

POST-CASTING PROCESSING

19

The cutting off of gates and feeders is a chore, significantly aided by not having feeders, or even running systems, if possible.

Zinc castings are special because they can be cleaned from flash by cryogenic tumbling as a result of embrittlement of the alloy by cooling to -196°C in liquid nitrogen, or possibly using a thermal deburring operation in which the castings are momentarily heated to 3000°C in a sealed chamber in which a mix of natural gas and oxygen are ignited to burn off flash and melt corners (Birch, 2000a).

High pressure die castings (HPDCs) of Zn, Mg and Al alloys can be clipped as a result of (1) their high accuracy and (2) their cleanness. The clipping of sand castings might be expected to result in a very blunt clipping tool, bringing the press to a full stop. However, these disadvantages cannot be too great because the process is currently being promoted for aluminium and iron sand castings (SERF, 2002; Ricken et al., 2009). The process is highly productive, taking seconds (not minutes by grinding for instance). Clean clippings are efficiently remelted, again contrasting with grindings. Operator problems such as white finger are also avoided.

Other alloy systems are much less easy. Feeder heads on steel castings can sometimes be removed by a powerful hammer blow, or with hydraulic wedges, although I suspect these fracturing processes are successful because of the density of bifilms, created by the poor filling technique, concentrating high up in the casting, particularly in the region of the feeder neck. The impact or wedge removal technique would probably not be possible with good quality well-made castings. Alternatively, large steel feeder heads are most often removed by flame or arc cutting. The troublesome cracks generated by the process are avoided if the filling system is improved so as to avoid oxide entrainment (founders are amazed and delighted that the cracks no longer appear after a good naturally pressurised filling system is applied to the casting!).

The removal of cores can be a challenge. Mechanical shock and vibration systems are widely used, as are high pressure water systems. Donahue (2006) describes a novel high velocity jet of liquid nitrogen, of similar density to water, as being highly effective for the removal of residual shell and core material from ceramic investment moulded castings.

Having completed de-moulding, de-gating, de-flashing and removing feeders, there is often still much to do to get the casting into the condition the customer has ordered (working in a foundry is a challenge).

This final chapter deals with these final issues. When these are complete, it is always a pleasure to see the castings going out of the door. (However, it is worth bearing in mind that finally getting paid for them is more certain if the castings are good.)

19.1 SURFACE CLEANING

A common treatment for the cleaning of castings of all types was *sand blasting*, in which particles of sand are entrained in a powerful blast of air directed at the casting. The process is now almost universally unused as a result of the concern about the generation of harmful silica dust.

Treatments nowadays include *shot blasting*, in which steel or iron shot is used, centrifuged to high speed by spinning wheels in so-called airless blasting. Stainless steel shot is, naturally, significantly more expensive than conventional carbon steel shot, but has the advantage for aluminium castings that they remain bright and free from rust stains.

Other treatments include *grit blasting* in which the grit is commonly a silica-free hard mineral such as alumina. The many other blast media include glass beads, plastic particles, walnut shells—almost anything seems to be used.

The particular benefit of blasting with a heavy shot such as steel is that the surface of the casting is deformed by the impact of each particle, generating a localised compressive stress at the site of impact. This is a peening process. Fathallah et al. (2000) finds 100% coverage of the surface is optimum to improve the fatigue resistance of a steel, whereas at 1000% coverage, the overlaps and scaling of the surface damage reduces the fatigue resistance once more.

Many other studies have confirmed that the compressive stresses in the surface of shot-blasted or shot-peened castings enjoy an improved fatigue performance as a result of the inhibition of the growth of surface cracks when subject to compressive stress. Improvements to alloy Al-7Si-0.5Mg were shown by Pitcher and Forsyth (1982), Ji et al. (2002) showed the benefits for ductile iron; Naro and Wallace (1967) improved steel castings. These experimental demonstrations have been supported by some excellent theoretical models by Fathallah et al. (1998) and Evans (2002).

Although the improvement of fatigue resistance following shot peening is perhaps to be expected, the effect of grit blasting is less easy to predict, because the fine notch effect from the indentations of the grit particles would impair, whereas the induced compressive stresses would enhance fatigue life. From tests on aluminium alloys, Myllymaki (1987) finds that these opposing effects are in fact tolerably balanced, so that grit blasting has little net effect on fatigue behaviour.

Kasch and Mikelonis (1969) noticed that the compressive stresses introduced into the surface by the peening effect caused significant movement of the casting on machining. Mroz and Goodrich (2006) found for both grey and ductile irons the depth of cold work and the residual stress did not change significantly after the first shot blast cycle.

However, the distortion effect is used to good effect in the sheet metal industry to induce the controlled forming of curved surfaces; aircraft wing panels are formed from flat sheets in this way; the flat product gradually curls up, becoming convex towards the direction of the impingement of the shot. The distortion effect of the peening treatment is controlled by the use of Almen strips. These are small strips placed alongside the article to be peened, and the curvature of the strips (measured as the arc height) is measured as a function of processing time (Cao et al., 1995). It seems that the use of controlled peening to adjust or vary the shapes of castings seems, so far as I am aware, not used at this time, but might be a useful technique to generate subtle curves on castings that might be difficult to mould.

Sand retention on as-cast surfaces of sand castings can be a problem. Figure 19.1 shows such a surface for an Al alloy cast in a zircon sand mould. A 10 kg cylinder head casting might have 0.5–1 g of sand retained on its surface. This compares with values of up to 0.05 g for a permanently moulded cylinder head which would have relatively few sand cores.

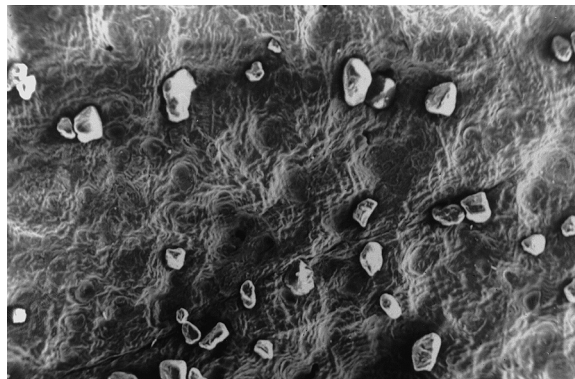


FIGURE 19.1

Surface of an Al alloy sand casting, showing adhering sand.

After a blasting treatment, the surface of the casting will have achieved some degree of uniformity of appearance and texture. For sand castings, however, not all of the sand particles are removed; a significant percentage of adhering particles are pounded in to the surface, thus becoming even more attached by the plastic flow of metal around and over them. This residual adhering sand is not important and not noticed for most applications. It might even be slightly beneficial for applications requiring paint adherence. However, it does affect the rate of wear of machine cutting tools if extensive machining of the surface is required.

The elimination of such sand from Al alloy sand castings can be achieved by a caustic etch. This seems to release most of the particles and improve machinability. However, of course, such chemical treatments are to be avoided if at all possible for environmental reasons.

When grit-blasting Al alloy castings, whether sand castings or permanent mould castings, residual grit is always found to be embedded in the surface. Simply picking up the casting and dropping it on to a cast iron surface table will reveal a ghostly outline of the casting on the table delineated by the grit that has been shaken out by the impact of the fall.

Little research appears to have been carried out on the prevention or reduction of sand retention on castings. We may speculate that, as suggested in Section 9.12 on surface finish, any process that thickens the oxide on the advancing melt would be beneficial. Thus slower filling will allow the oxide to grow thicker, and the surface will experience a slower build-up of pressure, delaying mould penetration while the surface oxide thickens further. Chemical additions to the alloy may also help, as suggested for Sr addition to Al-Si alloys. Finally, of course, the final overpressure applied by counter-gravity filling systems will also require to be reduced so far as possible.

19.2 HEAT TREATMENT

The heat treatment of castings is one of the most expensive of all the post-casting operations. Furthermore, high temperature heat treatments are, of course, the most expensive of these. The high temperature treatments include (1) stress relieving followed by a slow cool, and (2) solution or homogenisation treatments followed by a quench. In addition to the high costs, the treatments requiring a quench are significant dangers to the integrity and accuracy of the casting as we have discussed. There are both considerable economic and technical incentives to avoid these processes if at all possible.

The heat treatment of ferrous castings is a huge subject that cannot be covered here. In relation to castings, it is worth mentioning that alloy steel castings require to be homogenised at a temperature of at least 1300°C (Flemings, 1974). These are challenging temperatures, requiring highly specialised furnaces with vacuum or protective atmosphere, and specialised quenching facilities.

