys refined by AlTiB master alloy. Two suggestions have been proposed to explain this behavior. One is that higher silicon contents produce narrower dendrite arms that grow faster leading to larger grains. The other proposal is that the TiB_2 particles may become covered by TiSi_2 thereby rendering them incapable of nucleating α -Al grains. However, silicon poisoning still occurs when there is no titanium present and when other intermetallic and oxide particles are used instead of TiB_2 particles. Thus, neither theory is a satisfactory explanation for the cause of silicon poisoning, and the actual cause still needs to be established.

Magnesium alloys Grain refinement is also very important for magnesium alloys where it improves most mechanical properties, and for many alloys, also improves creep and corrosion resistance. Quite different refining practices are needed depending upon the presence or absence of zirconium.

The group of alloys based mainly on the Mg–Al system tends to have large and variable grain size unless cast by HPDC. The first method devised to control grain size was to superheat the melt to a temperature of 850°C and above for periods of about 30 min, after which the melt was quickly cooled to the normal casting temperature and poured. A comparatively fine grain size was achieved with fair success. The probable explanation is that foreign nuclei with suitable crystal structures such as Al_4C_3 precipitate on cooling to the casting temperature and act as nuclei for the magnesium grains during subsequent solidification. The superheating effect is only significant in Mg–Al alloys and presents problems because crucible and furnace lives are reduced and power requirements are increased.

An alternative technique was developed in Germany in which a small quantity of anhydrous FeCl_3 was added to the melt (Elfinal process) and grain refinement was attributed to nucleation by iron-containing compounds. This

method also had its disadvantages because the deliquescent nature of FeCl_3 made it hazardous, and the presence of as little as 0.005% iron could decrease the corrosion resistance of the alloys. The addition of manganese was made to counter this latter problem but effectively prevented grain refinement by FeCl_3 .

The method in current use for alloys containing aluminium as a major alloying element is to add volatile carbon-containing compounds to the melt and hexachlorethane (0.025–0.1% by weight) is commonly used in the form of small briquettes which are held at the bottom of the melt while they dissociate into carbon and chlorine. Grain refinement has been attributed to inoculation of the melt with particles of Al_4C_3 , Al_2MgC_2 , $\text{AlN}\cdot\text{Al}_4\text{C}_3$ or magnesium oxide (MgAl_2O_4). However, rod-like substances rich in aluminium, carbon, and oxygen have now been detected inside the α -Mg grains which appear to serve as nucleation centers. In this regard, it has also been noted that the compound Al_2OC has lattice dimensions $a = 0.317\text{ nm}$ and $c = 0.5078\text{ nm}$ that are similar to magnesium ($a = 0.320\text{ nm}$ and $c = 0.520\text{ nm}$) and the same hexagonal crystal structure. Release of chlorine causes some degassing of the melt which is a further advantage of the method.

The ability of zirconium to provide excellent grain refinement (Fig. 3.10) of most other magnesium alloys can also be attributed to two factors related to the dissolved zirconium and the undissolved zirconium particles. The undissolved particles are the α -allotrope (stable below 862°C) which has a hexagonal crystal structure and lattice dimensions ($a = 0.323\text{ nm}$, $c = 0.514\text{ nm}$) close to those of magnesium. The zirconium solute provides a high level of growth restriction (high Q value, Table 3.1) which also promotes nucleation. These two factors imply that zirconium can nucleate magnesium during solidification and microprobe analysis has revealed the presence of zirconium-rich cores in the centers of magnesium grains due to the segregation of zirconium at the zirconium particle interface. The cores have formed as a consequence of peritectic solidification after nucleation on the zirconium particles as shown in the Mg–Zr phase diagram (Fig. 3.11). Both zirconium particles and zirconium solute make a significant contribution to the degree of grain refinement achieved.

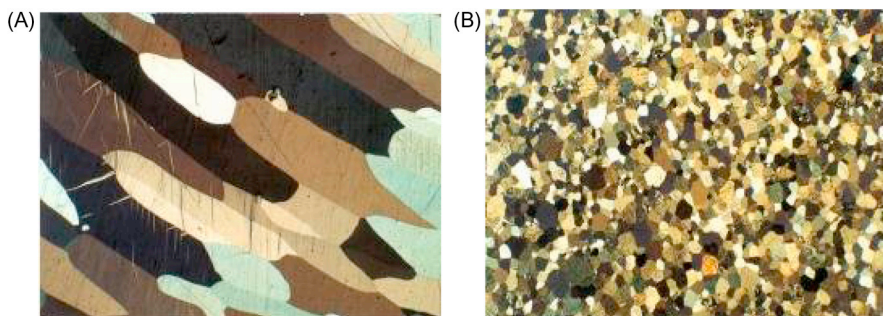


Figure 3.10 Microstructures of (A) pure magnesium and (B) magnesium with 1 wt% zirconium ($\times 50$ for each photo).

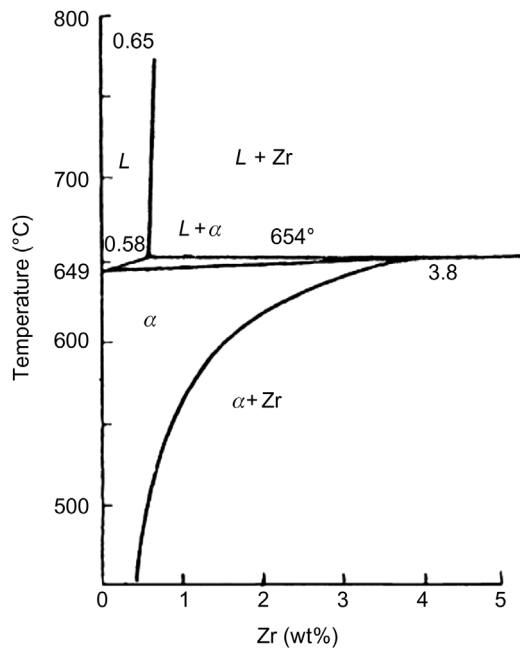


Figure 3.11 Section of the Mg–Zr phase diagram.

Undissolved zirconium particles tend to settle to the bottom of a crucible and, the longer the time elapsed after alloying with zirconium, the coarser will be the grain size of castings that happen to be poured from the top of a melt. If feasible, it is desirable to stir a melt before pouring. Because of the settling problem, it is also desirable to pour zirconium-containing alloy castings directly from the crucible in which they are melted. If transfer to another crucible is necessary, the alloy should be replenished with zirconium. Since the settling rate of the zirconium particles is dictated mainly by their size, attention is being directed to producing master alloys containing finer dispersions of particles. One possibility is to deform these master alloys, e.g., by rolling, in order to fragment the particles into smaller sizes.

Another potential problem is that zirconium will react preferentially with any iron present in the melt to form an iron–zirconium intermetallic compound. The resultant very low iron content in the melt will then provide a driving force for the rapid uptake of iron from the crucible, which is usually made of mild steel, if the temperature exceeds about 730°C. Practical solutions are to maintain a considerable excess of zirconium which increases costs, or to keep the melt temperature as low as possible and delay introducing the Mg–Zr master alloy until just before casting.

3.4 CASTABILITY

In general terms, an alloy can be described as being castable if it consistently and reliably produces castings that have sound microstructures and are

dimensionally accurate for a wide range of products, processes, and plants. The property of castability of an alloy is the sum of several factors of which the most notable are fluidity, mold-filling ability, volume shrinkage, susceptibility to hot tearing, porosity-forming characteristics, and surface quality. These factors are discussed below. Some depend strongly on chemical composition although mold design and process conditions are frequently more important. For example, aluminium foundries occasionally experience intermittent outbreaks of high reject rates because of porosity and shrinkage defects even though the chemical composition of an alloy appears to meet the required specification. In some instances, the batch of aluminium appears to be the culprit and it should be noted that several sources are used by foundries including:

- Primary metal supplied from a smelter as ingots that are provided in grades usually ranging from 99.5% to 99.85% aluminium, or as prealloyed ingots.
- Secondary metal ingots prepared by melting mixtures of recycled aluminium alloy products adjusted in composition to meet particular specifications.
- Molten metal delivered directly from a smelter or secondary producer to the foundry.
- In-house scrap returns.

The use of primary aluminium minimizes remelting and therefore reduces the formation of surface dross which is a mixture of aluminium metal, aluminium oxide, and other melt reaction products. Secondary metal can be cheaper, and certainly environmentally beneficial, to purchase because remelting of aluminium scrap consumes only about 5% of the energy needed to extract aluminium from its ore. However, impurity levels are inevitably higher and there is also a danger that undesirable elements may accumulate. An example is the buildup of iron from recycled scrap in Al–Si casting alloys that can lead to the formation of large intermetallic particles beyond a critical amount of iron resulting in the formation of damaging porosity. The additional remelting associated with the use of secondary metal also increases both the loss of aluminium to dross and the opportunity for the entrainment of oxide. Because of this, many foundries are adopting the practices of degassing and filtering melts prior to casting. Both magnesium and titanium have a strong propensity to oxidation and the melts need to be protected by a cover gas or vacuum respectively.

3.4.1 Fluidity

The fluidity of an alloy is commonly measured by pouring into a sand-molded spiral and alloys are compared by measuring the respective “running lengths” of travel before solidification occurs. Composition influences fluidity due to the effects of alloying elements on viscosity, surface tension, freezing range, and mode of solidification. Fluidity is reduced as the purity of aluminium

decreases and alloying elements are added. This is due mainly to a widening of the freezing range, the formation of intermetallic compounds, and to changes in the solidification pattern from a planar front to a mushy mode. This situation reverses close to a eutectic composition when fluidity is usually at a maximum.

