Best practices for making high integrity lightweight metal castings – molten metal composition and cleanliness control

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Abstract: To make high integrity lightweight metal castings, best practices are required in various stages of casting and heat treatment processes, including liquid metal composition and quality control, casting and gating/riser system design, and process optimization. This paper presents best practices for liquid metal processing and quality assurance of molten metal in both melting and mold filling. Best practices for other aspects of lightweight metal casting will be published separately.

Key words: best practices; high integrity casting; lightweight; metal casting; molten metal cleanliness; alloying; trace element

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With the increasing demands for reducing automobile weight for improved fuel efficiency, the application of lightweight metal castings continues to increase in automotive components including engine blocks, cylinder heads, intake manifolds, brackets, housings, chassis, transmission parts, and suspension systems. As many of these applications are important structural components, mechanical properties and particularly fatigue performance of the castings are critical to their success [1].

Mechanical properties of lightweight metal castings strongly depend upon the size, amount, and distribution of defects and multi-scale microstructures. In aluminum castings, volume fraction of defects dominates tensile behavior [2-3], while in dynamic loading it is the defect size (pore and oxide film) that controls the fatigue performance [4-22]. Reducing defect size improves fatigue properties. When the pores and oxide films are smaller than a critical size, the cracked/debonded eutectic particles and persistent slip bands from the aluminum matrix become the fatigue crack initiation sites with a significant increase of fatigue life. As shown in Fig. 1, the fatigue life of the samples failed by porosity is significantly lower than those by slip bands [1, 13].

Similar conclusions can be drawn in magnesium castings [23-25]. Casting defects also dominate the tensile and fatigue properties. In the presence of defects, fatigue cracks always initiate from the defects, leading to a low fatigue life. In the absence of defects, fatigue cracks can initiate from eutectic particles, slip band, or twin bands. Figure 2 shows an example of fatigue cracks initiated from the twin bands in the NZ30K1-T4 Mg alloys [24].

To improve strength and particularly fatigue performance of lightweight metal castings, therefore, casting defects should be minimized and eutectic particles should be refined and uniformly distributed in the microstructure. In aluminum and magnesium castings, the size and population of multi-scale defects and microstructural constituents depend upon not only the melt quality, mold filling and solidification conditions, but
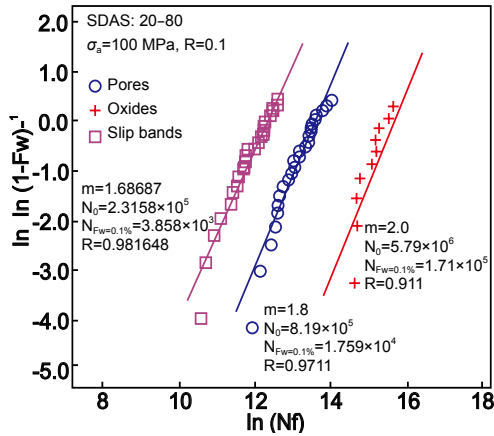
Fig. 1: Two-parameter Weibull plot for fatigue life of a Sr-modified A356 casting alloy sorted by type of crack origin (pore, oxides, or slip bands) observed on fracture [1, 13]
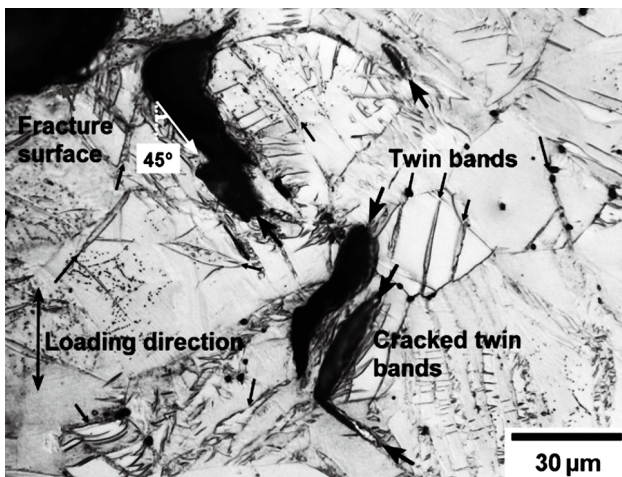

Fig. 2: SEM image showing crack initiation from twin bands in NZ30K1-T4 Mg alloy [24]

also the alloy composition, gating/riser system design and heat treatment process. The heat treatment process can also affect residual stress magnitude and distributions in the castings. This paper presents best practices for alloying and trace element control as well as molten metal contamination avoidance and elimination in both melting and mold filling. Best practices for other aspects such as quality assurance during casting and residual stress and distortion control during heat treatment will be published separately in this journal.

1 Alloying and trace element control

For a given casting process and a design of casting geometry and gating/riser system, alloy composition including trace elements such as iron (Fe), strontium (Sr), phosphorus (P), bismuth (Bi), calcium (Ca), and others, can play an important role in solidification characteristics and formation of defects and multi-scale microstructure.

Shrinkage tendency of an alloy is highly related to freezing range (also called solidification range) and solidification

sequence which is determined mainly by key alloying elements such as silicon in Al-Si based alloy system. There are a number of alloys widely used in lightweight metal casting. If high strength is required, heat-treatable alloys must be used. The alloy options can be further narrowed when considering other service requirements, such as pressure tightness, corrosion resistance and machinability. When the casting method is determined, the alloy choice is limited because not all alloys can be used with all casting methods. Sometimes, the alloy that shows the best properties on paper may have production characteristics that make it less desirable on an overall basis than other eligible alloys. Therefore, it is necessary to consider alloy overall manufacturability (i.e. castability, heat treatability, machinability, etc.) in addition to the mechanical properties the alloy can offer. For instance, aluminum alloys considered to be premium-strength compositions are listed in specification AMS-A-21180, which is extensively used in the United States for premium casting procurement. They include A210.0, A206.0, 224.0, 249.0, 354.0, C355.0, A356.0, A357.0, D357.0, 358.0, and 359.0. Among these aluminum alloys, the 200 series of alloys have high tendency of hot-tearing and wider solidification freezing range and are thus difficult to cast.