Similarly, for light alloys, the solution treatment is also expensive. It is performed at temperatures as near as possible to the melting temperature of the alloy, generally in the region of 500–550°C followed by a quench to retain solutes in solution.

The high temperature is a risk from the possibly incipient melting (or worse, wholesale melting). The quench is risky from the point of view of introducing several problems including residual stress and distortion. We shall devote some space to considering the important questions as to how these risks can be reduced or avoided.

19.2.1 HOMOGENISATION AND SOLUTION TREATMENTS

For a single-phase alloy, Flemings (1974) describes a simple and elegant model of the microsegregation, or coring, present in the dendrites, and how it can be reduced by a high-temperature heat treatment termed homogenisation. He defines a useful parameter that he calls the index of residual microsegregation, δ , as

$$\delta = \frac{C_M - C_m}{C_M^O - C_m^O} \quad (19.1)$$

where C_M = maximum solute concentration of element (in interdendritic spaces) at time t , C_m = minimum solute concentration of element (in centre of dendrite arms) at time t , C_M^O = maximum initial concentration of element, and C_m^O = minimum initial concentration of element.

The parameter δ is precisely unity before any homogenisation treatment. If homogenisation could be carried out to perfection, then δ would become precisely zero. After any real homogenisation treatment, δ would have some intermediate value that depends on the dimensionless group of variables Dt/l^2 . Here, D is the coefficient of diffusion of the homogenising element, t is the time spent at the homogenising temperature, and l is the diffusion distance, of the order of the dendrite arm spacing (DAS). Assuming a sinusoidal distribution of the concentration of the element across the dendrite, Flemings finds the solution, approximately, as:

$$\delta = \exp\left(-\pi^2 Dt/l_o^2\right) \quad (19.2)$$

where $l_o = (\text{DAS})/2$. Equation (19.2) is useful for the approximate prediction of times and temperatures required to homogenise a given cast structure. Flemings shows that, for a low-alloy steel, carbon is always homogenised by the time that the steel is heated to about 900°C for all normal values of DAS because of its high value for D (see Figure 1.4(c)). However, for the substitutional elements manganese and nickel, little homogenisation occurs below 1100°C, and for homogenisation to be about 95% complete ($\delta = 0.05$) requires 1 h at 1350°C if DAS = 50 µm. The fine DAS value is obtained by ensuring that such critical parts of the casting are within 6 mm of a chill. If no chill is used and the DAS = 200 µm or more, then practically no homogenisation is achieved ($\delta = 0.95$) at this time and temperature.

Flemings emphasises that the normal so-called homogenisation treatments for steels based on temperatures of 1100°C achieve only the homogenisation of carbon. The more recent use of vacuum heat-treatment furnaces capable of 1350°C and higher has produced very large improvements in the mechanical properties of cast steels.

The term *homogenisation treatment* is reserved for treatments designed to smooth out concentration gradients within a single-phase alloy.

The term *solution treatment* applies to those treatments designed to take into solution (i.e. dissolve) one or more second phases. These are also discussed by Flemings (1974). His presentation is summarised next.

Flemings considers a binary alloy containing a non-equilibrium eutectic. The dendrites are again assumed to be cored, having a sinusoidal distribution of solute as before, but containing interdendritic plates of divorced eutectic; for instance, in the case of the Al-4.5Cu alloy, a single plate of CuAl₂ phase separates the cored aluminium-rich dendrites. Dissolution of the interdendritic second phase is assumed to be limited by diffusion in the α -phase. If f and f_o are the volume fractions of eutectic at times t and t_o respectively, then the answer is similar to that seen in Eqn 9.10, approximately:

$$f/f_o = \exp\left(-2.5Dt/t_o^2\right) \quad (19.3)$$

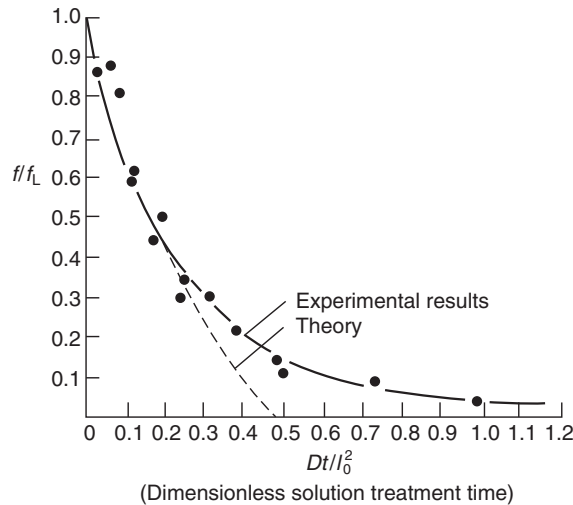
If the DAS is 100 µm, and if impurity levels in the alloys are kept low, so that solution temperatures within 10 or 20°C of the melting point can be employed without danger of melting the alloy, then a 10 h solution treatment is needed to dissolve all the second phase. This large DAS is easily reduced to below 30 µm by chilling the casting, so that times shorter than 1 h are easily attained.

Experimental tests of the theory show good agreement, particularly at short times (Figure 19.2). At long times, the dissolution of the last traces of segregate require more time than the simple theory predicts because more segregate exists between primary arms than between secondary arms, and so the last remnant of solute must diffuse over larger distances than simply the secondary DAS.

19.2.2 HEAT TREATMENT REDUCTION AND/OR ELIMINATION

There is strong motivation to abandon or significantly modify the high temperature solution treatment for Al alloys. This has taken three different options:

1. The total avoidance of any artificial heat treatment, using those alloys that exhibit natural ageing at ambient temperatures such the 7xx alloys based on the Al-Zn-Mg series (Sigworth et al., 2006). At room temperatures, the development of properties is particularly slow, taking days or months, but only an hour or less at economic ageing temperatures in the region of 100°C.

**FIGURE 19.2**

Rate of solution of eutectic in Al-4.5Cu alloy as a function of (dimensionless) time.

Data from Singh et al. (1970).

- Avoidance of solution heat treatment and quenching, but using heat treatments that involve only ageing at temperatures in the region of only 150–250°C. These treatments often take several hours, although I personally aim to get the treatment done within an hour at most by opting for higher temperatures.
- Using only minimal solution treatment times and temperatures, followed by quenching and minimal ageing. For instance, a typical solution heat treatment for the common structural engineering alloy Al-7Si-0.4Mg might be 6–12 h at 530°C followed by quenching and finally ageing for up to 12 h at 160°C. Personally, I always aim for a treatment not longer than 1 h on the grounds that, if cast well, the alloy will display more than adequate properties to pass all specifications. The various ways to reduce these excessive treatment times (optimised by metallurgists in an age of plentiful and low cost energy, and now completely inappropriate) are discussed later. In the meantime, as an instance, the treatment described previously would give nearly equivalent results at 1 h at 540 or 545°C if possible, and 1 h at 200°C.

The use of the rule ‘a 10°C rise doubles reaction rates’ is illustrated in Figures 18.8 and 18.9. The curves of strength versus time all collapse onto a single curve if allowance is made for the 10°C rule. It is strongly recommended that all heat treatment results are presented in this way. Occasionally, the curves will not precisely superimpose, but this effect is easily distinguished by a fanning outwards of the curves, or the generation of a set of curves that are clearly part of a family, and so largely parallel or uniformly diverging. As an example the treatment of an Al alloy at 180°C for 4 h is more or less exactly equivalent to a treatment at 190°C for 2 h, or 200°C for 1 h. These are all effectively *equivalent* treatments. One would expect an identical microstructure and identical properties from all of these treatments because they are *equivalent*. Although, of course, there is a limit to how far this translation of the treatment can be pushed to higher temperatures and shorter times, for most practical purposes on the shop floor, it is often surprisingly effective over many iterations.

From time to time, various stepped solution treatments are proposed. From the point of view of achieving the very best properties, such treatments are probably necessary. For instance, Gauthier and Samuel describe solutionising at 515°C for 12 h followed by 540°C for a further 12 h which gives excellent strength and ductility for the common automotive alloy 319.2 (Al-6Si-3.5Cu). Their optimum ageing treatment is 155°C for 12 h. However, such a daunting

high-cost treatment for such a low-cost alloy can hardly be recommended. For instance, the ageing treatment can be seen to be approximately equivalent to 1 h at 200°C. By contrast, Rometsch et al. (2001) study the solution treatment of 356 alloy (Al-7Si-0.4Mg), finding with a DAS of 40 µm that the alloy can be solution treated at 540°C in a time somewhere between 4 and 15 min.