Another aspect of fluidity is the ability to feed the interdendritic spaces remaining once the mold has been filled, but solidification is incomplete. As these spaces between the dendrites continue to narrow, melt viscosity and surface tension become more important and the pressure head needed to maintain flow increases. Volumetric shrinkage assists in this regard as normally there is an overall contraction during the change from liquid to solid.

3.4.2 Volumetric shrinkage

Total volumetric shrinkage during solidification also depends on alloy composition since each phase present has its own density characteristics. Shrinkage amounts to 6% for pure aluminium to as little as 1–2% for hypereutectic Al–Si alloys. Most alloying elements cause less change. Shrinkage is about 4% for magnesium alloys and varies from 3.5% for pure titanium to 5.5% for titanium alloys. Shrinkage may be evident as an overall contraction, as localized effects at surfaces, and at internal defects such as large isolated voids, interconnected porosity, or microporosity. The actual type that may occur in a casting depends on alloy composition, mold design, cooling rate, and mode of solidification. In pressure die castings, shrinkage will also occur if metal in the gate freezes before feeding of the casting has been completed.

3.4.3 Porosity

Most casting processes result in some internal porosity in a casting. This problem is usually most apparent in pressure die castings because of the speed of the operation, turbulent metal flow, and the opportunity to entrap gases. As mentioned earlier, it is because of the presence of porosity that aluminium die castings cannot normally be solution treated since this leads to surface blistering. In general, porosity in castings must be controlled and minimized so as to avoid detrimental effects on mechanical properties, pressure tightness, and surface appearance, particularly if machining is necessary. By attention to mold design and control of casting conditions, every effort is made to concentrate porosity in risers, or to redistribute it uniformly throughout a casting as less harmful micropores. Porosity arises primarily from shrinkage, the formation of internal oxide films, and from the entrapment of gases (air, steam, dissolved hydrogen, or products derived from the burning of organic lubricants). Porosity is usually worst in pressure die castings because of the highly turbulent metal flow and rapid solidification rates, although the entrapment of gases can be overcome by evacuating the mold prior to the entry of the molten metal. With regard to alloy composition,

porosity formation is closely related to the extent of the freezing range. Long ranges tend to result in the formation of dispersed, interdendritic, or intergranular micropores, whereas short freezing ranges may promote the formation of more localized regions of macro-shrinkage or larger intergranular pores. Porosity is also increased by the presence of intermetallic compounds with morphologies that may impede interdendritic feeding during casting. Examples are needles of the Al_5FeSi and $\text{Al}_{15}(\text{MnFe})_3\text{Si}_2$ compounds which have the appearance of Chinese script.

3.4.4 Hot tearing

Hot tearing (or hot shortness) arises when the tensile stresses generated within a solidifying casting exceed the fracture stress of the partially solidified metal. The result is usually evident as surface tearing and cracking which tend to occur at the sites of local hot spots where the casting is physically restrained. Generally, alloys are more susceptible if they have wide solidification ranges and low volumes of eutectics that can fill and repair hot tears. The change in susceptibility for a range of alloys is often referred to as a lambda curve because susceptibility initially increases from the pure metal as the alloy composition increases and then decreases after a critical composition is reached. This trend in susceptibility can also apply to ternary alloys. Fig. 3.12 presents a contour map of the hot tearing susceptibility factor for a range of ternary Al–Si–Mg alloys where a contour with a high value (and lighter color) of the susceptibility factor is very susceptible to hot tearing. The change in the hot

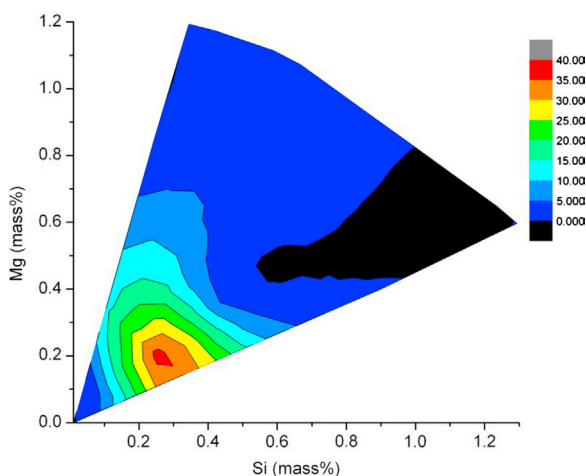


Figure 3.12 The measured hot tear crack width plotted as a contour map against the Si and Mg contents for a range of ternary Al–Si–Mg alloys. Larger value contours of a lighter color mean a higher susceptibility to hot tearing while lower value contours of darker colors mean there is a low likelihood of hot tearing. From Easton, MA et al.: *Metall. Mater. Trans. A*, 43, 3227, 2012.

tearing factor with alloy composition corresponds to a lambda curve and the maximum susceptibility occurs at about 0.15% Mg and 0.1% Si. Alloys based on the Al–Cu and Al–Mg system are most prone to hot tearing at similarly low Cu or Mg contents, whereas those based on Al–Si show the highest resistance to hot tearing.

A method of eliminating hot tearing of noncritical components is to promote porosity formation. In the distant past, potatoes were thrown into a melt to generate porosity during solidification. Another method is to generate more oxide films during pouring. The porosity reduces the need for feed liquid thus preventing hot tearing. On the other hand, where porosity needs to be eliminated, a study of gravity die cast aluminium automotive wheels showed that increasing the amount of Al–Ti–B master alloy led to large porosity and/or hot tears at the spoke–rim junction of the wheels. By gradually decreasing the amount of master alloy addition, the large localized porosity and hot tears gradually decreased being replaced by smaller internal porosity. The reason for this response is that the highly fined grains would flow into the melt and clog the mushy zone channel between fully solidified alloy against the spoke cavity walls placing considerable negative pressure on the hot spot at the spoke–rim junction which was cut off from the feed liquid, resulting in porosity and hot tears. The reduction in refiner meant fewer floating grains and a freer channel for liquid to flow into and feed the spoke–rim junction.

3.5 CASTING PROCESSES

3.5.1 Introduction

The first manufacturing step for most metallic products is a casting process. There are a variety of casting processes described in this section, many of which are used to cast aluminium and magnesium alloys. A typical aluminium producer will DC cast extrusion and rolling billets and remelt ingots on site to take advantage of producing product directly from the molten metal thereby removing a solidification and energy consuming remelting step. Magnesium producers will have ingot casting machines for producing foundry alloys nearby as most magnesium metal is then cast into shaped components via HPDC. Extrusions, sheet, and foil begin their manufacturing journey when the newly extracted molten metal is alloyed and then DC cast into billet or slab which is then extruded or rolled into semifabricated products. The alloys may also be cast by twin-roll or strip casting to reduce the need for many rolling passes to achieve thin sheet or foil. DC and ingot casting are described in more detail in Chapter 4.

Shape casting of components manufactured at foundries are produced from prealloyed remelt ingots which are cast by ingot casting machines. Sometimes molten-recycled aluminium is directly transported to foundries from recycling plants. The recycled melt is then blended with remelted casthouse ingots to

reduce the risk of casting defects. In general, alloys are classed as “primary” if prepared from new metal and “secondary” if recycled materials are used. Secondary alloys usually contain more undesirable impurity elements that complicate their metallurgy and often lead to properties inferior to those of the equivalent primary alloys.

The most commonly used casting processes are sand casting, permanent mold (gravity die) casting, and cold-chamber and hot-chamber pressure die casting. Sand molds are fed with molten metal by gravity except for low pressure and Cosworth precision casting processes where the melt is pumped upward into the mold cavity. The metal molds used in permanent mold casting are fed either by gravity or by using low-pressure air or other gas to force the metal up the sprue and into the mold. The following sections describe the casting processes commonly used in the commercial manufacture of aluminium and magnesium. Also covered are new casting processes that are used either for special applications or are under development and show promise for future adoption.

3.5.2 Melting

Aluminium alloys Aluminium and its alloys can be readily melted in air up to 750°C because of the protection of the tenacious surface oxide film and are therefore suitable for a wide range of casting processes. However, care needs to be taken in handling, pouring, and stirring the melt as new oxide films readily form whenever the protective oxide layer is broken. These films lead to the formation of dross which also traps aluminium and reaction products between the films producing waste material that is difficult to process and if left in the melt reduce the mechanical properties of the cast components. Graphite or refractory crucibles are suitable for melting aluminium but iron crucibles need to be coated with refractory materials to prevent dissolution of the iron into the melt.

Magnesium alloys It is usual for magnesium to be melted in mild steel crucibles for both the alloying and refining or cleaning stages before producing cast or wrought components. Unlike aluminium and its alloys, the presence of an oxide film on molten magnesium does not protect the metal from further oxidation. On the contrary, it accelerates this process. Melting is complete at or below 650°C and the rate of oxidation of the molten metal surface increases rapidly with rise in temperature such that, above 850°C, a freshly exposed surface spontaneously bursts into flame. Consequently, suitable fluxes or inert atmospheres must be used when handling molten magnesium and its alloys.

For many years, thinly fluid salt fluxes were used to protect molten magnesium which were mixtures of chlorides such as MgCl_2 with KCl or NaCl . In Britain, it was usual to thicken this flux with a mixture of CaF_2 , MgF_2 , and MgO

which formed a coherent, viscous cake that continued to exclude air and could be readily drawn aside when pouring. However, the presence of the chlorides often led to problems with corrosion when the cast alloys were used in service.

