CHAPTER

THE 10 RULES FOR GOOD CASTINGS

10

The 10 Rules are my personal checklist (Campbell, 2004), ensuring that I have not forgotten any essential aspect of casting manufacture. It cannot be emphasised too strongly that the failure of only one of the rules can result in total failure of the casting. This is not meant to be alarmist, but simply practical. No one has ever promised that making castings would be easy. However, following the rules is a great help.

We start off with a quick summary, followed by a detailed assessment of each rule in turn in the remainder of this section.

Rule 1. Start with a good quality melt

Immediately before casting, the melt shall be prepared, checked, and treated, if necessary, to bring it into conformance with an acceptable minimum standard. Prepare and use so far as possible only near-defect-free melt.

Rule 2. Avoid turbulent entrainment of the surface film on the liquid

This is the requirement that the liquid metal front (the meniscus) should not go too fast. Maximum meniscus velocity is approximately 0.5 m/s for most liquid metals. This requirement also implies that the liquid metal must not be allowed to fall more than the critical height corresponding to the height of a sessile drop of the liquid metal. The maximum velocity may be raised to 1.0 m/s or even higher, and the critical fall height might be correspondingly raised to approximately 50 mm, in sufficiently constrained running systems or thin section castings.

Rule 3. Avoid laminar entrainment of the surface film on the liquid

This is the requirement that no part of the liquid metal front should come to a stop before the complete filling of the mould cavity. The advancing liquid metal meniscus must be kept “alive” (i.e. moving) and therefore free from thickened surface film that may be incorporated into the casting. This is achieved by the liquid front being designed to expand continuously. In practice, this means progress only *uphill* in a continuous *uninterrupted* upward advance (i.e. in the case of gravity-poured casting processes, from the base of the sprue onwards). This implies

- Only bottom-gating is permissible.
- No falling or sliding downhill of liquid metal is allowed.
- No horizontal flow of significant extent.
- No stopping of the advance of the front due to arrest of pouring or waterfall effects, etc.

Rule 4. Avoid bubble entrainment

No bubbles of air entrained by the filling system should pass through the liquid metal in the mould cavity. This may be achieved by:

- Properly designed off-set step pouring basin; fast back-fill of properly designed sprue; preferred use of stopper; avoidance of the use of wells and all other volume-increasing features of filling systems (such as expanding channels sometimes known as ‘diffusers’); small volume runner and/or use of ceramic filter close to sprue/runner junction; possible use of bubble traps. A naturally pressurised filling system fulfils most of these criteria.
- No interruptions to pouring.

Rule 5. Avoid core blows

- No bubbles from the outgassing of cores or moulds should pass through the liquid metal in the mould cavity. Cores to be demonstrated to be of sufficiently low gas content and/or adequately vented to prevent bubbles from core blows.
- No use of impermeable clay-based core or mould repair paste.

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Rule 6. Avoid shrinkage

- No feeding uphill in larger section thickness castings. Feeding against gravity is unreliable because of (1) adverse pressure gradient and (2) complications introduced by convection.
- Demonstrate good feeding design by following all seven Feeding Rules, by an approved computer solidification model, and by test castings.
- Once good feeding is attained, fix the temperature regime by controlling (1) the level of flash at mould and core joints, (2) mould coat thickness (if any), and (3) temperatures of metal and mould.

Rule 7. Avoid convection

Assess the freezing time in relation to the time for convection to cause damage. Thin- and thick-section castings automatically avoid convection problems. For intermediate sections, either (1) reduce the problem by avoiding convective loops in the geometry of the casting and rigging, (2) avoid feeding uphill, or (3) eliminate convection by rollover after filling.

Rule 8. Reduce segregation

Predict segregation to be within limits of the specification, or agree out-of-specification compositional regions with customer. Avoid channel segregation formation if possible.

Rule 9. Reduce residual stress

No quenching into water (cold or hot) following solution treatment of light alloys. (Polymer quenchant or forced air quench may be acceptable if casting stress can be shown to be acceptable.)

Rule 10. Provide location points

All castings to be provided with location points for pickup for dimensional checking and machining. Proposals are to be agreed on with quality auditor, machinist, etc.

10.1 RULE 1: USE A GOOD-QUALITY MELT

10.1.1 BACKGROUND

The melt needs to be demonstrated to be of good quality. A good-quality liquid metal is one that is defined as follows.