For those alloys produced by HPDC that would blister and distort with even my kind of reduced treatment, Mehta (2008) reports that treatments as short as 10–15 min at temperatures of only 430–490°C, followed by quenching into water, result in useful properties when combined with ageing treatment for T4 condition (22°C), T6 (150°C) and T7 (200°C). Lumley et al. (2007) confirm the top temperature of 290°C for 15 min for 380 alloy to avoid blisters, although the 20 h age at 150°C should be substituted by 15 min at 210°C to create a useful energy saving assuming the 10°C rise equivalence to doubling rate.

Thus the conventional wisdom that HPDC material cannot be heat treated would be normally true for conventional HPDC and conventional solution heat treatments, but it seems it may be possible to develop a useful degree of improvement as outlined previously. The reduced temperatures and times are possible because of the extremely fine DAS. In addition, improvements in HPDC involving evacuation of the die and careful control of the shot stroke all contribute to increasing the soundness and reduced bifilm content of HPDC material, further allowing the benefit of some kind of solution treatment if necessary.

Another related aside might usefully be introduced here. It is unhelpful to see heat treatment results plotted on a linear time scale because the short times are necessarily too compressed to be properly discernible. I find a logarithmic plot based on the factor of 2 extremely convenient for the plotting of heat treatment results. Thus the time axis has equal divisions labelled 1, 2, 4, 8, 15 and 30 s; 1, 2, 4, 8, 15 and 30 min; and 1, 2, 4, 8, 16, 32 h etc. The small error introduced by using 15 instead of 16 in this series can easily be shown to be quite negligible in relation to the other inevitable experimental inaccuracies that will be naturally present in the plots. Another useful approximate log base two scale I sometimes use is 1, 2, 4, 8, 16, 32, 64, 125, 500, 1000, 2000, 4000 etc. Figure 18.8 shows an example of this kind of plot, combined with the 10°C effect of the doubling of reaction rates, allowing all of the different temperature results to be collapsed, with acceptable accuracy, onto a single curve. This technique for recording all kinds of diffusion-rate-limited reactions is warmly recommended.

19.2.3 BLISTER FORMATION

A further problem can occur for those alloys cast turbulently, such as by HPDC. The hydrogen absorbed from the atmosphere of the heat treatment furnace can diffuse into the part, inflating internal bifilms. Thus blisters can occur (a localised distortion) if the bifilm is close to the surface, whereas if the inflated bifilms are deeper inside the casting distortion of the whole casting can occur. Figure 19.3 shows blisters developed on a rolled product produced from a defective cast ingot. Blisters are common on plaster moulded castings of Al alloys produced from relatively dirty metal and are even seen on the surface of premium quality forgings produced from less-than-well-cast products.

In the past, of course, the essential role of the bifilm has not been appreciated because of all attention being focussed on the role of hydrogen. This is understandable because the introduction of a little NaBF₄ in the furnace atmosphere completely eliminates the problem (Dorward, 2001). This illustrates that some kind of absorption mechanism is taking place in which the absorption of hydrogen at the casting surface is inhibited by some kind of passivating action of the fluoride. In similar work, Zurecki (1996) finds that low levels of sulphur hexafluoride gas suppress blister formation during the heat treatment of Al-Mg alloys. Other halogen gases (those containing Cl and/or F) would also be expected to be effective to different extents. These solutions to blister problems are unwelcome because of the toxicity of the chlorides and fluorides.

As a personal experience of these problems, Al alloy castings produced by counter-gravity casting never displayed blisters from heat treatment even in atmospheres containing high levels of water vapour (the work basket was unfortunately constructed from square tubular steel, with the result that the tubes filled with water during the quench, but boiled off when returned into the furnace, giving a nearly pure steam atmosphere). However, gravity poured products treated in the same furnace, even after heavy forging, would, on occasion, produce crops of unsightly blisters, causing the parts to be scrapped.

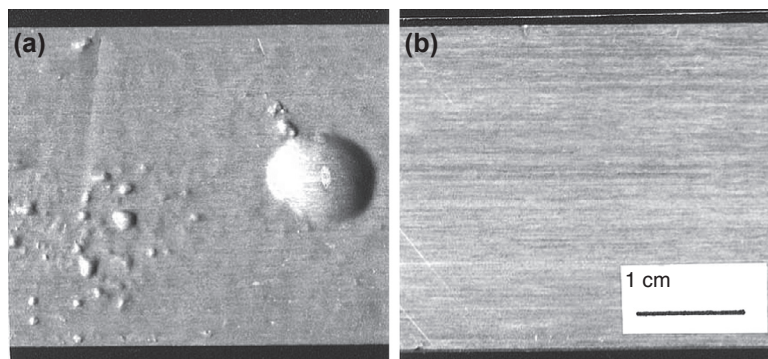


FIGURE 19.3

Surfaces of stretched sheet of 7475 (Al-Zn-Mg-Cu) alloy heated to 515°C (a) in air atmosphere; (b) with NaBF₄ in furnace atmosphere (Dorward, 2001).

Blister defects in rolled products are likely to be the consequence of the decoherence of near-surface films, often aided by the precipitation of hydrogen into such ready-made fissures. Celik and Bennett (1979) have shown that inclusions are implicated, and filtering of the melt before casting is an effective way to control the blistering behaviour of the rolled product. This observation is further confirmation that rolling is not especially effective in bonding or welding such defects to render them harmless.

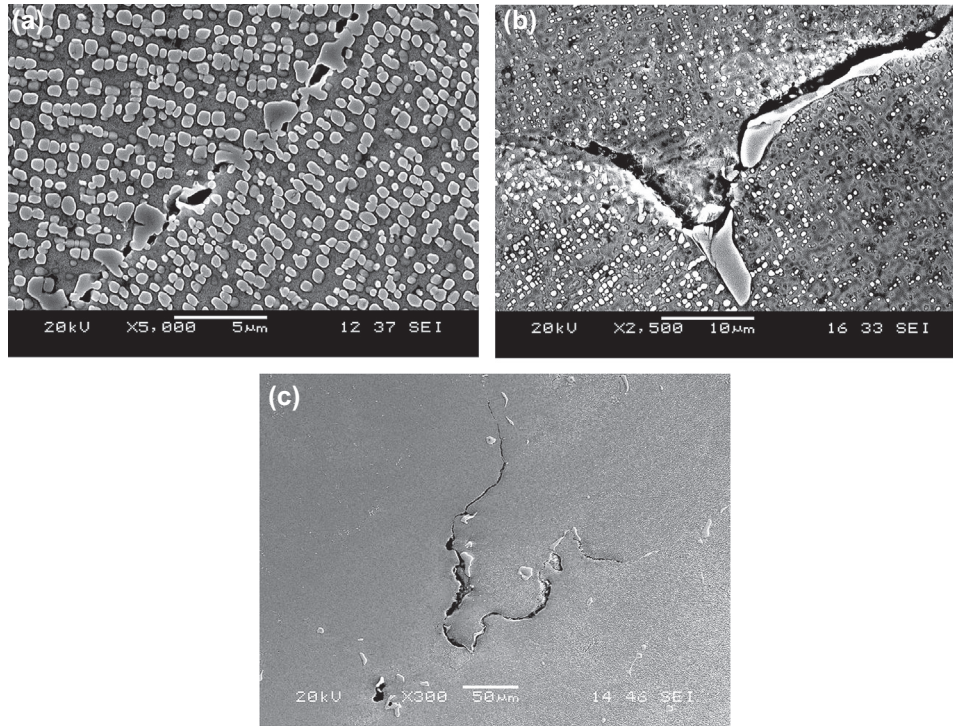
19.2.4 INCIPIENT MELTING

The use of solution treatment temperatures as high as possible, taking advantage of the general rule that an increase of 10°C will double the rate of reaction, is of course limited to the temperature at which the alloy will melt. The sight of the castings flowing out from under the door of the heat treatment furnace is to be avoided if at all possible.

However, well before the castings become too hot to melt entirely, they can suffer from overheating; a phenomenon known as ‘incipient melting’. This phenomenon was a puzzle in the sense that on heating above some low melting point constituents of the alloy, that might constitute only 1–2 volume % of the alloy, these constituents would clearly melt. However, on subsequent cooling, they should re-solidify so as to result in no net effect. However, this was untrue for most alloys and most castings. Their mechanical properties, particularly ductility, would be significantly impaired by overheating.