Grain refinement and microstructure modification has long been recognized and widely utilized in lightweight metal castings for improving castability and mechanical properties, particularly ductility. Beneficial effects on solidification, feeding, and properties are well established. In cast aluminum alloys, master alloys of aluminum with 5% titanium and 1% boron are commonly used since TiB_2 and $TiAl_3$ together are more effective grain refiners than $TiAl_3$ alone [3]. To avoid formation of Ti containing sludge, the titanium concentration in the alloy should be controlled no more than 0.2% and preferably below 0.18% for most cast aluminum alloys.

It is well known that modification of the acicular silicon present in cast Al-Si based alloys to a fine fibrous form results in an improved strength and, particularly, improved ductility for these alloys. Modification technology has matured since Pasz first introduced sodium (Na) to modify eutectic silicon in 1920. Additions made for modification and refinement of silicon structure now include strontium (Sr) and antimony (Sb). Several reports have appeared in literature concerning negative interactions between antimony and either strontium or sodium. It was reported that Sb concentrations as low as 100 ppm can significantly affect the tensile elongation of Sr modified A356 alloy [26]. This deterioration in mechanical properties was exacerbated by the presence of phosphorus (P). To achieve the similar eutectic silicon morphology in the presence of varying levels of P, sufficient Sr needs to be added. Figure 3 quantitatively shows that higher Sr concentrations are required for retaining good modification when P neutralization of the Sr effects is considered [27].

Phosphorus is an impurity associated with silicon used in the aluminum-silicon based alloys. The effect of P, at or beyond concentrations of a few of ppm, is not only to perform the
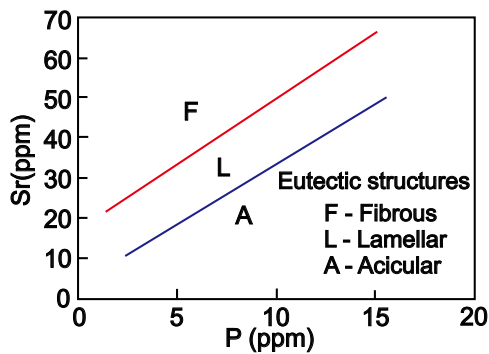
Fig. 3: Sr and P interaction in Al-7%Si alloy when solidification time is 60 s [27]

function of nucleating primary Si in eutectic or hyper-eutectic but also to yield a distinctly acicular eutectic silicon structure in Al-Si hypoeutectic alloys. It was also found that both the number of primary α-dendrites and the dendrite arm spacing (DAS) was increased in the high-purity Al-10%Si alloy by the addition of 0.005% (50 ppm) of phosphorus [28].

Magnesium also tends to coarsen the eutectic silicon structure and thus reinforces the effect of P. For example, an Al-7%Si alloy containing 2 ppm P still exhibits a lamellar silicon structure, whereas Al-7%Si-0.3%Mg alloy also containing 2 ppm P is acicular [27].

Like phosphorus, bismuth also neutralizes the effect of Sr modification. To retain full modification, Sr/Bi mass ratio higher than 0.45 is required when bismuth is present in the melt [29-31].

In spite of the positive effect of Sr modification on tensile strength and ductility in particular, excessive modification increases the tendency of microporosity and segregation of intermetallic phases formed in late solidification due to the change of solidification characteristics and formation of dual primary and eutectic grain structures [32]. It has been also reported that excessive eutectic modification delays the formation of an impermeable casting skin and thus increases core gas penetration from the sand cores, resulting in gas bubbles in the solidified castings. (These exogenous gas bubbles are distinct from gas porosity resulting from the rejection of hydrogen dissolved in the liquid aluminum during solidification). Therefore, it is important in aluminum casting to properly control the eutectic modification level to minimize both macro shrinkage porosity and gas bubbles simultaneously.

Iron is another important impurity in cast aluminum alloys. Its detrimental effect on casting quality and properties has long been recognized and well understood [3, 33]. Given below are practical guidelines summarized by Taylor for iron level control in Al-Si based casting alloys [33]:

(1) Wherever possible, iron levels in Al-Si based alloys should be kept as low as practical. This means minimizing iron contamination through careful selection of raw materials (i.e. ingots, silicon, etc.) and the maintenance of good refractory coatings on all steel tools used to prepare and handle melts.

(2) Iron levels above the critical level for the silicon content

of the alloy should be avoided as these can cause serious loss of ductility in the final cast product and decrease casting productivity through increased rejects due to shrinkage porosity, and particularly "leakers". The critical iron content (in wt.%) for an alloy can be calculated using $Fe_{crit} \approx 0.075 \times [\%Si] - 0.05$. If solidification (cooling) rates are very high (e.g. high pressure die casting), the Fe levels above critical contents may not be detrimental, but as the cooling rate decreases (gravity die casting, sand casting, etc.) the probability of super-critical iron levels causing problems dramatically increases.

(3) Traditional heat treatment regimes for Al-Si alloys, e.g. T6, do not alter the nature of the offending Fe-containing phases. Although the overall performance of an alloy may be improved by heat treatment even with the remnant as-cast intermetallic phases, it would be better still with low iron levels initially.

(4) Adding Mn to neutralize the effects of iron is common, at Mn:Fe ratios of ~ 0.5, however, the benefits of this treatment are not always apparent. Excess Mn may reduce β-phase and promote α-phase formation, and this may improve ductility but it can lead to hard spots and difficulties in machining. Mn additions do not always improve castability nor reduce porosity in high Fe alloys. Its effect is sensitive to alloy composition. The addition of Mn to melts with high iron levels can also promote the formation of sludge, if the sludge factor (derived by [%Fe] + 2[%Mn] + 3[%Cr]) exceeds a particular value for a given alloy and melt holding temperature. This is a serious problem particularly for die-casters who usually use low melt temperatures and high impurity secondary alloys.