In the 1970s, it became usual to replace the salt fluxes with cover gases comprising either a single gas (e.g., SO_2 or argon), or a mixture of an active gas diluted with CO_2 , N_2 , or air. Sulfur hexafluoride (SF_6) was widely accepted as the active gas because it is nontoxic, odorless, colorless, and effective at low concentrations. Replacing the fluxes and SO_2 also improved occupational health and safety, and increased equipment life because corrosion was reduced. SF_6 is, however, relatively expensive, and is now realized to be a particularly potent greenhouse gas with a so-called global warming potential (GWP) of 22,000 to 23,000 on a 100-year time horizon. Efforts are therefore being made to find other active gases containing fluorine and one alternative is the organic compound HFC 134a (1,1,1,2-tetrafluoroethane) that is readily available worldwide because of its use as a refrigerant gas. It is also less expensive than SF_6 . HFC 134a has a GWP of only 1600, and an estimated atmospheric lifetime of 13.6 years compared with 3200 years for SF_6 . Moreover, less is consumed on a daily basis so that the overall potential to reduce greenhouse gas emissions is predicted to be 97%.

Most alloying elements are now added in the form of master alloys or hardeners. Zirconium has presented special problems as early attempts to use either zirconium metal or a Mg–Zr hardener were ineffective. Success was achieved eventually by means of mixtures of reducible zirconium halides, e.g., ones containing fluorozirconate, K_2ZrF_6 , together with large amounts of BaCl_2 to increase the density of the salt reaction products. These salt mixtures were supplied under license to foundries. Subsequently it was found possible to prepare hardener alloys from the weighted salt mixtures and these proprietary hardeners made in this way were used for adding zirconium to magnesium alloy melts. Prior to the use of BaCl_2 , severe problems were encountered with persistent flux inclusions which arose through entrainment of salt reaction products in the melt and could not be removed by presolidification or any flux-refining step.

Since the late 1960s, Mg–Zr master alloy hardeners have been the preferred method for introducing zirconium as a grain refiner and these master alloys usually contain between 10% and 60% of zirconium in weight percentage depending on the supplier. Since the approximate maximum solubility of zirconium in molten magnesium is only 0.6% (Fig. 3.11), almost all the zirconium present in the hardener is undissolved. Because of this, the Mg–Zr master alloys have microstructures comprising zirconium or zirconium-rich particles often embedded as clusters in an Mg–0.6Zr matrix. The major difference between the different master alloys lies in the size of the particles and how they are distributed in the matrix.

As with aluminium, hydrogen is the only gas that dissolves in molten magnesium although it is less of a problem in this case because of its comparatively

high solid solubility (average of $\sim 30 \text{ ml } 100 \text{ g}^{-1}$). The main source of hydrogen is from water vapor in damp fluxes or corroded scrap/ingot, so pickup can be minimized by taking adequate precautions with these materials. A low hydrogen content reduces the tendency to gas porosity which is common in Mg–Al and Mg–Al–Zn alloys and these materials should be degassed with chlorine. The optimum temperature for degassing is $725\text{--}750^\circ\text{C}$. If the melt is below 713°C , then solid MgCl_2 will form which gives little protection from burning, while at temperatures much above 750°C , magnesium losses through reaction with chlorine become excessive. Gas porosity is not normally a problem with zirconium-containing alloys since zirconium will itself remove hydrogen as ZrH_2 and it is generally unnecessary to degas these alloys. However, it should be noted that such a treatment does improve the tensile properties of certain Mg–Zn–Zr alloys, presumably by minimizing the loss of zirconium as the insoluble ZrH_2 . In such a case, the degassing operation is completed before zirconium is added.

3.5.3 High-pressure die casting

Large tonnages of pressure die castings are produced in both aluminium and magnesium alloys because the process is well established with high productivity. Approximately half of all aluminium alloy castings and about 80% of all magnesium alloy castings made worldwide are manufactured in this way and used in the automotive industry and for a wide range of other consumer goods. The castings produced are often housings and other components for nonload-bearing applications that do not require a high degree of structural integrity throughout.

A disadvantage with HPDCs is that they may contain relatively high levels of porosity. This restricts opportunities for using heat treatment to improve their properties because exposure to high temperatures may cause the pores to swell and form surface blisters (e.g., [Fig. 4.10](#)). For comparison, sand castings and low-pressure permanent mold castings generally contain less porosity and are used to produce components having more complicated shapes. They can then be heat treated if the alloys respond to age hardening. With permanent mold casting, turbulence can be minimized by introducing the molten metal into the bottom of the mold cavity, under a controlled pressure, thereby allowing unidirectional filling of the mold. This method has features in common with the Cosworth process for producing high-quality aluminium castings ([Fig. 3.22](#)).

Aluminium alloys In HPDC, molten aluminium is forced into a steel die through a narrow orifice (or gate) at high speeds ranging from 20 to 100 ms^{-1} ([Fig. 3.13](#)). This is achieved by means of a piston and cylinder (or hot sleeve) where the piston is driven by a hydraulic ram capable of exerting a pressure of up to 100 MPa on the metal. The aim is to continue feeding the casting as it solidifies

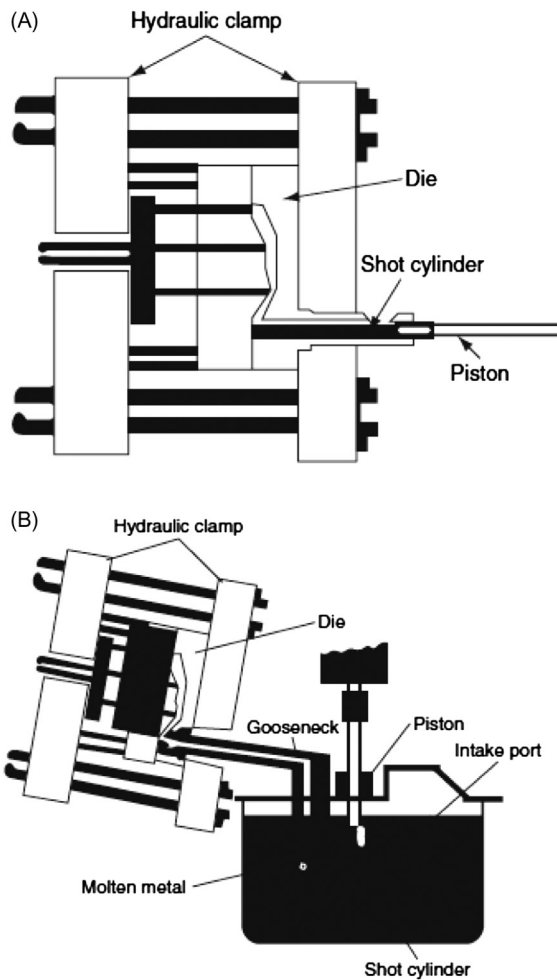


Figure 3.13 (A) Cold-chamber and (B) hot-chamber pressure die casting machines. Courtesy H. Westengen.

rapidly in the die. The following sections address two issues with alloy selection for die casting of aluminium alloys: die soldering and heat treatment.

Die soldering A particular problem with the pressure die casting of aluminium alloys is the tendency for the casting to stick (solder) to ferrous dies during solidification because of the high affinity iron and molten aluminium have for each other. Rapid interatomic diffusion can occur when they come into contact

resulting in the formation of a series of intermetallic compounds at the die surface. Die soldering may prevent automatic ejection of a casting which then has to be removed manually. This can cause delays and serious cost penalties, particularly if die surfaces are damaged and require repair.

Soldering is influenced by several factors:

1. *Die casting alloy composition.* Iron has the greatest beneficial effect and increasing amounts up to saturation level progressively alleviate soldering. For the widely used Al–Si casting alloys, the critical level of iron that is required can be estimated by the empirical relationship:

$$\text{Fe}_{\text{crit}} \approx 0.075 [\text{Si}\%] - 0.05$$

2. Manganese and small amounts of titanium are also beneficial, whereas the presence of nickel has been found to be detrimental. Magnesium, silicon, and strontium have no significant effect.
3. *Die steel composition.* The plain carbon steel H-13 (Fe–0.35C–0.35Mn) is commonly used. More highly alloyed compositions can better resist attack by molten aluminium but are often not used because of the cost of these steels.
4. *Melt injection temperature.* This should be as low as possible and an upper limit of around 670°C before the molten metal enters the shot sleeve has been found to be critical to minimizing sticking to some die steels.
5. *Use of die surface coatings.* Such coatings can serve as diffusion barriers to the iron–aluminium reaction and thereby retard soldering. A number of coatings are used by the die casting industry and boron-based ceramics have proved particularly effective. Boron nitride deposited by physical vapor deposition (Section 8.7) also provides a barrier providing it is compatible with the steel substrate. An alternative method is to aluminize the surface by immersing the die in molten aluminium at 760–800°C after which it is removed, held for 12–24 h at 300°C to form a controlled layer of iron aluminide, and then the surface is highly polished.

Heat treatment Two features of conventionally produced HPDCs are the extreme turbulence experienced by the molten alloy as it is forced at high speed into a die, and the very rapid rate at which it solidifies. Because of this, the resulting castings usually contain internal pores in which gases such as air, hydrogen, and vapors that are formed by the decomposition of organic die wall lubricants become entrapped. Porosity may also result from metal shrinkage that occurs during solidification. Whereas it is normal to accept some level of porosity in HPDCs, the presence of internal gas-filled pores makes it undesirable to expose them to high temperatures. This follows because the pores expand and may cause both unacceptable surface blistering on components and dimensional instability due to swelling.

As given in Table 5.2, several aluminium alloys used for producing HPDCs are based on the systems Al–Si–Cu and Al–Si–Mg, both of which are amenable to strengthening by age hardening. However, this opportunity has not been exploited because conventional practice has required the alloys to be solution treated for several hours at temperatures around 520°C prior to quenching and ageing. Recently, however, it has been recognized that most of the solute elements copper, magnesium, and to some extent, silicon, dissolve in the α -aluminium grains during the early part of the solution treatment cycle. Because of this, it has been demonstrated that reducing both the solution treatment temperatures and times has enabled commercial HPDCs made from the above classes of alloys to undergo significant age hardening without causing blistering or swelling.