An exception known to me was the early Cosworth castings, which were exceptionally clean and did not suffer loss of properties if heated above the melting point of some phases, but were in fact observed to continue to improve in properties. Researchers have occasionally reported (for instance, Zhang et al., 1998) that temperatures up to 560°C have been perfectly acceptable for Al-7Si-0.4Mg alloy in contrast to most of the industry that works at a maximum temperature of 540° or perhaps 545°C, in view of the known melting points of Fe-rich phases in the region of 550–555°C. This gave a clue that the presence of bifilms, as an uncontrolled influence, may be a factor contributing to the permanent degradation of properties when overheated. In confirmation of this suspicion, the author has proposed a mechanism that appears to fit the facts (Campbell, 2009a). This is outlined next.

On heating over the melting point of some of the low melting point grain boundary constituents, these phases will melt, and so expand in volume. This expansion is not easily contained because both the newly arrived liquid phase and the surrounding solid are both effectively incompressible, resulting in the plastic deformation of the surrounding solid matrix. These millions of tiny expansions will result in a tiny growth of the whole casting. The resulting high compression stress around the melting phases will ensure that any bifilm on which the phases formed will remain tightly shut. However, on subsequent cooling and solidification, the previously melted phase will now occupy less

**FIGURE 19.4**

Grain boundary melting phenomena in the HAZ of welded Ni-base alloy 738LC showing (a) cracked and decohered grain boundary carbides; (b) asymmetric features of the crack and melted regions suggestive of an asymmetric bifilm; (c) a convoluted crack reminiscent of an opened bifilm (Sidhu et al., 2005).

volume as a result of natural solidification shrinkage. The overall plastic growth of the casting is not reversed because the nearby bifilm accommodates this reversed stress by simply opening (because there is no bonding between its internal surfaces). This is vastly easier than the problem of reversing all the plastic flow. Thus, the alloy now suffers from grain boundary porosity in the form of expanded bifilms.

The presence of bifilms at the sites of incipient melting is strongly indicated by Ni-base alloys, and is common in heat affected zones (HAZ) of welds that experience temperatures not only in the incipient melting range but naturally up into the complete melting region. Figure 19.4 shows instances of incipient melting in the HAZ of welded 738LC Ni-base superalloy (Sidhu et al., 2005). In the case of Figure 19.4(a), the bifilm is thin and symmetrical, suggesting it is a rapidly entrained alumina type; the $M_{23}C_6$ particle has cracked through its centre, and grain boundary γ' particles also contain traces of cracks. In contrast, Figure 19.4(b) shows MC carbides on only one side of the crack, suggesting the originating bifilm to be asymmetrical, perhaps an alumina plus a spinel? Figure 19.4(c) shows a typical convoluted bifilm, opened by the melting of the grain boundary phase in the heat affected zone of the weld.

19.2.5 FLUID BEDS

Over recent years, there has been significant interest in the use of fluidised beds for heat treatment, particularly for Al alloys (Chaudhury and Apelian, 2006). The advantages include extreme uniformity of temperature and good heat transfer

leading to rapid heat up of parts. In contrast, a muffle furnace with circulating air often has a problem to achieve the 2.5°C uniformity required for demanding work, in which all nine checked points—the eight corners and the centre—of the furnace are required to be within the 2.5°C tolerance.

19.2.6 QUENCHING

When used for quenching, the rate of cooling in a fluid bed is only about half that of a water quench. A slower quench is almost certainly a benefit for some castings, where the water quench is usually too fast, creating severe residual stress problems. In fact, the relevant question is, ‘is the fluid bed quench slow enough?’ In general, as we have seen in Section 10.9, the loss of properties as a result of a slower quench is relatively minor compared to the overall benefit, which can be extremely valuable. Ragab et al. (2013) describe its use for quenching of Al-7Si-0.4Mg alloy.

Ramesh and Prabhu (2013) propose the use of a cylindrical probe which can be used to quantify a quench action. They define a dimensionless cooling parameter

$$D^2(dT/dt)/\alpha\Delta T$$

where D is the probe diameter, dT/dt is the cooling rate, α is the probe diffusivity, and ΔT is the initial temperature difference between the probe and the quench medium.

For those who want to know the response of properties to differing quench rate, the technique described by Dolan et al. (2005) is recommended. These researchers used a variant of the Jominy end quench technique, usually reserved for the investigation of steel heat treatments, in which water is jetted on to the end of a pre-heated 25 mm diameter bar of the alloy to be tested. Thermocouples are inserted in holes drilled at intervals of 2, 38 and 78 mm long, the length of the bar to ascertain the spectrum of cooling rates with increasing distance. These are correlated with hardness measurements at 2 mm intervals along flats machined on both sides of the bar. In addition, of course, the structural changes induced by different quench rates can also be studied.

19.3 HOT ISOSTATIC PRESSING

In his history of hot isostatic pressing (HIPping), Mashl (2008) describes how pressing from all directions on a component, using a gas at high pressure, was found to densify ceramics as early as the 1950s. In 1970, it was first used on Al alloy castings, and within 5 years had started to become widely used for light alloy aerospace castings. In a significant historical paper, Coble and Flemings (1971) confirmed that pores, if sufficiently fine, and in a fine-grained matrix, would gradually disappear given a few tens of hours and a temperature high enough for a ‘sintering reaction’ to occur. They found that the application of modest pressure, about 20 atm (2 MPa), greatly assisted the process. The development of HIPping was the outcome. Much higher pressures were employed, usually nearer 1000 atm (100 MPa), and temperatures as near the melting point as was practical. The optimum conditions for HIP have been defined in an elegant study by Arzt et al. (1983) in which they characterise tool steel, superalloys, alumina and ice.

The significant improvement in mechanical properties, particularly average fatigue life, reported for certain alloys after HIPping is probably at least in part due to the contribution towards the de-activation of entrained double oxide film defects (bifilms) as fatigue crack initiators. We shall evaluate the evidence here for HIPping as a solid-state process for the closing of pores and bifilms.

When a cast Al-7Si-0.3Mg alloy with oxide film defects, like that shown in Figure 19.5(a), is subjected to HIPping treatment, the applied pressure at temperature close to its melting temperature induces a substantial plastic deformation in the casting causing the defects to collapse and their surfaces to be forced into contact (Figure 19.5(b)). This plastic collapse phase occurs almost immediately, probably requiring only a few seconds or a few minutes. The slightly extended time may be required for the residual air to be reacted to form oxides and nitrides. At this stage, the pores and cracks are closed but, of course, not necessarily bonded. Thus only a modest, if any, benefit from HIPping may be expected at this early stage. Also noticeable in Figure 19.5(b) are residual pores among the collapsed oxide bifilms that appear to have survived the complete HIP process. It seems likely they contain the insoluble 1% argon constituent from the air. Such pores are never expected to close, but will contain the argon at the huge pressure experienced during the HIP process and

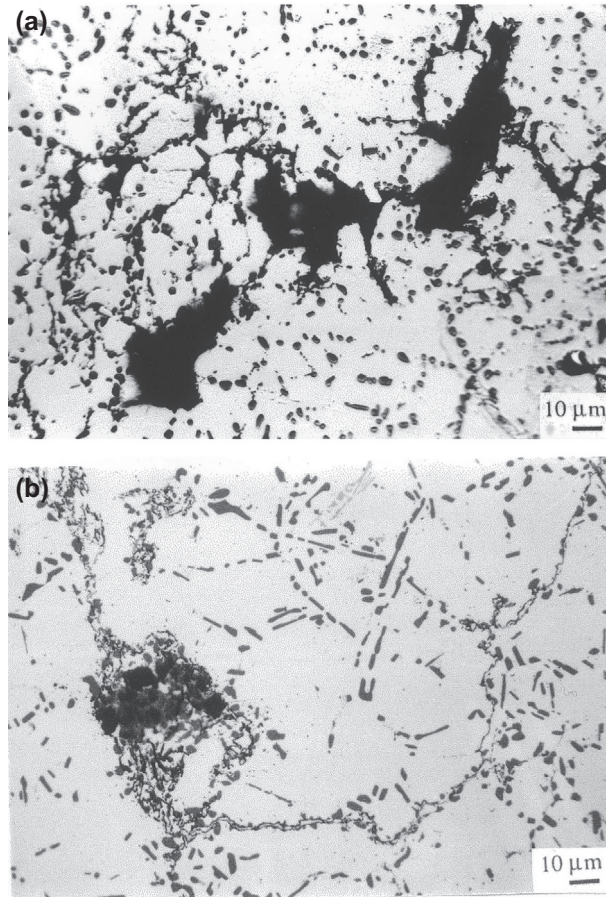


FIGURE 19.5

Optical micrograph of an Al-7Si-0.4Mg alloy showing (a) a network of oxide films and associated pores in the as-cast condition; (b) the network of films and collapsed pore in the HIPped condition (Nyahumwa et al., 2000).

will pressurise most of their associated bifilm, and thus will expand again if the temperature is raised sufficiently. Such pores might be expected to re-grow to some extent, triggering failure during such processes as creep or super plastic forming.