The control of alloying and trace elements in cast magnesium alloys is much less sophisticated in comparison with cast aluminum alloys. This is partly because cast magnesium alloys are relatively new and the understanding of their metallurgical aspects is still limited. Nevertheless, the solidification of existing magnesium alloys differs in a number of essential ways from the solidification of aluminum foundry alloys, as summarized by StJohn et al. [34]. Magnesium alloys have a much lower volume fraction of eutectics and the composition of most magnesium alloys is within the range of the maximum solid solubility of the major alloying elements such as aluminum and zinc. This means that commercial magnesium alloy microstructures usually have divorced eutectics that form due to non-equilibrium solidification. This can be a problem for structural alloys as it decreases ductility, but the high solidification rates seen in high pressure die casting (HPDC) or heat treatment can improve ductility. On the other hand, divorced eutectic formation is likely to be an advantage for creep resistant alloys as the stable intermetallic phase present in the grain boundaries increases the microstructure's resistance to deformation at elevated temperatures. Grain refinement of cast magnesium alloys can also improve the distribution of the intermetallic phases. A good example is the non-aluminum containing magnesium alloys grain-refined by adding zirconium. Zirconium provides
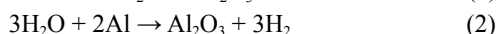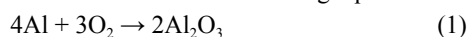
very potent nucleant particles and high segregating solute, both of which ensure very good grain refinement. Unfortunately, a similarly effective grain refiner for aluminum containing magnesium alloys has not been developed.

Another difference compared to cast aluminum alloys is the large freezing (solidification) range in most cast magnesium alloys, particularly those available for high pressure die casting. This reduces the alloy castability, increases shrinkage porosity, and promotes the formation of banded defects in HPDC. These defects reduce the ductility of the cast components but the large solidification range is responsible for the ability of magnesium alloys to fill long thin sections.

2 Liquid metal quality assurance

The control of liquid metal quality is absolutely vital for the production of high-quality castings since the majority of defects in final casting are usually related to inclusions and gases coming from the liquid metal [35-37]. The liquid metal should be cleaned to the highest level possible before it is introduced into the mold cavity. The highest level of quality of liquid metal means that the oxide inclusions and the dissolved gases are minimized to the point that they will not cause casting defects during solidification. To ensure the highest possible quality of liquid metal, the starting point is the cleanliness of the furnace charge that could comprise a mixture of primary ingot, secondary (recycled) ingot, bought-in scrap, and in-house returns from the gating system and scrap parts. In general, recycled ingots and particularly scrap returns tend to have higher impurity contents and higher level of oxides and dissolved gases in comparison with primary ingots, especially those supplied from continuously cast billet. When a high quantity of scrap return is used, a special caution should be taken to avoid liquid metal contamination because all scrap return surfaces are full of oxides and may also contain moisture and other contaminations. One technology has been recently developed to reduce liquid metal contamination with scrap return charge [38]. The method comprises preheating a scrap charge to remove moisture and contaminants. The scrap charge is then coated on all free surfaces with a layer of flux. Subsequently the scrap charge is melted in a furnace to form a melt bath of liquid aluminum suitable for casting. The flux layer removes the naturally occurring oxide film from the scrap charge surfaces as well as provides a cover flux to protect the melt bath from oxidation.

During melting, more oxides and dissolved gases can be generated when the liquid metal surface is not properly protected and directly contacts the atmosphere. For example, liquid aluminum can form aluminum oxides very rapidly when it is exposed to the atmosphere. Meanwhile, hydrogen gas will be dissolved into the liquid aluminum since atmosphere also contains moisture. This is shown in the following equations:

$$4Al + 3O_2 \rightarrow 2Al_2O_3 \tag{1}$$
$$3H_2O + 2Al \rightarrow Al_2O_3 + 3H_2 \tag{2}$$

$$H_2 \rightarrow 2[\underline{H}]_{melt} \tag{3}$$

Although aluminum oxides (Al_2O_3) are generally considered protective, they are unable to maintain their continuous protection particularly at elevated temperatures when any of the IA or IIA oxide groups (except BeO) are present. In addition, more oxides and spinel form when Mg > 0.005% and melt temperature > 745 °C. When the aluminum alloy contains more than 0.5% magnesium, in particular over 2%, its oxidation process does not follow a typical parabolic law for pure aluminum; and in contrast, MgO and/or spinel ($MgAl_2O_4$) form almost exclusively, in addition to aluminum oxides (Al_2O_3), Fig. 4. A significant amount of commercially viable casting and wrought aluminum alloys (e.g. A356, 319, A380, A390, A206, A6061, A6101, 530, etc.) contain magnesium. During the melting of these alloys, magnesium oxides and spinel continue to form regardless of whether or not surface film rupture occurs. Figure 4 shows an example of the weight increase with time, due to oxidation, of Al-7%Si liquid containing different Mg and Sr contents at 730 °C. As can be seen, Mg and Sr additions significantly increase the oxidation rate of the liquid Al-7Si alloy. This also explains why strontium modification effectiveness fades during melting or re-melting.
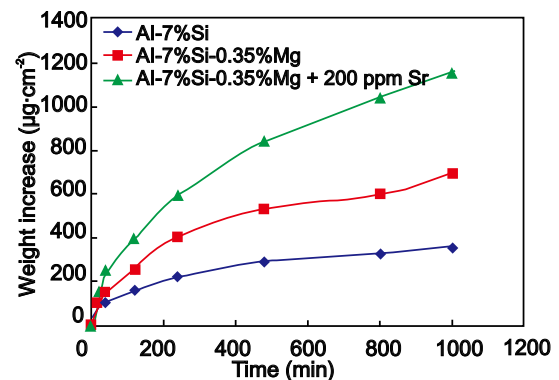

Fig. 4: Thermogravimetric analysis of oxidation rate of aluminum alloy (Al-7%Si) with or without Mg and Sr addition at 730 ºC

The dissolved hydrogen in aluminum liquid can produce a lot of gas porosity if the melt is not properly degassed. This is because of the large difference in hydrogen solubility between liquid and solid (Fig. 5). To make high integrity aluminum castings, the hydrogen level in the liquid aluminum should be controlled below 0.15 mL per 100 g aluminum and preferably below 0.1 mL per 100 g aluminum.

As the dissolved gas in liquid metal is so vital for the production of high-quality castings, degassing has become an unavoidable step to perform in practice since un-degassed melts usually have a hydrogen level well above 0.15 mL per 100 g aluminum.