One example of a modified ageing schedule is to solution treat these HPDC alloys at 480°C for 0.25 h, quench into cold water and age to peak strength at 175°C (T6 temper). Fig. 3.14 shows results for HPDC aluminium

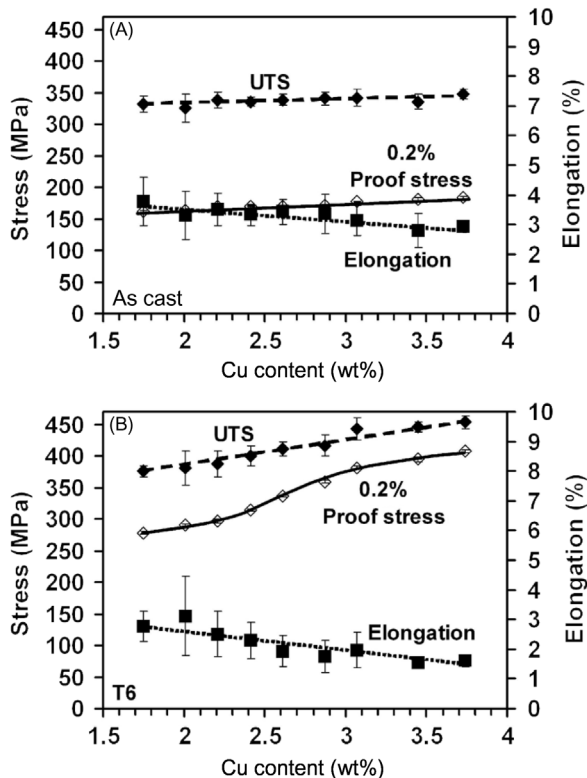


Figure 3.14 Effects of increasing levels of copper on the tensile properties of an HPDC alloy with nominal composition Al–10.5Si–0.8Fe–0.75Zn–0.22Mg–0.18Mn in the (A) as-cast and (B) T6 age hardened conditions. From Lumley, RN: *Fundamentals of Aluminium Metallurgy*, p. 283, Woodhead Publishing Ltd., Oxford, 2011.

alloys containing 10%Si and 0.7%Mg with Cu levels ranging from 1.7% to 3.7%. Comparisons between the tensile properties in the normal as-cast condition, and after applying this modified age hardening heat treatment, have shown that the 0.2% proof stress can be more than doubled if these HPDC alloys contain more than 2.7% copper. With the absence of surface blistering and swelling, there is the prospect of achieving significant weight savings by redesigning HPDC components to take account of their much higher mechanical strength.

Magnesium alloys Most magnesium alloy components are now produced by HPDC (Fig. 3.13). Cold-chamber machines are used for the largest castings and molten shot weights of 10 kg or more can now be injected in less than 100 ms at pressures that may be as high as 150 MPa. Hot-chamber machines are used for most applications and are more competitive for smaller casting sizes due to the shorter cycle times that are obtainable. Magnesium alloys offer particular advantages for both these processes, namely:

1. Most molten alloys show high fluidity which allows casting of intricate and thin-walled parts, (e.g., 2 mm, Fig. 3.15). Magnesium may be used for castings with thinner walls (1–1.5 mm) than is possible with aluminium (2–2.5 mm) or plastics (2–3 mm).
2. Magnesium has a latent heat of fusion per unit volume that is two-thirds lower than that of aluminium. This means that magnesium castings cool more quickly and die wear is reduced.
3. High gate pressures can be achieved at moderate pressures because of the low density of magnesium.

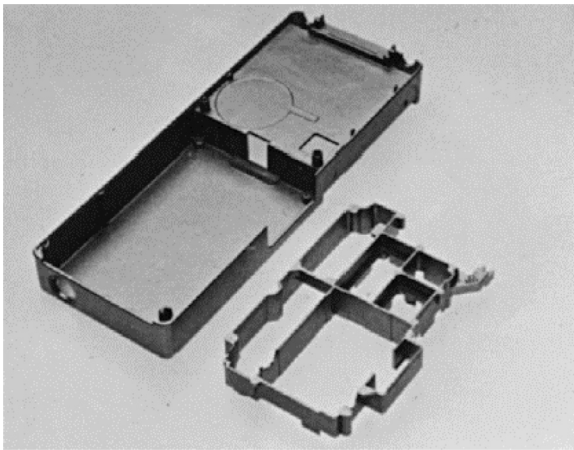


Figure 3.15 Thin-walled magnesium alloy case and chassis for a hand-portable cellular telephone. Courtesy Magnesium Services Ltd, Slough, UK.

4. Iron from the dies has very low solubility in magnesium alloys, which is beneficial because it reduces any tendency to die soldering.

3.5.4 Semi-solid processing

As metals and alloys freeze, the primary dendrites normally grow and intermesh with each other so that, as soon as a relatively small fraction (e.g., 20%) of the melt has frozen, viscosity increases sharply and flow practically ceases. If, however, the dendrites are broken up by vigorously agitating the melt during solidification, reasonable fluidity persists until the solid content reaches as much as 60%. Each fragment of dendrite becomes a separate crystal and very fine grain sizes can be achieved without recourse to the use of grain refiners (Fig. 3.16). Moreover, the semi-solid slurry has thixotropic characteristics, in that viscosity decreases on stirring, and this has interesting implications for casting processes.

In early laboratory studies of semisolid processing, a slurry was produced by agitating an alloy in a narrow, annular region between the furnace wall and a central cylindrical stirrer. A temperature difference was maintained so that the alloy within this region was in the two-phase semi-liquid state, while above it was fully molten. The slurry was then transferred and cast in a pressure die casting machine. The process was known as rheocasting. Slugs of the slurry could also be removed and forged in a die (thixoforging).

Commercialization of the slurry production process presented two major problems. One was the design of a furnace-stirrer system that would provide adequate quantities of slurry. The other was rapid chemical attack and

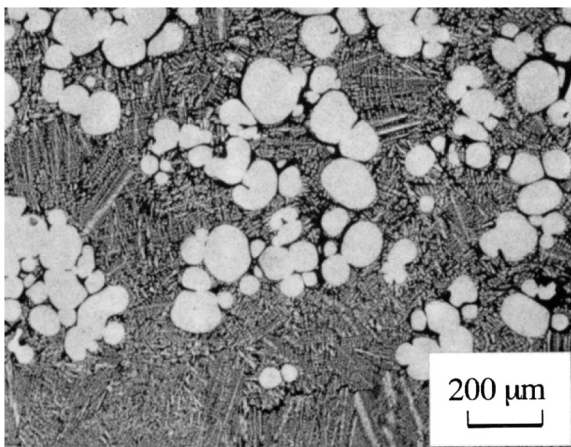


Figure 3.16 Microstructure of the alloy Al-6.5Si sheared at a rate of 180 s^{-1} showing rounded α -Al dendrite fragments of the primary α -Al phase in a matrix of finely dispersed eutectic. From Ho, Y et al.: *Nature and Properties of Semi-Solid Materials*, Sekhar, JA and Dantzig, J (Eds.), TMS, Warrendale, PA, USA, p. 3, 1991.

erosion of materials used for the stirrer. These limitations were overcome with the development of the thixomolding process which was begun by Merton Flemings and his students at MIT and then further developed by the Dow Chemical Company. The technology (Fig. 3.17) is now marketed by Thixomat Pty Ltd, and applied to magnesium alloys predominantly AZ60 and AZ91D. In 2016, there are more than 480 machines worldwide in 13 countries, which range in clamp tonnages from 100 to 1600 tons enabling very large magnesium components such as large screen TV surrounds, to be cast in one piece.

Magnesium alloys are amenable to thixotropic casting which offers the opportunity to produce high-quality fine-grained products more cheaply than by high-pressure die casting. The mechanical properties of thixomolded parts compare well with those obtained with standard HPDC.

Thixomolding is a one-step process which combines plastic injection molding technology with metal pressure die casting (Fig. 3.17). Feedstock in the form of solid alloy pellets is heated to the semisolid temperature ($\sim 20^\circ\text{C}$ below the liquidus temperature) to achieve 5–15% solid fraction and then sheared by a high torque screw drive) into a thixotropic state in a screw feeder before passing directly into the die casting machine. Very thin sections down to 0.7 mm thickness can be cast. An example of a thin-walled casting produced by thixomolding is shown in Fig. 3.18. An advantage of this process is that oxidation of the molten alloy is easily prevented by maintaining an inert atmosphere in the relatively small entry chamber (Fig. 3.17). Other advantages include reduced microporosity, prolonged die lives because heating and cooling cycles are less extreme, and there is no need for a furnace to melt the magnesium alloys prior to casting.

It has been demonstrated that thixoforming of rechipped sprues, gates, runners, and scrapped parts results in mechanical properties comparable to virgin

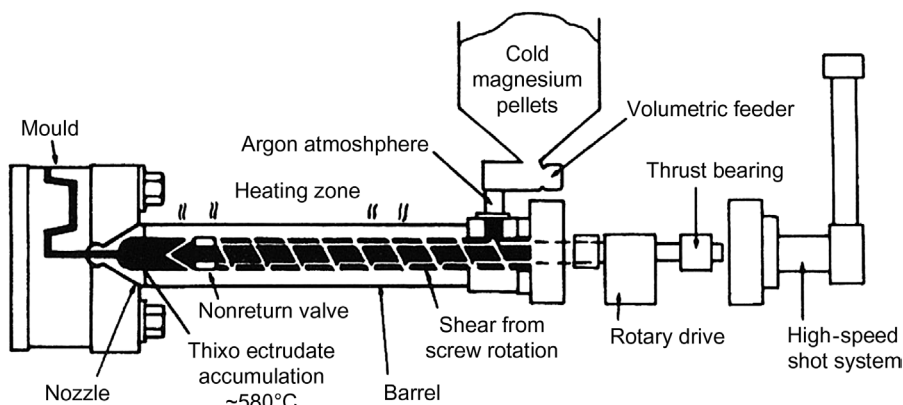


Figure 3.17 Schematic of a thixomold machine for producing die cast magnesium alloy components. Courtesy Thixomet Inc.