For those Al alloys containing no Mg for which the oxide is the very stable ceramic, alumina, Al_2O_3 , there is little or no benefit to be expected from extending the Hip cycle. Similarly, for those alloys containing more than approximately 1%Mg, once again the oxide is extremely stable, being pure magnesia, MgO. For such alloys, no bonding is expected to develop between the extremely stable ceramic surfaces dividing the central interface of the bifilm.

The situation is interestingly different for many aluminium alloys of intermediate Mg content in the range of approximately 0.05–0.5%, as for instance typified by Al-7Si-0.4Mg alloy. For such alloys, an alumina (Al_2O_3) film is first formed during the rapid entrainment process. However, after an incubation period that is a function of time and temperature, the entrained film transforms to magnesium aluminate spinel, MgAl_2O_4 , (Aryafar et al., 2010). This involves a volume change and atomic re-arrangement of the crystal structure that would be expected to encourage diffusion bonding

across any oxide–oxide interface that happened to be in contact. Analysis of bifilms that have acted as fatigue crack initiation sites have confirmed their conversion to a spinel and confirms that fatigue properties are improved by HIPping (Nyahumwa et al., 1998, 2000). A further positive finding from this work was that compared to filtered castings, the unfiltered but HIPped castings exhibited higher fatigue performance, despite larger maximum defect sizes, implying some degree of bonding across the crack. The application of HIP to castings shown in Figure 9.34(a) resulted in fatigue test samples that did not fail; the run-outs at the stress 150 MPa reached nearly 10^8 cycles (not shown). At the higher stress of 240 MPa, improved fatigue lives were still recorded (Figure 9.34(b)) although it is of interest that in all failures, with the possible exception of the most resistant specimen, the fatigue failures still occurred from oxides, being probably healed or partly healed bifilms.

The liquid-state healing mechanism as described in Section 2.4, and the HIP solid-state healing mechanism are suggested to be analogous. However, there are interesting differences. In the liquid state, where bifilm are floating freely in the melt, the healing mechanism operates at high temperature (i.e. approximately 700°C for an aluminium alloy) and with only the moderate external applied pressure (<0.1 MPa) because of depth in the liquid. When the casting has solidified, the solid-state healing process due to HIP operates at lower temperature but very much higher pressure (i.e. 100 MPa at 500°C for Al-7Si-Mg alloy). In both the liquid and solid conditions, pressure inside the bifilm will be expected to fall as the oxidation reaction proceeds. When all the oxygen is consumed, the nitrogen is subsequently consumed to form nitrides (Raiszadeh and Griffiths, 2008).

Whether bifilms are actually ‘healed’ (i.e. effectively welded) clearly relates to the chemistry of the alloy and its films. Some bifilms appear to heal as we have seen previously, whereas others are resistant. These important factors are not well researched at this time. Thus, HIPping has limitations that do not appear to be widely known or understood. However, evidence for possible mechanisms is discussed below.

In contrast to this beneficial action of HIPping found by Nyahumwa and others as described previously, Wakefield and Sharp (1992) observed HIPping to have no beneficial effect on the fatigue properties of Al-10 Mg alloy castings, despite the closure of pores and cracks. The inference is that the bifilms in this alloy proved impossible to de-activate, resisting effective welding. To restate and expand on the phenomenon, this is attributed to the magnesia (MgO) film formed during oxidation of Al-10 Mg and entrained during casting. The magnesia film is (1) thicker and (2) has a stable structure that does not transform during HIPping (Nyahumwa et al., 1998, 2000). This lack of any substantial atomic movement explains the inert nature of the highly stable MgO compound. (Similarly, again, as discussed previously, it would be expected that Al alloys with very low Mg contents, that would be expected to contain entrained alumina films would be similarly resistant to HIPping because of the great stability of alumina in the absence of sufficient Mg to convert it to a spinel structure.)

A further sobering example can be quoted from failed attempts to hip Ni-base superalloys. During the early development of the Pegasus engine for the Harrier Jump Jet, 25 polycrystalline Ni-based alloy turbine blades that had previously been scrapped because of their content of porosity were subjected to HIPping, and were fitted to a test engine alongside sound blades to evaluate whether HIPping might be a satisfactory reclamation technique for blades. The HIPped blades failed within a few hours, damaging the engine and forcing a rapid shut down of the test. The failures had occurred by creep cavitation at the grain boundaries of recrystallised regions in the centre of the castings. Almost certainly the original porosity would be caused by bifilms, probably of aluminium and/or chromium oxides entrained by the severe turbulence that is usual during the vacuum casting process. (As we have observed many times, the vacuum is known to contain plenty of residual air to ensure the creation of surface films.) The great stability of the films, formed at the high casting temperature, would ensure that they were resistant to any re-bonding action. The recrystallisation would have happened because of the large plastic strains that were a necessary feature of the collapse of the porosity. However, the subsequent grain growth would expand grains up to local barriers such as bifilms. Thus, the bifilms, effectively unbonded, and so acting as efficient cracks, were automatically located at the grain boundaries from where the failures were seen to occur.

The closure of internal cavities usually causes negligible changes to the overall dimensions of the casting if the pores are small and/or deep seated. For large or near-surface pores, however, the collapse of the surface of the casting in the form of a localised sink may scrap the casting if the depression exceeds the machining allowance. In a severe case, the surface may puncture, opening up the internal cavity to the surface (Zeitler and Scharfenberger, 1984).

Naturally, HIPping cannot work if the pores are already connected to the outer surface of the casting. Such pores will never heal. Unfortunately, this is all too common. The existence of the various forms of surface-connected porosity is well known to those who HIP castings. In particular, pressure die castings are woefully resistant to the benefits of HIPping because of their many surface-connected bifilms (traversing the so-called 'dense' outer layers of the casting). However, many gravity filled and even counter-gravity filled castings exhibit surface-connected pores as a result of bifilms intersecting the surface.

In addition, although an excellent start, even good filling of the castings will not guarantee a satisfactory response to HIP treatment if the alloy has a long freezing range and is poorly fed. As we have seen in Section 7.1.5, poor feeding in a long freezing range alloy can easily lead to conditions in which residual liquid is sucked away from the surface of the casting, subsequently expanding as shrinkage porosity inside the casting, to create surface-connected porosity.

For reliable HIP response, the surface of the casting must be sound.

Finally, therefore, although the mechanical properties of castings usually exhibit an improvement, in the sense that their *average* properties are raised, the Weibull modulus most often *falls*. This is a direct result of the closure and welding of some defects, improving the properties of some castings, but leaving some castings unaffected, for the various reasons we have seen. Thus the *scatter* of properties is increased. Regrettably, this is one of the greatest disadvantages of HIPping, often overlooked.

19.4 MACHINING

There are numerous text books on machining, so we shall limit our attention to only a few of those aspects of relevance to castings.

An early experience of mine taught me that casting defects can have a profound effect on the difficulties experienced by the machinist. A Ni-base superalloy turbine blade for a land turbine was top poured through the root of the blade, creating oxide flow tube defects that opened into vertical cracks when grinding the 'fir tree' roots. The founder asserted the problem was the result of severe grinding, the heat of grinding producing grinding cracks. However, in fact, even the lightest grinding had difficulty to eliminate the problem.

Turning to the wide field of cast irons, Teague and Richards (2010) review the information on the age strengthening of grey irons, finding that ageing for 5 days at room temperature greatly improves tool life. Apparently, the effect is the result of the precipitation of nitrides. They report an investigation that attempted to reduce this time by ageing at 271°C for 75 min which yielded no clear result. It seems to me this attempted treatment was way over the top. Taking the 5 days at room temperature as 120 h at 20°C, and using the useful approximation that a 10°C rise in temperature will double the rate of reaction, this gives equivalent treatments at 60 h for 30°C, 30 h for 40°C etc.; finally indicating 1 h at 90°C. This extrapolation of times and temperatures makes no claim to be more than an approximation, but I have often found it usually to be impressively accurate. It seems more modest times and temperatures have not yet been tested.

Machining of grey cast iron can pose additional difficulties (Hoff and Anderson, 1968) because the skin typically contains the following.