The principles of sparging for the removal of dissolved hydrogen had been developed in the late of 1920s and 1930s. The use of active gases such as chlorine and physicochemical separation of entrained oxides and other nonmetallics by fluxing
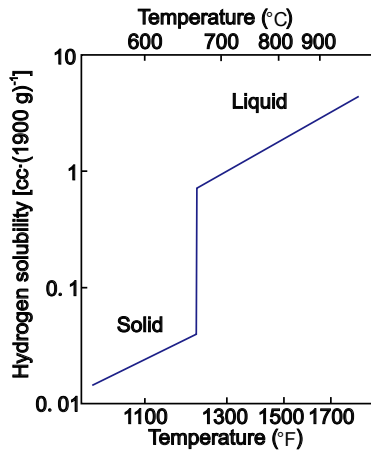
Fig. 5: Hydrogen solubility in pure aluminum [40]

became known in the 1930s. The use of diffusers for more efficient gas fluxing was developed later. By 1950, particulate filtration and countercurrent fluxing using nitrogen, argon, chlorine, and combinations of these gases became common for wrought alloy production, and variations of these processes were being used in gravity casting foundries [39]. The later development of rotary degassing systems was quickly adapted to foundry use. In the past several decades, numerous companies have developed a variety of degassing equipment. Sigworth has recently reviewed the most important existing techniques and also summarized the present scientific understanding in this area [40]. The methods and equipment available now to degas metal appear to be well established and reliable. Moreover, competition amongst different suppliers, better materials, and improved designs have resulted in lower operating costs.

Among various degassing techniques, rotary degassing is considered as one of the most efficient methods. In rotary degassing systems, degassing is accomplished by 'purging' liquid metal with an inert gas. The secret to good degassing performance is to produce small bubbles (Fig. 6), which is strongly related to the design of the rotary impeller heads. Some simple impeller heads like the notched rotary head developed by RFP technologies [41], or the rotating square block used by Metallics [42], are inexpensive to machine, but do not produce

very small bubbles. For the best degassing performance and high process efficiency, however, bubbles smaller than 2–3 mm in diameter are needed. For this reason, several high efficiency 'pumping' rotor heads have been developed. One example of this approach is a head design developed by Foseco [43].

In practice, operators should pay attention to the bubble size and bubble distribution on the surface of melts during degassing. An optimal operation schedule should be established for the facility in terms of purging gas flow rate and head rotation speed. When gas flow rate is too low, the number of bubbles is low although bubbles are relatively small. When the gas flow rate becomes too high, large bubbles will appear on the surface and result in volatile turbulence. In this case, the rotor has been 'flooded' with more gas than it can handle. But, if the rotor speed is increased in this case, the large bubbles may disappear. This is because faster rotor speeds can handle somewhat larger gas flow rates. The desired result is a good dispersion of small bubbles, while maintaining a relatively quiet surface. Small bubbles appear to 'dance' just underneath the metal surface [40]. The bubbles should be observed on the entire surface of the metal in the degassing area, not just in the center or at the edges. When trying to achieve evenly distributed small bubbles, it should be ensured that the head rotation speeds do not exceed the level at which the vortex forms. A vortex can 'pump' oxides from the surface into the liquid metal, severely degrading melt quality.

It is also a good idea to keep metal temperatures as low as possible during degassing. The gas solubility is usually an exponential function of temperature, as shown in Fig. 5. For instance, each 100 °F (56 °C) increase in liquid aluminum temperature doubles the time needed for hydrogen degassing. This has been well demonstrated by Sigworth (Fig. 7) [40]. Higher temperatures also result in more rapid oxidation and greater loss of metal to dross. Degassing can also be an effective way to remove inclusions and oxides suspended in the liquid metal, especially when chlorine or fluxes are employed.

Selection of the degassing location is also important as it significantly affects degassing efficiency. Degassing is usually accomplished at one of the following locations in a metal casting facility [40]:
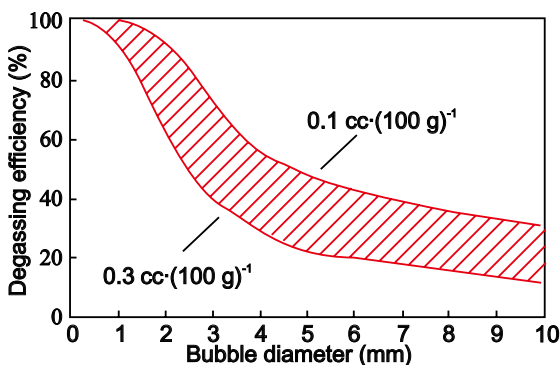

Fig. 6: Calculated degassing efficiency as a function of bubble size [40]
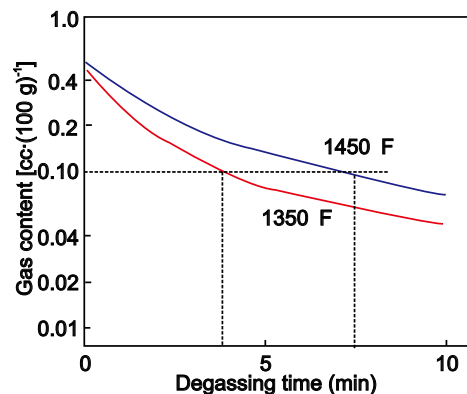

Fig. 7: Gas removal in A357 alloy at two temperatures [40]

(1) In a transfer ladle, Fig. 8, used to convey metal between the melting and holding furnaces,

(2) In an in-line system, when the metal is conveyed to pouring stations through launder system, Fig. 8, or dip well, and

(3) In crucible furnaces, usually just before casting the metal.