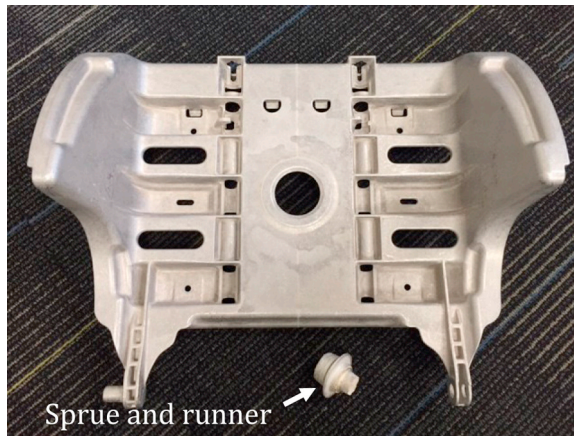


Figure 3.18 Child's safety auto seat manufactured by Molded Magnesium Products, 1.4kg part thixomolded with a hot long nozzle enabling miniature sprue and runner and 95% metal yield, 430mm × 380mm × 178mm, 2–5mm section thickness. Courtesy Ray Decker, Thixomat, Inc/nanoMAG LLC.

Table 3.2 Comparison of thixomolded AZ91D from recycled scrap to that from virgin chips

% Recycled scrap	YS (MPa)	UTS (MPa)	Elongation (%)
100	167	254	6.1
50	167	263	6.5
10	162	256	7.0
0 (virgin)	145–169	230–299	6–8

Courtesy Ray Decker, Thixomat, Inc/nanoMAG LLC.

YS, yield strength, UTS, ultimate tensile strength.

granules (Table 3.2). Also, high-quality melt-recycled magnesium alloy chips are now available in commercial quantities.

Semi-solid processing offers the following advantages:

1. Significantly lower capital investment and operating costs when compared with conventional casting methods. The whole process can be contained within one machine so that the need for melting and holding furnaces as well as melt treatment are all avoided. Foundry cleanliness is easy to maintain and energy requirements are less because complete melting is not required, cycle times are reduced, and scrap is minimized.
2. Shrinkage and cracking within the mold are reduced because the alloy is already partly solidified when cast.
3. Lower operating and pouring temperatures lead to an increase in the life of metal dies.

4. Composite materials can be readily produced by adding fibers or other solid particulates into the feedstock (compocasting).

The uptake of rheocasting and thixocasting has been limited but examples of automotive components in current production are master brake cylinders, and pistons and compressors for air conditioners where high structural integrity and pressure tightness is required. Thixomolding has been more successful commercially especially for magnesium alloys. It is ideal for the casting of thin components such as mobile phone and laptop computer casings and automotive components (e.g., [Fig. 3.18](#)).

3.5.5 Squeeze casting

This process involves working or compressing the liquid metal in a hydraulic press during solidification which effectively compensates for the natural contraction that occurs as the liquid changes to solid. Pressures around 200 MPa are used which is several orders of magnitude greater than the melt pressures experienced in conventional foundry practice. As a consequence, the flow of the melt into incipient shrinkage pores is facilitated and entrapped gases tend to remain in solution. The high pressure also promotes intimate contact between casting and mold walls or tooling which assists heat extraction thereby leading to refinement of the microstructure. Squeeze casting has also been used to prepare higher-quality castings from existing alloys and to produce castings in alloys that could not be successfully cast by conventional processes.

Squeeze casting may be carried out in what are known as direct and indirect machines. In direct squeeze casting, metered amounts of molten metal are poured into a die similar to that used in permanent mold casting and then pressure is applied to the solidifying metal via the second, moving half of the die. For the indirect method, molten metal is first poured into a shot sleeve and then injected vertically into the die by a piston which sustains the pressure during solidification. In both cases, a high casting yield is obtained because runners and feeding systems are not required so that virtually all the molten metal enters the die cavity. Casting mass may vary from 200 g to 35 kg depending upon the capacity of the machine.

Squeeze casting has been used for half a century in Russia and is now being exploited in other countries, notably Japan. For example, the Toyota Motor Company in Japan introduced squeeze-cast aluminium alloy wheels into their product line for passenger cars in 1979. An indication of the potential of this technique is evident from [Table 3.3](#), in which a comparison of the properties of the alloy 357 (Al–7Si–0.5Mg) prepared by sand, gravity die, and squeeze casting shows the latter clearly to be superior. Here the properties in each case are influenced mainly by the different levels of porosity, with improvements in ductility being particularly notable. GKN have developed their squeeze-forming process for the manufacture of aluminium wheels for demanding defense applications. The squeeze-formed wheels are 35% lighter than standard cast wheels and offer increased endurance four times that of a typical traditionally forged steel armoured fighting vehicle wheel.

Table 3.3 Mechanical properties of 357 aluminium alloy produced by different casting processes

Process	0.2% Proof stress (MPa)	Tensile strength (MPa)	Elongation (%)
Sand cast	200	226	1.6
Chill cast	248	313	6.9
Squeeze cast	283	347	9.3
Cosworth	242	312	9.8

From Lavington, MH: *Metals Mater.*, 2, 713, 1986.

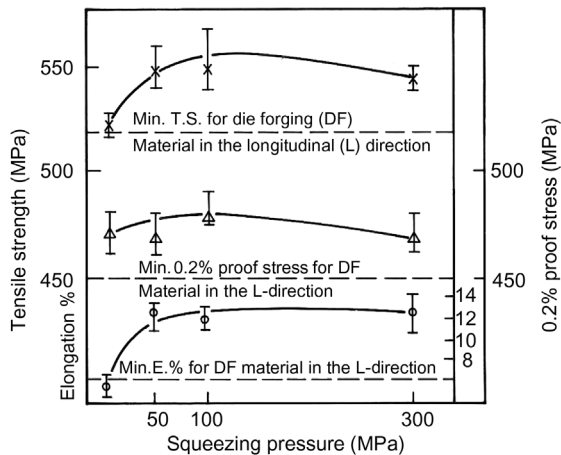


Figure 3.19 Tensile properties of squeeze-cast alloy 7010 as a function of squeeze-casting pressure. From Chadwick, GA: *Metals Mater.*, 2, 693, 1986.

Another advantage is that it is possible to squeeze cast some complex, high-strength alloys that are normally produced only as wrought products. For example, the Al–Zn–Mg–Cu aircraft alloy, 7010, has been successfully squeeze cast and, after a T6 ageing treatment, the minimum required properties for the wrought (die forged) condition are easily met (Fig. 3.19). A feature of the squeeze-cast material is that the casting has isotropic properties, whereas wrought products always suffer from directionality effects (see Fig. 2.42). What is also interesting is that the *S–N* curves obtained in fatigue tests on squeeze cast 7010 lie within the scatter band for wrought materials, whereas castings normally have much poorer performance (Fig. 3.20).

Properties may be further enhanced by incorporating fibers or particulates into squeeze-cast alloys, thereby producing the metal matrix composites that are discussed in the Section 8.1.3. In this regard, squeeze casting is superior to other low-pressure techniques for infiltrating liquid metal into a mesh or pad of fibers which has been a major limitation in the past. An advantage of such composites is that the fibers can be confined to strategic locations in castings where there is a need for increased strengthening or wear resistance. One recent example is the fiber reinforcement of regions of aluminium alloy pistons for cars that are now manufactured commercially.

3.5.6 Gravity and low-pressure casting processes

Gravity and permanent molds are usually machined or cast from steel or iron, that are fed either by gravity or by using low-pressure air or other gas to force the metal up the sprue and into the mold. Sand cores may be added to make hollow castings or cooling/air channels. Gravity casting is often used for short runs in castings for aftermarket or replacement components or special one-off components. For large campaigns of hundreds of castings, low-pressure die casting (Fig. 3.21) is used where consistent properties are required for

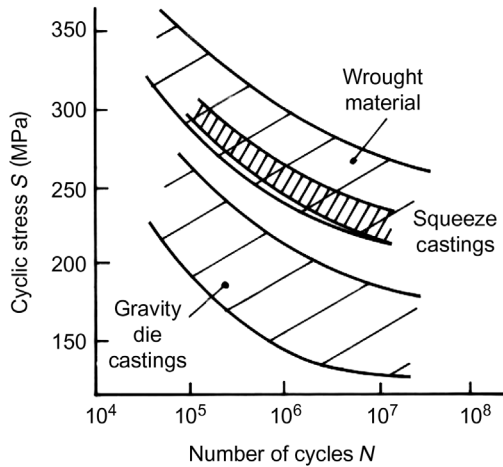


Figure 3.20 Fatigue (S - N) curves for alloy 7010 in wrought, gravity die cast, and squeeze-cast conditions. From Chadwick, GA: *Metals Mater.*, 2, 693, 1986.