1. No graphite but only fine pearlite (equivalent to a hard steel)
2. Oxides
3. Sand grains

For this reason, machinists are usually keen to ensure that the first machining cut is taken well under the casting skin. This is not always possible, however, and the effects of decarburisation and inclusions are noted below.

Rickards (1975) has researched the problems of poor machinability as a result of the formation of graphite-free surface layers on grey iron castings made in greensand moulds. It seems the problem is affected by section size and the carbon content of the sand. For instance for a 300 mm section thickness, the graphite-free layer can reach 1 mm. He found 5% coal additions to the sand to roughly halve the problem, and additions up to 10% would control the problem in thin-walled castings, but could cause a carburised layer in castings with heavy sections up to 100 mm thick.

The author recalls those working in a machine shop drilling ductile iron crankshafts who could tell the difference between those castings made with good and bad filling system designs simply by listening as they walked among the

machines. Those that were cast with significant surface turbulence contained inclusions rich in magnesium oxide, creating the characteristic high pitched singing noise that the machinists knew signalled a poor finish and short tool life.

The effect of oxides in castings is well known to machinists. The alumina inclusions in light alloys and mixed oxides in iron and steel castings are all known to cause defects on the machined surface by dragging out, leaving unsightly grooves. Worse still, the cutting edge of the tool is often chipped or blunted by encounters with such hard particles, that are in general vastly harder than the cast alloy, and even harder, in many cases, than the tool tip.

For sand castings of all types, but particularly the softer light-alloy castings, and particularly when turbulently filled, the presence of residual sand on the surface of the casting is often a problem. For iron castings, it is widely appreciated that the first cut should be sufficiently deep to get below the glassy surface inclusions.

In light alloys cast in sand, the complete removal of all sand is practically impossible short of dissolving away part of the surface in an acid or alkaline bath. Normally, some residual surface grains are broken and hammered into the surface by the so-called cleaning processes such as shot blasting. Grit blasting can be even worse because particles of grit (often hard, angular, abrasive grains of alumina) remain embedded in the surface.

Because of this residual surface aggregate problem, the slightly impaired machinability of sand castings of all types, ferrous and non-ferrous, compared with die (permanent mould) castings is one of the very few potential disadvantages of sand castings. This disadvantage seems little known and publicised, but needs to be acknowledged and accepted by all the parties in the casting supply chain to save future heartache and cancelled orders.

It is worth emphasising once again benefits of gentle uphill filling of mould cavities. The beautiful glossy shine that iron castings exhibit when filled by non-turbulent bottom-gated systems illustrates the protection provided by the lustrous carbon film acting as a mechanical barrier between the liquid metal and the mould. Because the liquid metal now no longer comes into contact with the sand mould, the surface of the casting should be completely free from embedded sand grains, and should exhibit excellent machinability. Similar benefits are achieved for most casting alloys by the insulating action of the oxide film when carefully uphill filling moulds. At this time, I am not sure that this prediction has ever been tested.

For Al alloys, the machining of high Si alloys (in the range approximately 10–20% Si) is well known to require diamond tipped tools. However, what is less well known is that even the lower Si alloys present some degree of tool wear. Armstrong and Martin (1974) draw attention to the excellent machinability of the 7%Zn alloy (771.0) that requires only a low temperature age (180°C for about 4 h) followed by an air cool (not a quench), resulting in a highly stable casting with low internal stress that maintained its highly accurate dimensions after machining, and enjoyed tool lives up to 20 times longer than the normal Si-containing alloys, plus machining speeds up to five times faster.

Dasch et al. (2009) describes investigations into the dry machining of Al alloys which would have significant environmental benefits compared with current wet machining methods, but has been prevented by the build-up of hot Al on tool tips. He finds only 0.15%Sn addition to an Al alloy achieved a 1000 fold increase in tool life, in addition to the ability to machine dry. Higher cutting speeds enhanced the effect, such that faster drilling worked better than low speed drilling.

For the future, the advent of other machining techniques will become increasingly important. These include machining or cutting by fluid bed, water jet, flame, plasma, oxygen, laser, spark and abrasive finishing. The smoothing of complex internal passageways in castings is sometime accomplished by forcing through abrasive mixtures of SiC in silicone putty. Even so, our current cutting technology based mainly on single point or multipoint cutting tools is likely to be with us for a long time yet.

As an interesting and potentially important aside, designers of light alloy castings often despair at the inability of threaded holes in castings to avoid stripping out of the threads when subjected to large loads. Thus relatively expensive heli coil inserts are specified. These strong steel inserts spread the pull-out load and significantly strengthen the fastener. However, these are costly and inefficient solutions. The author has experience of M8 self-tapping bolts (thread-rolling type rather than the thread cutting type) in an Al alloy casting that enjoyed no cost to machine the thread, and the better thread engagement and work hardening of the surrounding matrix resulted in 8 times greater pull-out load. The benefits of thread-forming fasteners in net-shape cast holes of light alloys has been further researched by Paxton et al. (2006). The work hardened region around the fastener acts as its own integral helicoil, spreading the pull-out load. Furthermore, the avoidance of a second metal minimises the potential for corrosion problems.

Although cast holes can be envisaged for HPDC, small diameter cast holes are not attractive for aggregate moulded castings because the small sand projections on cores and mould would be vulnerable to being slightly ill-formed or even broken off. Thus, for sand castings, it would be preferable to machine the tapping sized holes in the casting. This seems a relatively small penalty, still enjoying the benefit of avoiding tapping, and still achieving a greatly superior engineering feature which may be particularly important for certain critical castings.

19.5 PAINTING

For those painting processes where the paint is cured by heat, such as powder coating, the curing cycle is usually integrated with the heat treatment requirements for the casting, and thus constitutes the final heat treatment stage (the ageing) of the product. Aluminium alloy wheels for cars are usually heat treated and painted in this combined process.

However, problems arise if the casting exhibits any surface-connected porosity. This can cause a defect in the smooth surface of the paint known as a paint crater.

Before painting, the surface of the casting is usually machined to a bright finish and may be further mechanically polished and subsequently subjected to cleaning with a solvent. The paint powder is usually applied electrostatically to achieve a tolerably uniform deposit on a complex shape. During the heating of the casting for the baking cycle, the powder melts and slowly cures, while air expands from surface-connected cavities in the casting and bubbles through the melting paint. At a later stage, when the paint becomes viscous, the collapsing walls of the last blister tend to remain 'frozen' in place, becoming the crater wall (Figure 19.6(a)). Micks and Zabek (1973) investigated paint craters originally assuming them to be faults in the paint application process. However, they found that the craters were invariably linked to surface-connected porosity. The central hole in the crater was always connected to an internal cavity which was found to remain free from paint (Figure 19.6(b)).

Unfortunately, such craters are a serious source of scrap for aluminium alloy road wheels for cars, and kitchen utensils because the cosmetic requirement for such castings from the car owner and the cook, is simply perfection.

For Mg alloy castings, particularly those for aerospace applications, the special surface coatings to protect the casting from corrosion are a significant contribution to the cost of the product, and, because Mg alloys are so light, even make a contribution to its weight. The maintenance of the coating, often requiring replacement after weld repair or other disturbance represent further discouraging costs to the selection of these otherwise attractive alloys.

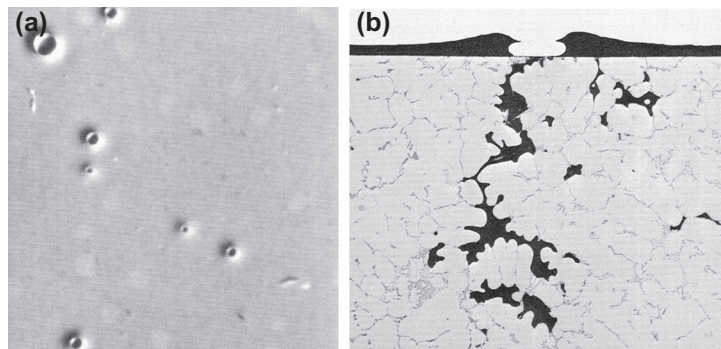


FIGURE 19.6

(a) Paint craters of varying sizes on the surface of an Al-5Si-0.5Mg alloy casting, acrylic powder coated and baked at 230°C (b) Optical micrograph of a section through a crater, showing its connection into surface-connected porosity (Micks and Zabek, 1973).

19.6 PLASTIC WORKING (FORGING, ROLLING, EXTRUSION)

The effects of bifilm populations in steels, and more particularly Ni-base alloys, can be so profoundly damaging that no plastic working can be applied; the cast ingot falls into pieces on the first stroke of the forge. Fortunately, most cast metals are not so bad, and plastic working can be successfully carried out as is discussed below.