In aluminum alloys, both metal quality and hydrogen levels can be measured using a number of different techniques, as listed in Tables 1 and 2 [45]. Molten aluminum quality can be measured directly or destructively from the solidified component. Direct measurement tends to be more expensive, but it is fundamentally more accurate.
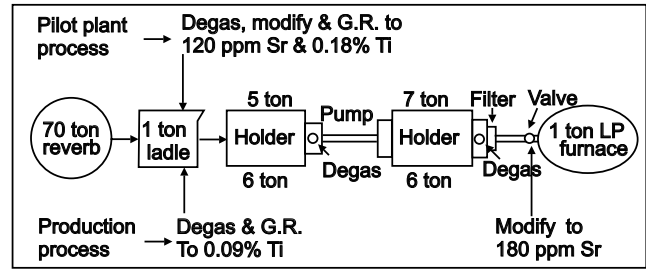
Fig. 8: Degassing locations used in both pilot plant and production plant at Nemak [44]

Table 1: Summary of metal quality tests for aluminum alloys [45]

Solid samples		Liquid samples	
Name	Technique	Name	Technique
Acoustic	"Wheel tapping" RFDA	PoDFA	Filtration (pressure)
Metallographic	Small laboratory samples	Prefil footprint	Filtration (pressure)
Extraction	Chemical/electrolytic	LAIS/VFT	Filtration (vacuum)
Fast neutron	Oxygen determination	Density separation	Molten metal centrifuge
Fracture bar	Mechanical test	LiMCA Coulter method	Electrical
Tool wear	Historical	Ultrasonic	Reflection of ultrasonic

Table 2: Summary of hydrogen measurement tests for aluminum [45]

Reduced pressure tests		Fundamental tests	
Name	Technique	Name	Technique
Straube-Pfeiffer	Fixed pressure	Ransley probe	Recirculating carrier gas
First bubble	Variable pressure	Telegas	Development of above
Vibrated vacuum	Encourage gas bubbles	AISCAN	Similar to above
Constant volume	Minimize shrinkage effects	CHAPEL	Direct partial pressure
Density index	Magnify porosity	NOTORP	Solid-state galvanic cell
Hyscan QRP	Hydrogen from sample	Vacuum solid extraction	Pressure rise
		Nitrogen carrier fusion	Similar to Telegas
		LECO	Remelting test for chilled samples

Among these measurement techniques, the reduced pressure test (RPT) is widely used by most metal casters. This is not only because the RPT method is relatively simple and easy to use but because it can measure both gas and oxide inclusions. To get accurate RPT measurements, however, a proper procedure must be followed. Given below is the list of the procedures initially summarized by Groteke [46] and then revisited by Sigworth [40]:

- Use a thin-walled, coated cup for sample collection
- Preheat cup in an adjacent surface area
- Skim oxides from the melt surface
- Back-fill sample cup to minimize collection of surface oxides
- Collect 125–150 g sample from the furnace (collect sample below the surface)
- Transport sample rapidly to a vacuum unit with good seals and pump capacity

- Place on an insulated pedestal
- Cover with a chamber that permits visual observation (if possible) of the solidifying sample
- Start the pump and minimize vibration of the sample during solidification
- Control vacuum to target level adjusted to an absolute vacuum reference
- Observe the sample while solidifying to develop an estimate of metal cleanliness (and gas content at low absolute pressures)
- Adjust the time of the cycle to allow for full solidification of the sample
- Upon cycle completion, remove sample and air cool both cup and sample
- Measure specific gravity of the sample and compare to process specifications (or count bubbles forming at the surface)

3 Bifilm avoidance during metal transfer

After molten metal is cleaned in furnace, getting the clean metal into the casting cavity without generating extra oxides and particularly bifilms becomes important in achieving high integrity metal casting. Any young (new) oxides and bifilms formed during metal transfer and mold filling will probably end up in the final casting and pose significant damage to casting properties. Bifilms are usually dry on the folded surfaces and easily opened up under the reduced pressure due to shrinkage during solidification. The young oxides and bifilms form in a relatively short time (during metal transfer and mold filling), the thickness of the films is usually thin ranging from a few tens of atoms to several micrometers so that they can be easily deformed [45]. Figure 9 illustrates a thin young oxide film draped over the dendrite tips in a high pressure die casting of 380 alloy, indicating the entrained oxide bifilms initiated shrinkage porosity. Figure 10 shows a young magnesium oxide film initiated fatigue crack in a NZ30K1 Mg alloy.
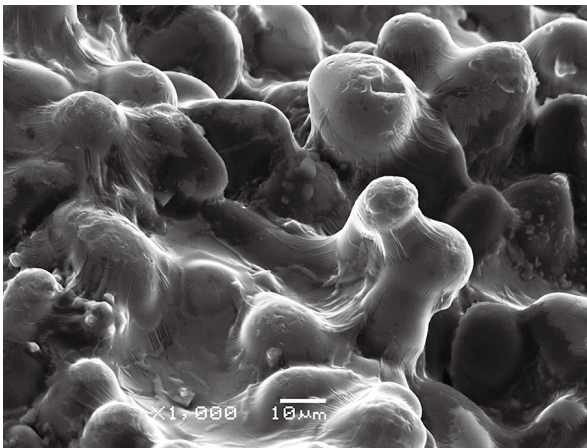

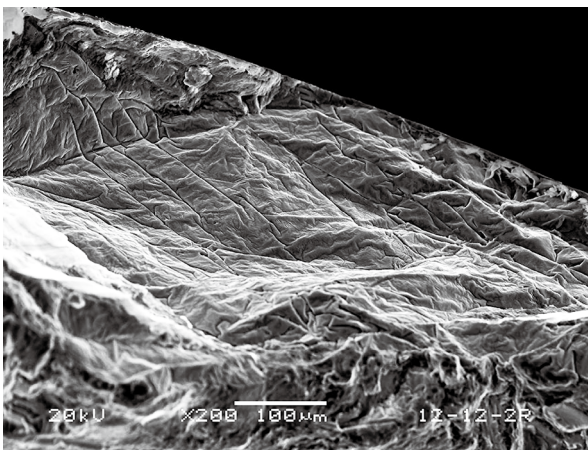**Fig. 9: An SEM picture of aluminum oxide film draped over dendrite tips in a 380 alloy**

Fig. 10: An SEM picture of magnesium oxide film initiated fatigue crack in a NZ30K1 Mg alloy

To understand the bifilm entrainment, Campbell [47] first put forward a concept of critical velocity. For every liquid metal there exists a critical velocity above which the surface will fold over and entrain itself in the bulk of the liquid metal. The critical velocity, V_{crit}, is approximately given by

$$V_{crit} \approx 2 \sqrt[4]{\frac{rg}{\rho}} \qquad (4)$$

where g is the acceleration due to gravity in m·s^{-2}, γ is the surface tension in Nm, and ρ is the liquid density in kg·m^{-3}.