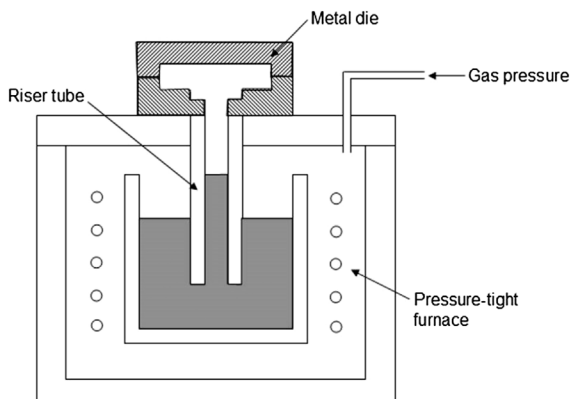


Figure 3.21 Schematic of low-pressure die casting machine. From *Fundamentals of Aluminium Metallurgy: Production, Processing and Applications*, p. 150, Lumley, R. (Ed.), Woodhead Publishing Limited, Oxford, 2011.

load-bearing applications such as aluminium wheels. Low-pressure die casting allows controlled flow of the molten metal with low turbulence to enter the mold from a holding furnace underneath the mold. Although productivity is significantly lower than HPDC, the quality of the castings is far superior. The metal dies need to be monitored for wear and attached dross to ensure a good surface finish is achieved. Casting design is also crucial to ensure solidification occurs toward the hottest part of the casting so that all parts of the casting are fed to account for shrinkage. The holding furnace below the mold needs attention to ensure the melt is free from dross and does not accumulate too much grain refiner such as TiB_2 particles which tend to sink in the relatively quiescent environment.

3.5.7 Sand and precision sand casting processes

Sand casting Sand casting is one of the oldest casting methods and the subject is so large that only a brief description will be given here. Many ferrous and aluminium alloys are cast in sand molds. The process has been developed with various degrees of automation, but the traditional practice is still used for short runs and prototyping.

A mold pattern consisting of the die cavity, pouring basin, sprue, sprue well, runner and gates, and risers for feeding is often made of wood and designed so that sand can be readily packed around the pattern. The sand is usually processed with binders to give it strength and plasticity. Once the sand is sufficiently strong, the patterns are removed and the sand molds are put together for casting. Once the melt has been poured into the mold and the casting has solidified, the sand is removed and recycled to various degrees depending on how much sand is damaged/fractured by the high temperatures near the surface layers of the mold cavity.

Magnesium engine blocks have been made for some high-performance vehicles. However, due to the reactivity of magnesium, many precautions are needed which make the process unsuitable for most applications other than for highly complex casting designs and low numbers. Magnesium alloys can be successfully sand cast providing some general principles are followed which are dictated by the particular physical properties and chemical reactivity of magnesium.

1. Suitable inhibitors must be added to the molding sand in order to avoid reaction between molten magnesium and moisture which would liberate hydrogen. For green sand or sand gassed with carbon dioxide to aid bonding, sulfur is used, whereas for synthetic sands compounds such as KBF_4 and KSiF_6 are also added. Boric acid is also used for some sands, both as a molding aid and as a possible inhibitor through its tendency to coat the sand grains.
2. Metal flow should be as smooth as possible to minimize oxidation.

3. Because of the low density of magnesium, there is a relatively small pressure head in risers and sprues to assist the filling of molds. Sand molds need to be permeable and must be well vented to allow the expulsion of air.
4. Magnesium has a relatively low volumetric heat capacity which necessitates provision of generous risers to maintain a reservoir of hotter metal. Because of this need to make special provision by way of risers and other feeding devices, the volume ratio of poured metal to actual castings may average as much as 4:1 for magnesium alloys.

A variation on mold pattern manufacture was to use polystyrene which volatilizes when in contact with the melt. This process is referred to as lost foam or evaporative casting. This method enables the pattern to remain in situ during casting but a downside has been the turbulence caused by the conversion of the foam into gas. Recent advances have led to 3D printing of the pattern which is a major advantage for the manufacture of complex mold patterns and is a cost-effective method of undertaking prototype casting trials.

For complex castings such as an engine block with thin cooling channels, sand cores are made and placed within the mold cavity at appropriate locations. Due to their delicate nature, they need to be strong enough for placement and to resist metal flow. For mass production, this process is too labor intensive and better control and automation are needed to produce complex castings of high integrity. Two such methods which are used commercially are the Cosworth process and the improved low-pressure casting (ILP) process.

Cosworth process Conventional methods for casting aluminium and its alloys all involve turbulent transfer which has the undesirable effect of dispersing fine particles, oxide films, and other inclusions through the melt. Such particles and films are known to act as nuclei for the formation of microporosity in solidified castings.

The Cosworth process allows quiescent transfer of metal from the stage of melting of the ingots to the final filling of the mold (Fig. 3.22). The melting/

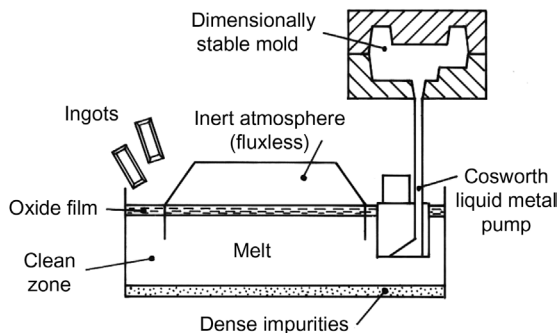


Figure 3.22 Diagrammatic representation of a casting unit using the Cosworth process. From Lavington, MH: *Metals Mater.*, 7, 213, 1986.

holding furnace has sufficient capacity to ensure that the dwell time for the alloy between melting and casting is long enough to allow oxide particles or inclusions to separate by floating or sinking. No fluxes or chemicals are used and a protective atmosphere is employed to minimize the risk of gas absorption and oxide formation. A unique feature of the process is that the mold is located above the furnace and gently filled with molten metal from below by a programmable electromagnetic pump which, itself, has no moving parts. The mold is permeable to allow air to escape from the cavity. The mold is then rolled over under an applied positive pressure immediately after filling. This practice further improves the quality of castings and allows higher production rates to be achieved. Taken overall, the generation and entrapment of oxide particles is minimized. Porosity is much reduced and this is reflected in improved tensile properties and ductility when compared with conventionally produced sand castings (Table 3.3). Fatigue properties at both room and elevated temperatures are also superior and scatter in the test results is much reduced.

Another feature of the Cosworth process is the use of reclaimable zircon sand rather than conventional silica for making molds and cores. This avoids the volume change associated with the phase transition from α to β quartz that occurs in silica at temperatures close to those used for melting aluminium alloys. The stability of both molds and cores is thus much improved which allows close dimensional tolerances of castings to be obtained and repeated. For example, within one mold piece, it is possible to apply tolerances of ± 0.15 mm up to sizes of 100 and ± 0.25 mm up to 800 mm. This feature, when combined with the greatly reduced porosity, allows thin-walled, pressure-tight castings to be produced. These have proved particularly beneficial in the manufacture of complex cylinder heads and other engine components for high-performance motor cars (e.g., Fig. 3.23). Size ranges of castings have so far been within the range 0.5–55 kg and integrated production lines have been established.

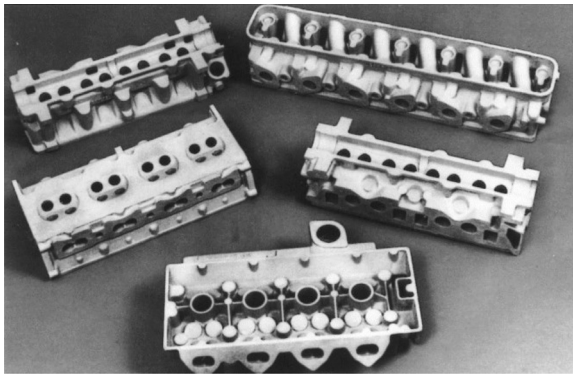


Figure 3.23 Cylinder head castings produced by the Cosworth process. Courtesy Cosworth Research and Development Ltd.

ILP Process Another precision casting method that utilizes the transfer of molten metal vertically up a riser tube into the bottom of the mold cavity is the ILP process developed in Australia. Degassed and filtered metal is delivered to the casting furnace, which uses a pressurized atmosphere of nitrogen to enable quiescent, computer-controlled filling of the mold. Once filled, the mold is sealed and immediately removed so that solidification occurs remote from the casting section. This allows another preassembled mold to be positioned on the station which facilitates high productivity, and cycle times of around 60s become possible.

As shown in Fig. 3.24, a special feature of the ILP process is the use of a combination of thermal (metal) cores and resin-bonded silica sand for the mold which promotes rapid unidirectional solidification in those regions of the casting where optimal properties are required. Furthermore, the mold can be inverted to facilitate this controlled solidification, which may provide dendrite arm spacings less than $20\mu\text{m}$ adjacent to the metal cores, and less than 0.5% microshrinkage overall.

The ILP process has been adapted to allow robotic handling of the molds so that movements are precisely repeatable. The benefits of the Cosworth and ILP processes have been further developed by Nemak into their low-pressure precision sand process for the mass production of cylinder heads and blocks for engines of some models of automobiles that are manufactured for the European and North American markets.

3.6 NEW CASTING PROCESSES

3.6.1 T-Mag

The T-Mag process was developed by CSIRO (the Commonwealth Scientific and Industrial Research Organisation in Australia) to enable permanent mold

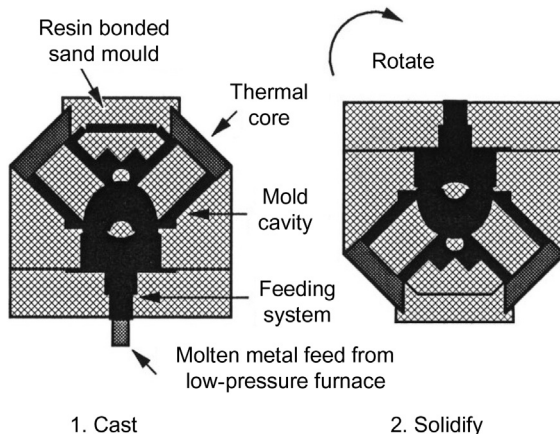


Figure 3.24 Diagrammatic representation of the ILP process. Courtesy Comalco Aluminium Ltd.

casting of magnesium alloys. The process, Fig. 3.25A, is contained within a protective atmosphere and the melt is gently poured to fill the die cavity from the bottom to reduce turbulence and oxide formation. The die cavity and melt transfer tube are rotated during filling so that the transfer tube is at the top of the casting to enhance feeding. Once solidification reaches the gate, the machine rotates back, the casting is ejected when sufficiently strong, and the cycle begins again. As the system is enclosed, cover gas consumption is minimized.