It seems curiously perverse that most metallurgical texts continue to foster the erroneous assumption that working eliminates casting defects. Usually, the strains involved in most working operations are too low to effect any significant welding of faults. Bifilms in general are merely pushed around, if anything, growing worse before (if ever) getting better. The growth of casting defects by plastic working is well known to those who work in the forging industry. There are good reasons for this behaviour.

During the working of cast material, for instance by rolling, it is to be expected that the defects will be elongated in the direction of working. The elongation of the defect necessarily rotates it, aligning the defect along the rolling direction. This is clearly seen in the lengthening of graphite nodules during the rolling and forging of ductile iron (Neumeier et al., 1976).

In addition, of course, elongation of the work piece increases the area of its defect. The newly extended surfaces in the bifilm would necessarily be oxidised by the remnant of air entrained in the defect. The entrapped air would be contained partly amid the microscopic pores between the crystals of the oxide, and partly in reservoirs formed by folds and in expanded regions constituted by those bubbles entrained as part of the bifilm. During rolling, the continued oxidation of the expanding surfaces would hinder the welding of the interfaces that were being newly created, and by this means assisting the defect to grow as a crack, possibly to many times its initial size. The great effectiveness of the creation of this newly oxidised extension to the crack would be a consequence of the small oxygen requirement; the oxide would grow to a thickness of only nanometres at the hot working temperature (which is, naturally, much lower than the temperature at which the film was first formed on the melt). In other words, the entrained residual oxygen is highly efficient at these temperatures to continue the oxidation of new area as it is formed.

Miyagi et al. (1985) used ultrasonics to observe this increased area of cavities in 5083 alloy (Al-Mg type) during the early stages of hot rolling. Micro-cavities near the surface of the rolled plate seemed to close early, but those nearer the centre were long-lived. Micro-examination revealed that they were smooth sided, and appeared to have opened and expanded along grain boundaries. (The association with grain boundaries is, of course, a feature to be expected of bifilms.) They were seen to reduce in size only after reductions of more than 50%.

Later, if the extension were sufficiently large to consume the remaining air, further extension would result in the welding up of new extensions to the crack. It seems likely that normal forging and rolling do not reach this stage for most alloys, and so do little to heal defects, simply extending and re-aligning them as described previously. Processes such as extrusion and multi-pass rolling may sometimes account for a more complete elimination of the casting defects because of the much greater strains that can be involved. Even so, it seems that many defects remain, as indicated, for instance, by the evidence from corrosion (Section 9.10).

Work by Harper (1966) on the hot rolling of wire-bar copper showed that rounded bubbles of gas are collapsed asymmetrically, becoming pinched off, and forming what appear to be trails of minute beads on a string, the beads being microscopic bubbles only 30–40 nm diameter. It is hard to avoid the conclusion that the bubbles are the residual insoluble argon sitting in the collapsed bifilm, or even the collapsed bubble trail, resembling a string on the two-dimensional polished section.

Many bifilms will be already linked to the surface. Alternatively, during working operations such as rolling, they can become linked to the air by local plastic failure of ligaments of sound material separating the defect from the outside world. Thus, a corroding environment is expected to penetrate most near-surface entrainment defect cracks in cast, forged, rolled, and even extruded materials as is seen in Section 9.10.

The huge amount of work on the behaviour of inclusions in steels during working cannot be covered in this short account. However, the early work by Wojcik et al. (1967) was notable because they were able to show that their steels, destined for piercing to make tubes, contained a distribution of inclusions that they found to be log normal. This finding allowed them to extrapolate their inclusion counts from relatively few microsections to assess the probability of the presence in the billets of a small number of inclusions sufficiently large to cause wall defects in the tubes.

Charles and Uchiyama (1969) and Wardle and Billington (1983) studied the plasticity of inclusions in steels as a function of composition and temperature, showing the regimes in which the inclusions would either flow, thereby elongating, or fracture, so forming chains of fragments. Leduc et al. (1980) checked the density of cast low carbon steels during hot rolling, but did not find any evidence for a fall in density during the early stages of rolling similar to that found for the non-ferrous materials. They found a progressive increase in density, becoming fully dense at about 75% reduction. Examination of the microstructure indicated that all the pores were eliminated at that stage. This is probably understandable in view of the low residual levels of Al in the steels (0.02–0.04%) and absence of other strong oxide forming elements. Any surface oxide would therefore have been expected to be a liquid iron-manganese silicate, and so any bifilms, if present, would have been easily welded shut.

19.7 IMPREGNATION

The impregnation process is a sealing technique, designed to seal porosity and eliminate leakage problems in castings.

Clearly, good casting processes do not need impregnation.

However, many do.

For instance HPDCs are highly prone to leakage problems because of their high content of bifilms and inflated bubble trails. Similarly, low pressure castings that have not had the benefit of the roll-over action following the filling of the mould are sometimes prone to an interconnected type of shrinkage porosity as a result of convection problems.

Although some casting operations use impregnation only to seal those castings that are shown to leak, others do not carry out an initial sorting test, but simply apply impregnation to all castings. Although, like heat treatment, impregnation is often carried out by specialist offsite operators at a price usually based on the weight of the casting, it is sometimes installed as an integral part of the foundry operation.

The impregnation process involves the placing of the casting in a vessel from which the air is evacuated. The casting is then lowered into the sealing liquid. When fully immersed, air is re-introduced into the vessel to restore atmospheric pressure. In this way, the liquid is forced into the evacuated pores in the casting. The casting is then raised out of the liquid and is allowed to drain. Excess sealant is then washed off and the sealant is cured.

There are two main curing processes based on two different sealing systems. These are (1) sodium silicate cured with the addition of a catalyst to the liquid and (2) a thermosetting resin hardened by a subsequent low temperature heat treatment.

A recent development is the streamlining of the process by the use of special sealants, allowing the process to be carried out in a single vessel with only a short treatment time (Young, 2002).

In a small percentage of cases, leakage of the casting is not cured by the first impregnation. Some casters and their buyers allow two or even three such attempts. Usually, if the casting still leaks after repeated attempts to seal it, it is finally scrapped.

19.8 NON-DESTRUCTIVE TESTING

Although destructive testing is the most discriminating and reliable way to determine the suitability of a casting for service, the destroyed casting is somewhat unfit for service. Clearly, it is essential to have reliable tests that do not involve destroying the product. Naturally, this criterion puts severe limitations on the appropriateness of the various approaches to the solution to this problem.

19.8.1 X-RAY RADIOGRAPHY

Radiography is a powerful technique for checking the presence of major defects such as open cracks, porosity or foreign (entrained) inclusions. The traditional radiographic limit to detection has been approximately 2% of the section thickness, although modern techniques are providing ever greater sensitivity of detection. If 1% could be achieved this would be 100 μm in a 10 mm section. This is a substantial defect, especially if it is a crack. In general, however, a crack

can usually only be detected if it is aligned with the viewing direction. Even then, if the crack is tightly closed, the crack can remain undetected. Thus it has to be concluded that radiography cannot offer reliability for the detection of cracks.

It is well to remember that the features noted on radiographs generally appear to be shrinkage porosity are nearly always *not* shrinkage, but clusters of bifilms together with their entrained internal air. Rounded pores generally identified as ‘gas’ are also most often entrained *air bubbles*. The reader needs to be aware that at the time of writing practically all radiographic inspections are diagnosed incorrectly. The accurate diagnosis of radiographs cannot be expected for many years.

The powerful new technique of three-dimensional radiographic examination by tomography can be attractive if the cost and the significant time required are available. Although the technique is clearly capable of identifying serious defects, the additional capability to assess inaccessible wall thickness deep inside complex castings or the presence of residual core material in an inaccessible passageway, are major potential advantages.

19.8.2 DYE PENETRANT INSPECTION

Dye penetrant inspection is a sensitive technique for surface-connected defects such as porosity and cracks. In this process, the casting is bathed or immersed in a dye, a liquid designed to be highly wetting so as to penetrate small holes and fissures rapidly. The castings is withdrawn from the die, washed and coated with a ‘developer’ in the form of a fine white powder that can absorb the dye by capillary attraction as a blotting paper, sucking the die out of its crevice and into the developer layer where it can be seen. The most sensitive dyes fluoresce in ultraviolet light, and are therefore viewed in a darkened area.

The dye penetrant inspection approach is of course limited to the detection of surface-connected defects; any internal defects, which may be major faults, remain undetectable.