The critical velocity is directly proportional to the fourth power of the ratio of surface tension to density. As summarized by Jolly [45], there is little change of critical velocity throughout the periodic table. V_{crit} falls between 0.25 m·s^{-1} (Se) and 0.60 m·s^{-1} (Be) for the metallic elements in the periodic table and it falls between 0.37 m·s^{-1} (Zn) and 0.50 m·s^{-1} (Ti, Al) for the most common engineering materials (i.e., steel, Al, Cu, Mg, Zn, and Ni alloys).

To further demonstrate the ease of bifilm entrainment in real casting, Campbell also calculated a critical metal falling distance. The critical distance, H_{crit}, for a falling stream to reach the critical velocity, V_{crit}, can be calculated by [46]

$$H_{crit} \approx \frac{v_{crit}^2}{2g} \qquad (5)$$

For most cast aluminum and magnesium alloys, the critical velocity is about 0.5 m·s^{-1} and the acceleration due to gravity is 9.81 m·s^{-2}, then the critical distance, H_{crit}, that the metal has to fall to reach the critical velocity is only 12.7 mm. This means that if liquid aluminum or magnesium falls a distance greater than 12.7 mm, then there will be surface turbulence (damage) and the probability of surface oxide generation and entrapment of the surface oxides into the bulk of the liquid metal. In real casting process, liquid metal often falls a distance much greater than this critical distance, either during transfer of liquid metal from one furnace to another, during the pouring of the casting itself, or even during the mold cavity filling stage.

To avoid bifilm entrainment, we should stop pouring, as initially proposed by Campbell [48], and also eliminate any possible waterfalls when the melt fills the mold cavity. Several developments in mold filling in the past decades have provided options to reduce bifilm entrapment.

Low pressure casting (LPC) process was first in production use in the US in the 1950s. A number of challenges that were unaddressed by commercial low pressure systems were successfully met. A cam-controlled back-pressure method based on gross casting weight was used to retain residual metal levels at the top of the feed tube. This feature prevented the inclusion spawning characteristic of normal low pressure cycles [39]. In-gate filtration and screen methods were also devised. The range of part designs and alloys that were cast would be considered unusual today when the low pressure process has become principally known for automotive wheel production. Instead, the low pressure method is considered a means of non-turbulent mold filling with a number of

additional advantages that include reduced gross/net weight (i.e. improved metal yield) and lower metal pouring temperature. Some examples are diesel engine and compressor pistons, air-conditioner compressor bodies, beatings, high-speed rotors and impellers, etc. Geometric symmetry, which is normally a criterion for low-pressure production was not considered a prerequisite, and many of the castings that were produced used conventional risering rather than exclusively relying on the in-feed for shrinkage compensation.

Although the low pressure casting process achieves non-turbulent mold filling, it does not mean that the low pressure filled castings are free of oxides. Significant amounts of oxides can be generated by the massive turbulence during the filling of the furnace, and can be formed in the feed tube as the melt washes up and down. Furthermore, following each casting, the melt falls down the feed tube each time when the pressure is cut off, the 'whoosh' effect in the bottom of the furnace stirring the oxide sediment and other inclusions back into suspension, ready to be introduced to the next casting. The LPC process involving the use of separated furnace bodies is clearly an improvement, but still does not address all of the fundamental limitations of this process [48].

To overcome some limitations associated with low pressure casting process, vacuum riserless casting (VRC) was developed in the early 1960s. Rather than pressurizing a contained molten reservoir, the application of a vacuum on the mold cavity drew metal from the bath through a short fill tube. The metal source was exposed for periodic treatment. The distance from subsurface metal entry to the casting cavity was minimal, dies were extensively chilled, and the process could be highly automated. While only relatively small and simple shapes were produced by the VRC method, productivity and mechanical properties were exceptional. Like the low pressure casting (LPC) process, however, the VRC process still has the problem of the 'whoosh' effect following each casting when the vacuum is cut off.

Counter-gravity mold fill methods were developed involving the use of mechanical or induction pumps. One of the excellent examples is the Cosworth casting process which was invented by Prof. John Campbell in 1970s [49]. As shown in Fig. 11, the Cosworth process uses the large holding furnace working at
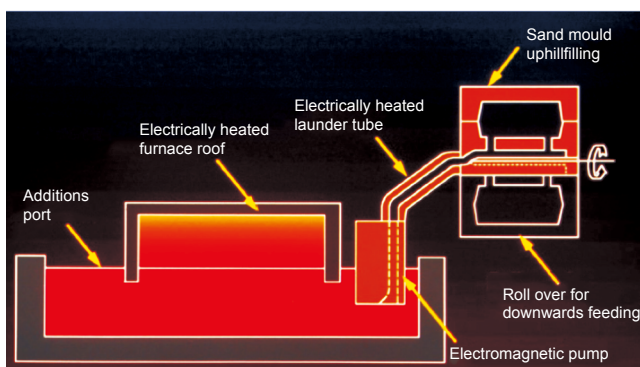

Fig. 11: Cosworth counter-gravity casting process [48]

an essentially constant level, plus retaining the melt at the elevated level close to the delivery point of the riser tube that effectively eliminates the issues associated with LPC and VRC processes.

Similarly, the level pour casting process also offers non-turbulent metal flow. It essentially grows a cast component with minimal oxides and also provides unidirectional solidification with a short feeding distance which is highly desirable for the production of quality cast components. The level pour process had its origins in the direct chill process that was developed for fabricating ingots. It was natural that the shared concerns for metal distribution and solidification principles would result in the synthesis of process concepts [39].