Benefits of the T-Mag process are lighter and stronger components allowing further weight reduction compared to traditional casting processes. An example component is shown in Fig. 3.25B.

3.6.2 Ablation casting

Ablation casting is a relatively new sand casting process where the sand is eroded away by powerful water jets which allows for a rapid increase in cooling rate while solidification occurs. The sand binder is water soluble and the process can be applied to large or small and thick or thin section castings. It has mainly been applied to aluminium alloys but should be suitable for other alloys such as magnesium with good process control. Ablation casting has been tested on magnesium alloy automotive control arm prototypes with some success. Honda applied the process to cast aluminium “large ultra-rigid nodes” for application in the crush zone of a car body. The nodes connect aluminium extrusions which are welded together as part of the spaceframe. This approach achieved the required energy absorption properties where conventional castings would have been too brittle.

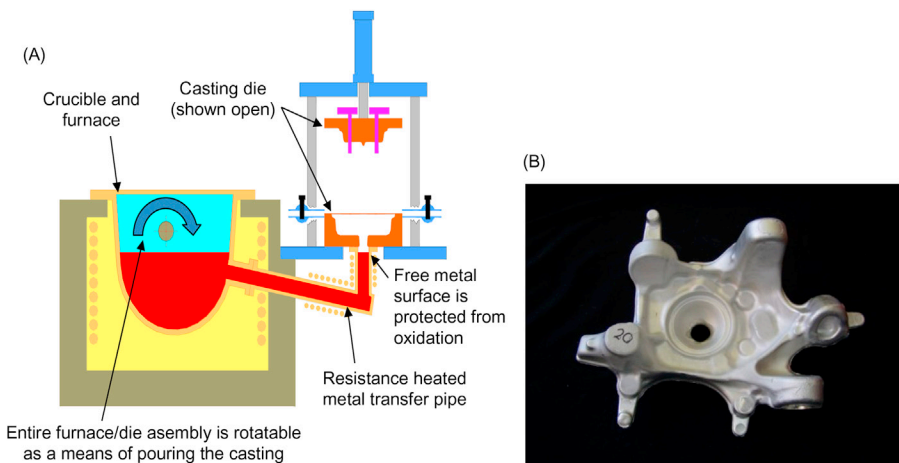


Figure 3.25 (A) Schematic of the T-Mag process and (B) as-cast passenger vehicle front suspension mount made by the T-Mag process. (A) From Wang, L. et al.: *Metal Casting Design and Purchasing*, July/August, p. 30, 2011. (B) Courtesy G. de Looze, CSIRO and S. Groat, T-Mag Pty Ltd.

3.6.3 Application of external fields

The application of external fields to molten melt above the liquidus temperature or during cooling through the nucleation stage of solidification has attracted considerable attention over recent decades as a means to produce a very fine as-cast grain size ($<100\mu\text{m}$) and as a means to produce nanocomposite materials. Several methods have been developed and three examples are presented that represent mechanical, ultrasonic, and magnetic fields.

Melt conditioning Melt conditioning is a process developed at Brunel University, London, where the melt is intensively sheared before entering the casting (Fig. 3.26). High shear is generated by a rotor/stator type mechanism, which comprises a stationary stator and a high-speed rotor capable of rotating at speeds of up to 10,000rpm generating a shear rate of up to 10^5s^{-1} . The process appears to shear and possibly wet oxide films or agglomerates and the resulting finer particles are well dispersed throughout the melt. The particles then act as nucleant particles for the formation of the primary phase of aluminium and magnesium alloys. The additional nucleation produces a highly refined microstructure. Recycled AZ91 magnesium alloy was treated by CSIRO using the melt conditioning device shown in Fig. 3.26 resulting in substantial grain refinement and a significant improvement in strength and elongation. Therefore, the high shear device can be used for physical grain refinement by dispersing naturally occurring oxides; for efficient degassing of aluminium melts from dissolved hydrogen; for the preparation of metal matrix composites; and also for preparation of semisolid slurries.

The melt conditioning technology has been applied to conventional casting processes such as ingot casting, sand casting, low- and high-pressure die

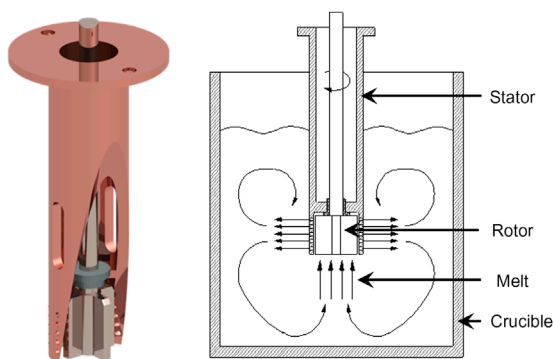


Figure 3.26 The newly developed rotor/stator type high shear device for conditioning of Al and Mg alloys melts by application of intensive melt shearing. Courtesy J. Patel and Z. Fan, BCAST, Brunel University, London.

casting, twin-roll casting, and DC casting, obtaining good refinement without the need for chemical grain refiner addition. Fig. 3.27 shows the dramatic reduction in grain size of a DC cast aluminium alloy billet when melt conditioning is applied.

Ultrasound and magnetic fields Although ultrasonic treatment is not new, there has been significant activity over the last decade in developing approaches to reduce the grain size well below that produced by conventional casting methods, even when grain-refining master alloys are added. As well as stimulating grain refinement, these processes show promise for manufacturing metal matrix nanocomposite materials. Fig. 3.28A is a simulation mesh of an ingot with an ultrasonic sonotrode submerged into the top of the melt. Underneath the sonotrode, a dark region is marked indicating the cavitation region.

Ultrasonic treatment has been studied for many years with limited industrial take-up. Most research has been undertaken on ultrasonic treatment of aluminium and magnesium alloys where ultrasonication above the cavitation threshold has proved to be an effective structural-refining method for metallic alloys. For example, high-intensity ultrasonication can increase the grain density

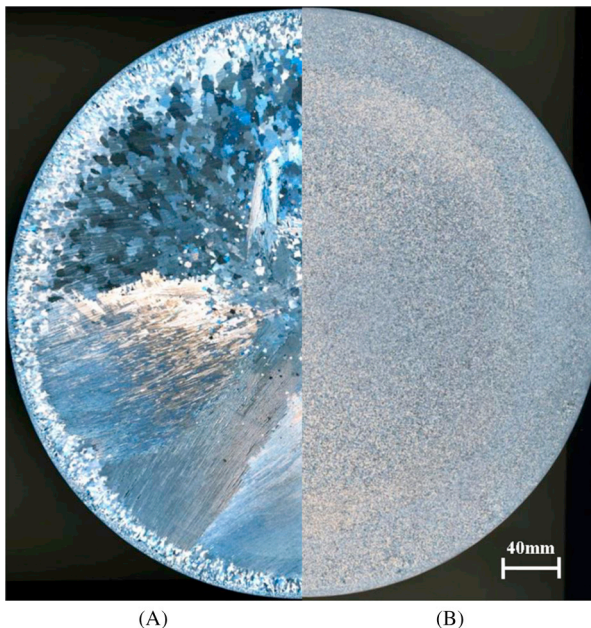


Figure 3.27 Macrostructure of the cross section of an A6063 Al-alloy billet without the addition of chemical grain refiner, 205 mm diameter and 2 m length (A) DC cast and (B) melt conditioned and DC cast. From Patel, JB et al.: *Mater. Sci. Forum*, 794–796, 149, 2014.

enormously from about $15 \text{ grains mm}^{-3}$ (grain size: $\sim 500 \mu\text{m}$) for Al–Mg–Zr alloys to $\sim 3 \times 10^4 \text{ grains mm}^{-3}$ (grain size: $\sim 40 \mu\text{m}$).

Two basic mechanisms of grain formation have been proposed: cavitation-enhanced heterogeneous nucleation and dendrite fragmentation. Direct and indirect evidence produced to date shows that fragmentation occurs when the treatment is applied to the semisolid mixture but it may not make a significant contribution to the density of successful nucleation events thus favoring cavitation-enhanced nucleation.

Fig. 3.28B is a simulation of convection generated by acoustic streaming which is produced by the high-frequency vibrations of the sonotrode. Recent research has highlighted the importance of acoustic streaming. This process enables fast thermal equilibration throughout the melt. The very low-temperature gradient means that nearly all of the melt is undercooled to the same amount at the same rate. The newly formed grains of low solid fraction within the semisolid state thus remain stable and are transported throughout the melt by the vigorous convection resulting in a fine and uniform as-cast grain size.