19.8.3 LEAK TESTING

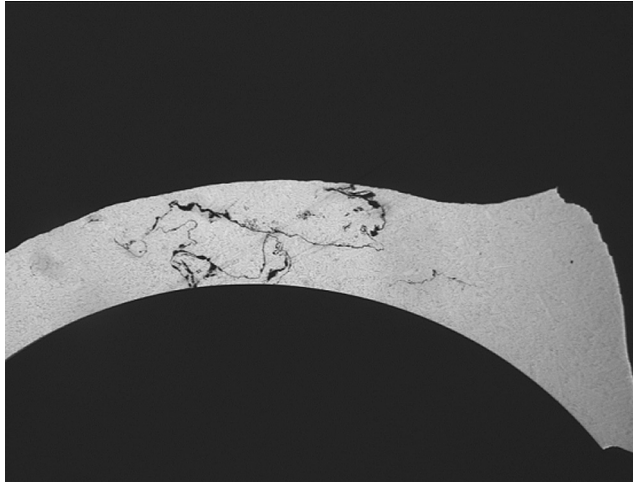
Section 9.11 describes how the leakage of castings is almost never from ‘porosity’ resulting from shrinkage or gas. Section 7.2.2 illustrates how leaks can result from core blows that have effectively punched holes through the walls of the casting, creating serious leaks. In general, however, leaks appear to be mainly the result of the presence of bifilms. [Figure 19.7](#) shows an approximately 20 mm long bifilm in an Al alloy casting which clearly had no problem to bridge the 5 mm wall thickness even though the path was long and meandering.

The finding of such leaks by leak testing is a costly and time-consuming activity which those counter-gravity casting operations that do not make leaky castings have the good fortune to avoid. Interestingly, the leak shown in [Figure 19.7](#) was discovered not by leak testing, but by a resonance test which is far quicker and easier, besides, of course, also discovering any other seriously deleterious features from which the casting may be suffering. This indicates an interesting future direction for the general testing of castings.

However, for those casters that need to conduct a leak test the many techniques are listed in some detail in Section 9.11. By far the commonest system is the time-honoured bubble test by pressurisation of the casting with air while immersing the casting in water. However, the test is easier described than done. Leakages from seals as a result of imperfect dressing of prints can give plenty of additional bubbles that boil to the surface of the water, obscuring any observation of the relatively tiny leaks that the test was designed to find.

Another aspect of the test which gives cause for some concern is the procedure that sets aside any failed casting, for instance an aluminium cylinder head, with a view to repeat the test on the following day. Often, the casting will pass the test on the following day, almost certainly as a result of the leak path being sealed by corrosion products. The rapidity of the corrosive action is likely to result from the association of intermetallic particles that have precipitated on the bifilm, creating efficient corrosion couples. Having reflected for some years on the honesty of this procedure, I find I have never been aware of any failures of these castings in service. Thus, it seems a proven technique for the rescue of expensive and valuable products. Perhaps the procedure would be less acceptable for products that had to withstand high pressures.

Overall, however, the shop floor experience that good quality melts transferred into moulds by appropriate counter-gravity casting operations do not suffer leakage problems is powerful evidence that the leaks are mainly the result of

**FIGURE 19.7**

A leak path formed by an oxide bifilm, causing the master brake cylinder to be rejected.

Courtesy Magnaflux Quasar.

bifilms. The clear message is that these features are avoidable using good filling technology as described in this book, and leak testing could, with relief, be yet another foundry chore consigned to the rubbish bin of history.

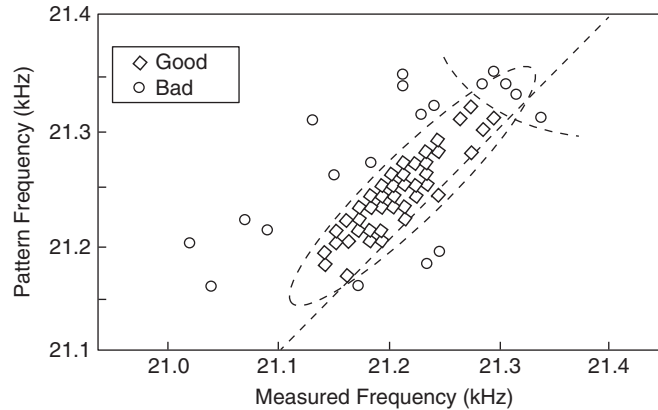
19.8.4 RESONANT FREQUENCY TESTING

Parts have been tested by checking their resonant frequency for many years. From time to time, various forms of the 'ding it and listen' method, sometimes enhanced with electronic equipment, have been tried but have not in general been demonstrated to be sufficiently discriminating for testing the serviceability of castings. The normal variations within castings often changed the 'ringing frequency' more than unacceptable changes in material or structural integrity.

It is worth devoting some space to one particular development of the resonance test that has become available over recent years to illustrate the degree of sophistication that has been necessary to evolve a useful and powerfully discriminating test. The technique is outlined next.

A casting is subjected to a swept spectrum of vibrational frequencies and its response is recorded as a frequency spectrum. Among the thousands of resonance peaks observed in the spectrum, about 50 prominent and recognisable peaks are selected as potentially useful for further analysis. These 50 frequencies are recorded for each casting in a sample 'teaching' set of 100 or more castings, some of which are normal, acceptable castings and some of which have been rejected for various reasons. These samples provide the resonant frequency information to 'teach' the system to accept changes in frequencies that are the result of normal process changes (such as slight changes in size or shape, chemical composition etc.) which do not affect its serviceability. The software selects and analyses about six resonant frequencies to incorporate the possible variations into one comprehensive 'pattern' that can encompass all the acceptable variations.

The presence of the unacceptable castings in the sample set is to ascertain whether the frequency pattern that accepts all the good parts is in fact capable of rejecting all of the castings in which there is a known defect. If defects are present, the frequency pattern of the unacceptable casting will change in ways that are different from those for normal castings. The amount of change is usually a measure of the seriousness of the defect. This concept relies on the quality assurance teaching phase defining as 'acceptable' those parts that are structurally acceptable, as measured by metallurgical, destructive and non-destructive testing. An iterative 'teaching' process is undertaken, so that suitable frequency patterns

**FIGURE 19.8**

A representation of a sophisticated resonant frequency sorting technique, here illustrating a two-dimensional elliptical and a curved boundary, separating good parts from bad. In practice, multi-dimensional criteria are used.

Data courtesy Magnaflux Quasar.

are progressively identified and confirmed by separate testing as useful indicators. Because of the number of castings required for this setting-up phase, the technique is limited to long series production parts, but achieves zero or near-zero false acceptance errors in millions of components. Typical production testing time is a few seconds per casting in a fully automated, computer controlled sorting operation.

A typical distribution chart of testing for leaking master brake cylinders is shown in Figure 19.8. 'Perfect' castings would be expected to fall on the 45° line for which observed and predicted vibration frequency pattern are identical. However, a spread of vibration pattern (1) within the region defined by the ellipse and (2) to the left of a second criterion is found to deliver acceptable products. Frequencies outside these two boundaries, 'the acceptance window', were always found to have some serious fault, although many had been accepted by X-ray and liquid dye penetrant inspection. Figure 19.8 shows only two discriminating frequency boundaries as a simplifying example, whereas in fact the technique always uses a pattern combining about six resonances analysed by sophisticated software. In view of the ability of the technique to allow for the harmless changes caused by slight process variations, it has been called 'process-compensated resonance testing'.

Current developments include the exploration of computer models of the component under test, to ascertain whether the 'teaching' stage, and the exploration of the effects of different kinds and sizes of defects and other changes, may be carried out in the computer.

Castings with structural weakness (previously accepted because the major defects were effectively invisible bifilms that gave little or no classical non-destructive testing indications) are rejected. Conversely, castings that would have been rejected for negative non-destructive testing indications are accepted if they are assessed to be structurally acceptable.

A further interesting capability of the technique, known as resonant ultrasound spectrometry, is the accurate measurement of the elastic (Young's) modulus, as well as other elastic properties such as shear modulus and Poisson's ratio. The elastic moduli are reduced if there is a large population of small bifilms present in the casting. The casting can be rejected if the elastic moduli fall sufficiently to indicate a serious concentration of bifilms that would, for instance, impair the ductility and toughness of the alloy. The ability of the technique to assess ductility and toughness without resort to the expense and delay involved with a destructive tensile test is a further impressive advantage.

The concept provides a quality assurance method that predicts the structural performance of the casting, reduces false reject scrap, and reduces testing costs. This new, fast and low-cost technique appears to out-perform our existing non-destructive testing techniques, and promises to be powerful for future high production and safety-critical products.