There are two types of level pour casting systems based on mold assembly and mold filling configurations. In one form, the liquid metal is introduced to the bottom of a mold through a moving pouring cup that traverses the length of the mold. Quiescent flow and continuous layering of molten metal provides improved internal quality. Excellent soundness can be obtained without the use of the extensive risering normally required for hot spots and those long solidification range alloys. In the other form, the assembled mold is lowered on a hydraulic or a mechanical platform through a trough arrangement that provides non-turbulent flow of metal through entry points that is usually parallel to the vertical traverse of the mold. In this case, the metal flow rate is controlled by the dimensions of the entries and the lowering rate of the mold, which can be modulated for mold cross-sectional variations as a function of mold travel. Characteristic of the direct chill process on which it is based, the level pour process features quiescent molten metal flow, minimized feeding distance, and reduced pouring temperatures.

4 Conclusions

Mechanical properties and, in particular, fatigue performance of metal castings are dominated by casting defects and, to a much lesser degree, by multi-scale microstructure. Therefore, the casting defects should be eliminated in the high integrity metal casting (or, at least, the defect size should be reduced to a level which is smaller than a critical size that will affect mechanical properties).

(1) A proper selection and control of alloy composition and particularly trace element contents is the first step to make high integrity metal castings since it is the alloy compositions that dictate the alloy thermophysical properties and solidification characteristics, which control the formation of defects and multi-scale microstructure. Whenever possible, one should optimize the alloy composition to achieve the best castability (minimal solidification range, low shrinkage tendency, and high feeding capability, etc.) while meeting the mechanical property requirements.

(2) Formation of casting defects is strongly related to molten metal cleanliness. Therefore, the liquid metal should be

cleaned to the highest level possible, meaning that the oxide inclusions and the dissolved gases are minimized to the point that they will not cause casting defects during solidification. The inclusions and dissolved gases can be reduced by various ways including flotation, sedimentation, and filtration. The most effective approach is flotation which can reduce both inclusion and dissolved gases simultaneously simply by introducing inert gas or active gaseous fluxes. To achieve the best results, the melt temperature, the bubble size, the number of bubbles and their distribution, and the bubbling location should be optimized. In general, low melt temperature and small bubble size is desired and it is also better to keep the bubbling location close to the dip well or pouring station.

(3) By reciting Campbell's claim, the necessity of stopping the pouring of liquid metals is becoming increasingly urgent. Pouring is the main source of entrained bifilms which are the root causes of many casting defects such as porosity, hot tearing, etc. As lattice dislocations explain plasticity, bifilms explain pore initiation and fracture initiation. Only when pouring is minimized (i.e. bifilms are reduced or eliminated) will casting processes begin to achieve their potential in delivering high integrity and reliable castings.

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References

[1] Wang Qigui. Latest Developments and Advances in Aluminum Shape Casting. China Foundry, 2013, 10 (1): 43–49.

[2] Caceres C H and Selling B I. Casting defects and the tensile properties of an Al-Mg-Si alloy. Mater. Sci. Eng. A, 1996, 220: 109–116.

[3] Cáceres C H, Davidson C J, Griffiths J R, et al. Hypoeutectic Al-Si-Mg Foundry Alloys. Materials Forum, 1997, 21: 27–43.

[4] Promise N E. Evaluation of non-ferrous materials (1956) cited by H E Boyer in "Atlas of fatigue curves", American Society of Metals, 1986.

[5] Major J F. Porosity Control and Fatigue Behavior in A356-T61 Alloy Castings. AFS Transactions, 1997, 105: 901–906.

[6] Couper M J, Neeson A E, and Griffiths J R. Casting defects and the fatigue behaviour of an aluminium casting alloy. Fatigue Fract. Engng. Mater. Struct., 1990, 13, 213–227.

[7] Skallerud B, Iveland T and Härkegård G. Fatigue Life Assessment of Aluminum Alloys with Casting Defects. Engineering Fracture Mechanics, 1993, 44: 857–874.

[8] Murakami Y and Endo M. Effect of defects, inclusions and inhomogeneities on fatigue strength. Fatigue, 1994, 16: 163–182.

[9] Odegard J A and Pedersen K. Fatigue Properties of an A356 (AlSi7Mg) Aluminum Alloy and Its Life Prediction. SAE technical paper, SAE-940811, 1994: 25–32.

[10] Ting Chih-Hsien. The Long-Life Regime Fatigue Analysis for Lost Foam Cast Al-Si Alloy 319. Ph.D. Dissertation, University of Illinois at Urbana-Champaign,1991.

[11] Lee F T, Major J F and Samuel F H. Effect of silicon particles on the fatigue crack growth behavior of Al-12Si-0.35Mg-(0-0.02)Sr casting alloys. Metall. Mat. Trans., 1995, 26A: 1553–1570.

[12] Lee F T, Major J F and Samuel F H. Fracture Behaviour of Al12wt.%Si0.35wt.%Mg(0–0.02)wt.%Sr Casting Alloys Under Fatigue Testing. Fatigue Fracture Engng Mater. Structures, 1995, 18: 385–396.

[13] Wang Q G, Apelian D, and Lados D A. Fatigue Behavior of A356-T6 Aluminum Cast Alloys. Part I. Effect of Casting Defects. Journal of Light Metals, 2001, 1: 73–84.

[14] Wang Q G. Fatigue Fracture Mechanism and Fatigue Life Assessment of Aluminum Castings. In: Proceedings of Materials Lifetime Science & Engineering, Edited by P.K. Liaw, R.A. Buchanan, D.L. Klarstrom, R.P. Wei, D.G. Harlow, and P.F. Tortorelli, TMS (The Minerals, Metals & Materials Society) 2003, Warrendale, PA: 211–222.

[15] Wang Q G, Crepeau P N, Griffiths J R, et al. The Effects of Oxide Films and Porosity on Fatigue of Cast Aluminum Alloys. Shape Casting: The John Campbell Symposium, eds: Murat Tiryalioglu and P.N. Crepeau, TMS 2005: 205–214.

[16] Gao Y X, Yi J Z, Lee P D, et al. The effect of porosity on the fatigue life of cast aluminium-silicon alloys. Fatigue & Fracture of Engineering Materials & Structures, 2004, 27(7): 559–570.

[17] Chan K S, Jones P, and Wang Q G. Fatigue crack growth and fracture paths in sand cast B319 and A356 aluminum alloys. Materials Science and Engineering A, 2003, A341: 18–34.