Similarly, research on the application of pulsed currents to aluminium alloys showed that the nuclei are formed at the top of the casting and, due to thermal

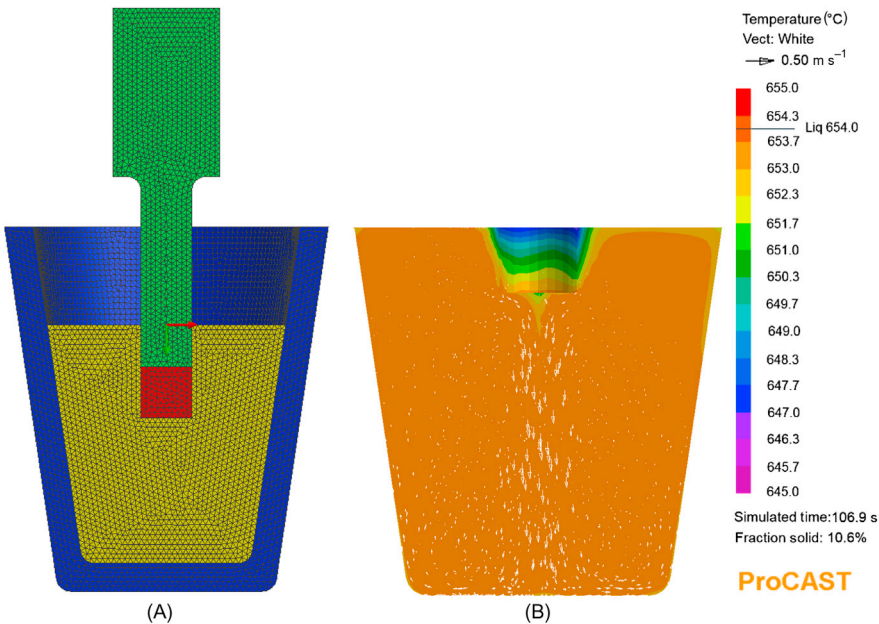


Figure 3.28 ProCAST simulation: (A) a meshed experimental arrangement for ultrasonic treatment of alloys and (B) a simulation of the flow pattern within the melt due to vigorous convection generated by acoustic streaming from the vibrating tip of the sonotrode. Courtesy G. Wang, The University of Queensland.

currents and gravity, the nuclei are distributed throughout the casting. The nuclei are formed when the pulses are applied during nucleation and not during the growth phase. Recent work on pulsed magnetic oscillation (illustrated in Fig. 3.29) of a pure aluminium melt applied from above to just below the melting point was able to generate equiaxed grains throughout most of the cast ingot. In order to be able to do this, the whole melt must be slightly undercooled and the cooling rate must be slow enough to allow time for growth and transport of grains throughout the melt. With good control, this can be achieved due to the action of acoustic streaming causing rapid convection and therefore rapid heat transfer generating a very low-temperature gradient.

A theoretical analysis has indicated that cavitation-enhanced nucleation on existing nucleants is driven by increased undercooling due to the pressure pulse from the collapse of cavities. In addition to affecting the nucleation rate, cavitation may enhance wetting of inoculants, deagglomeration of particle clusters, and mass transport in the melt.

Recently it has been shown that solute plays a major role in controlling the as-cast grain size when ultrasonic treatment is applied and this is illustrated in Fig. 3.30. It was shown that significant ultrasonic grain refinement occurs only in the presence of adequate solute and that the effectiveness of ultrasonic treatment increases with increasing solute content. For instance, under identical ultrasonication and solidification conditions, the resulting grain density in a 70-mm diameter pure Mg (99.98%) ingot reached only about 60 grains mm^{-3} , compared with about 5600 grains mm^{-3} in binary Mg–9Al and about 5700 grains mm^{-3} in commercial AZ91 (Mg–9Al–1Zn). The role of solute has since been verified on Al alloys and Mg–Al alloys.

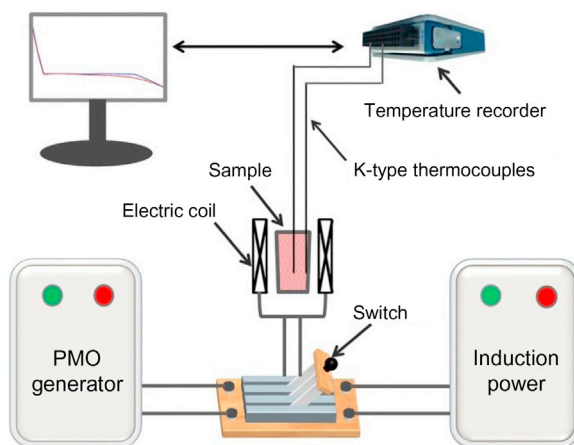


Figure 3.29 Schematic diagram of the pulsed magnetic oscillation (PMO) experimental apparatus. From Liang, D et al.: *Mater. Lett.*, 130, 48, 2014.

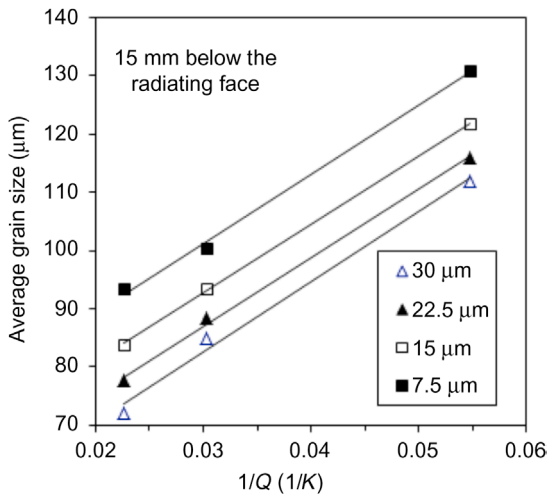


Figure 3.30 Grain size vs $1/Q$ (the inverse of the Q values for the range of Mg–Al alloy compositions corresponding to AZ31, AZ61, and AZ91), subjected to ultrasonic amplitudes between 7 and 30 μm . From Ma, Q and Ramirez, A: *J. Appl. Phys.*, 105, 013538, 2009.

3.6.4 Twin-roll casting of magnesium

Ingots for producing wrought products have mainly been cast in permanent metal molds. However, due to a demand for magnesium alloy sheet and extrusions, more attention is being given to semicontinuous DC methods which are broadly similar to those described for aluminium alloys (Section 4.1.1). Following extensive pilot scale trials, a vertical DC caster for magnesium alloys has been constructed in Austria that is capable of producing extrusion billets with a maximum weight of 4000 kg. Smaller, horizontal DC casters have also been developed.

For the conventional production of magnesium alloy sheet, cast ingots are usually produced with dimensions up to 0.3 m \times 1 m \times 2 m. These ingots are homogenized (e.g., at 480°C) for several hours, scalped, and hot rolled in stages in a reversing hot mill to a thickness of 5–6 mm. The sheet is then annealed (e.g., at 340°C) before each subsequent rolling pass which reduces the thickness by only between 5% and 20% because of the hexagonal crystal structure of the alloys. This latter part of the rolling process is costly and time consuming so that the current usage of magnesium alloy sheet worldwide is limited. This makes the prospect of strip casting an attractive alternative if sheet can be cast to near final thickness and given a final finish roll. Casting of magnesium sheet does, however, present some additional problems apart from the tendency of the molten metal to oxidize with the danger of catching fire. These problems arise because magnesium alloys freeze faster than aluminium alloys and generally have longer freezing ranges which makes as-cast sheet more prone to

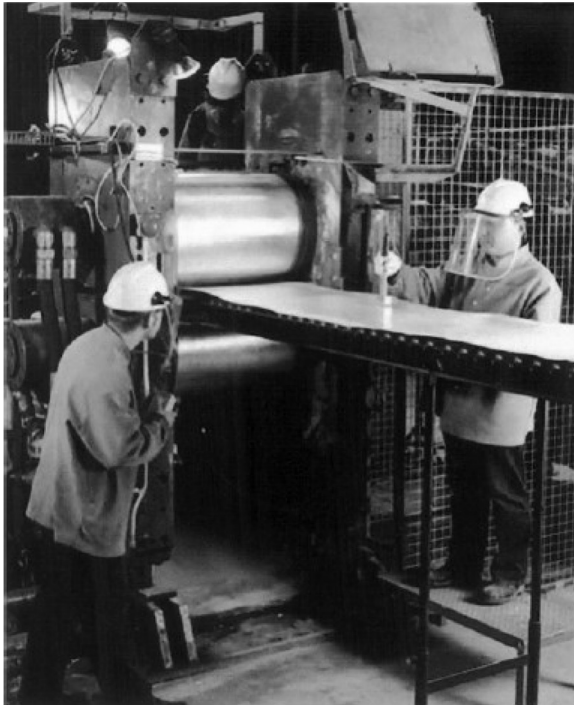


Figure 3.31 Pilot plant for twin-roll casting of magnesium alloys. Courtesy D. Liang, CSIRO Manufacturing, Melbourne.

surface defects and internal segregation. Nevertheless, some success has been achieved through the adaption of twin-roll casting practices developed for strip casting aluminium alloys. As an example, [Fig. 3.31](#) shows a pilot plant that can directly cast a range of magnesium alloys as sheet up to 600 mm wide which has a thickness of 2.5 mm that can be further reduced down to 0.5 mm using a conventional finishing mill.

3.7 JOINING

Most aluminium casting alloys can be arc welded in a protective atmosphere of an inert gas, e.g., argon, provided they are given the correct edge preparation. Ratings of weldability are included in [Table 5.3](#). In addition, some surface defects and service failures in sand and permanent mold castings may be repaired by welding. Filler metals are selected which are appropriate to the compositions of alloy castings with 4043 (Al–5Si) and 5356 (Al–5Mg–0.1Mn–0.1Cr) being commonly used. Special care must be taken when repair welding cast components such as automotive wheels that have previously been heat treated. Unless an appropriate reheat treatment schedule is available, this

process should be avoided. With respect to joining of castings by brazing, similar conditions apply to those discussed for wrought aluminium alloys in Section 4.5.2.

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