[18] Zhang B, Poirier D R, Chen W. Microstructural effects on high cycle fatigue-crack initiation in A356.2 casting alloy. Metall. Mater. Trans, 1999, 30A: 2659–2666.

[19] Wang Q G and Jones P E. Prediction of Fatigue Lives of Aluminum Shape Castings Containing Defects. Metall. Mater. Trans. B, 2007, 38(4): 615–622.

[20] Wang Qigui, Jones P E. Methods and Systems to Predict Fatigue Life in Aluminum Castings. U.S. Patent 7623973 B1, Nov. 24, 2009.

[21] Wang Q G, Jones P E. Fatigue Behavior and Life Prediction for Aluminum Castings in the Absence of Casting Flaws. SAE Int. J. Mater. Manuf., 2011, 4: 289–297 (SAE doi:10.4271/2011-01-0193).

[22] Wang Q G. and Jones P E. Fatigue and Life Prediction in Aluminum Castings. Modern Casting, 2013, 103(1): 1–10.

[23] Li Z M, Wang Q G, Luo A A, et al. Improved high cycle fatigue properties of a new magnesium alloy. Mater. Sci. Eng. A, 2013, 582: 170–177.

[24] Li Z M, Wang Q G, Luo A A, et al. High Cycle Fatigue of Cast Mg-3Nd-0.2Zn Magnesium Alloys. Metall. Mater. Trans., 2013, 44(11): 5202–5216.

[25] Li Z M, Luo A A, Wang Q G, et al. Effects of grain size and heat treatment on the tensile properties of Mg-3Nd-0.2Zn (wt.%) magnesium alloys. Mater. Sci. Eng. A, 2013, 564: 450–460.

[26] Bonsignore P V, Daniels E J and Wu C T. Calcium Metal as a Scavenger for Antimony from Aluminum Alloys. Argonne National Laboratory, Technical Report, October 4, 1994.

[27] Garat M, and Scalliet R. A review of recent French casting alloy development. AFS Transactions, 1978, 86: 549–562.

[28] Loper C R and Cho J-I. Influence of trace amounts of phosphorus in Al casting alloys – A review of the literature. AFS Transactions, 2000, 108: 667–672.

[29] Farahany S, Ourdjini A, Idris M H, et al. Effect of bismuth on microstructure of unmodified and Sr-modified Al-7Si-0.4Mg alloys. Trans. Nonferrous Met. Soc. China, 2011, 21: 1455–1464.

[30] Farahany S, Ourdjini A, Idrsi M H, et al. Evaluation of the

effect of Bi, Sb, Sr and cooling condition on eutectic phases in an Al-Si-Cu alloy (ADC12) by in situ thermal analysis. Thermochimica Acta, 2013, 559: 59– 68.

[31] Rathod N R, Manghani J V. Effect of Modifier and Grain Refiner on Cast Al-7Si Aluminum Alloy: A Review. Intl. Journal of Emerging Trends in Engineering and Development, 2012, 5(2): 574–582.

[32] Wang Q G, Apelian D, Arnberg L, et al. Solidification of the Eutectic in Hypoeutectic Al-Si Alloys. AFS Transactions, 1999, 107: 249–256.

[33] Taylor J. Metal related castability effects in aluminium foundry alloys. Cast Metals, 1996, 8: 225–252.

[34] StJohn D H, Dahle A K, Abbott T, et al. Solidification of Cast Magnesium Alloys. Magnesium Technology, Edited by Howard I. Kaplan, TMS (The Minerals, Metals & Materials Society), 2003: 95–100.

[35] Zhao Lei, Liao Hengcheng, Wang Qigui, et al. In-situ observation of porosity formation during directional solidification of Al-Si casting alloys. China Foundry, 2011, 8(1):14–18.

[36] Liao Hengcheng, Wang Qigui, Song Wan, et al. Effect of Oxide Level on Pore Formation in A356 Alloy by X-ray Imaging and Directional Solidification Technology. In: Proceedings of the 13th International Conference on Aluminum Alloys (ICAA13), Edited by: Hasso Weiland, Anthony D. Rollett, William A. Cassada, TMS (The Minerals, Metals & Materials Society) 2012: 1457–1462.

[37] Liao Hengcheng, Wang Qigui, Song Wan, et al. In-situ Characterization of Porosity Evolution in A356 Alloy Directionally Solidified Under Different Solidification Velocities.

JOM, 2012, 64(1): 22–27.

[38] Wang Q G and Gerard D. Methods of reducing (old) oxides in aluminum castings, 2012 (GMP020334).

[39] Kaufman J G and Rooy E L. Aluminum Alloy Castings: Properties, Processes, and Applications, AFS and ASM International, ISBN: 978-0-87170-803-8.

[40] Sigworth G K. Best Practices in Aluminum. AFS, June 2014.

[41] Sigworth G K. Practical Degassing of Aluminum. Modern Casting, 1988, 78(3): 42–44.

[42] Neff D V. Clean Metal for Aluminum Foundries -New Technology Using a Rotor Degasser and Filter Pump. AFS Transactions, 1990, 98: 579–584.

[43] Schmeisser D. Rotary Stirring Device for Treating Molten Metal. U.S. Patent 7,669,739, 2010.

[44] Wang Q G, Crepeau P N, Gloria D, et al. Improvement of Fatigue Strength in Aluminum Castings, in Advances in Aluminum Casting Technology II, eds: M. Tiryakioglu and J. Campbell, ASM International, Materials Park, OH, 2002: 209–218.

[45] Jolly M. Prof. John Campbell's Ten Rules for Making Reliable Castings. JOM, 2005: 19–28.

[46] Groteke D E. Improve Your Vacuum Test. Modern Casting, 2007, 8: 28–31.

[47] Campbell J. Castings. 2nd edition, Butterworth-Heinemann Limited, London, UK, June 12, 2003.

[48] Campbell J. Stop Pouring, Start Casting. Intl Journal of Metalcasting, 2012(3): 7–18.

[49] Tiryakioglu M, and Crepeau P N. Shape Casting: A Tribute to Prof. John Campbell. Metall. Maters. Trans. B, December 2006, 37B: 855.