1. Substantially free from suspensions of non-metallic inclusions in general and bifilms in particular.
2. Relative freedom from bifilm-straightening and bifilm-opening agents. These include certain alloy impurities in solution such as Fe in Al alloys and hydrogen or other gases. However, it should be noted that Si can precipitate in its primary form as a bifilm straightening element and is, of course, inescapable when using unmodified Al-Si alloys, but is avoided in modified alloys.

Please note that the good quality of the melt should not be taken for granted, and, without proper treatment, most often fails this requirement.

There are a few exceptions in which good quality might be assumed. Such metals may include pure liquid gold, iridium, platinum, perhaps mercury and possibly some liquid steels whilst in the melting furnace at a late stage of melting. These instances are, however, either rare or tantalisingly inaccessible, such as the steel in the melting furnace; in the process of getting it out of the furnace, much damage is done to the liquid by the pouring that is currently an integral feature of our conventional steel foundries.

An important distinction is useful to identify four major oxide inclusion types as listed in Section 2.9. All of these different oxide films will necessarily have the structure of a bifilm as a result of them becoming submerged, having penetrated the oxidised surface of the liquid metal. Clearly, the major task is to eliminate the macroscopic bifilms. However, for reliable properties, it is also essential to eliminate the majority of the larger fraction of mesoscopic and microscopic bifilms.

Regrettably, many liquid metals are actually so full of sundry solid phases floating about that they begin to more closely resemble slurries than liquids. In the absence of information to the contrary, this condition of a liquid metal should be assumed. Over recent years, the evidence for the real internal structure of liquid metals being crammed with defects has been growing as investigation techniques have improved. Some of this evidence is described later. Much of the evidence applies to aluminium and its alloys where the greatest research effort has been. Evidence for other materials is presented elsewhere in this book.

It is sobering to realise that many of the strength-related properties of metals can only be explained by assuming that the initial melt is full of defects. Many of our theoretical models of liquid metals and solidification that are formulated to explain the occurrence of defects neglect to address this critical fact. Classical physical metallurgy and solidification science has been unable to explain the important properties of cast materials such as the effect of dendrite arm spacing and the ease of formation of pores and cracks. During the early stage of fatigue, the problem of the anomalously rapid growth of short cracks despite the low stress intensity is easily explained if the cracks pre-exist. We shall see that in general the behaviour of cast metals arises naturally from the population of defects.

However, it has to be admitted that is often not easy to confirm the presence of non-metallic inclusions in liquid metals, and even more difficult to quantify their number and average size or spread of sizes. McClain et al. (2001) and Godlewski and Zindel (2001) have drawn attention to the unreliability of the standard approach studying polished sections of castings. A technique for liquid aluminium involves the collection of inclusions by pressurising up to 2 kg of melt, forcing it through a fine filter, as in the porous disc filtration analysis and Prefil tests. Pressure is required because the filter is so fine. The method overcomes the sampling problem by concentrating the inclusions by a factor of about 10,000 times (Enright and Hughes, 1996 and Simard et al., 2001). The layer of inclusions remaining on the filter can be studied on a polished section. (The total quantity of inclusions is assessed as the area of the layer as seen in section under the microscope, divided by the quantity of melt that has passed through the filter. The unit is therefore the curious quantity mm^2kg^{-1}. We can hope that at some future date this unhelpful unit will, by universal agreement, be converted into some more meaningful quantity such as volume of inclusions per volume of melt. In the meantime, the standard provision of the diameter of the filter in reported results would at least allow readers the option to convert the values for themselves.)

To gain some idea of the range of inclusion contents an impressively dirty melt might reach 10 mm^2kg^{-1}, an alloy destined for a commercial extrusion might be in the range 0.1–1, foil stock might reach 0.001 and computer discs 0.0001 mm^2kg^{-1}. For a filter of 30 mm diameter, these figures approximately encompass the range 10^{-3} (0.1%) down to 10^{-7} (0.1 part per million) volume fraction.

Other techniques for the monitoring of inclusions in Al alloy melts in the past included Liquid Metal Cleanness Analyser (LiMCA), in which the melt is drawn through a narrow tube while measuring the voltage drop applied along the length of the tube. The entry of an inclusion of different electrical conductivity (usually non-conducting) into the tube causes the voltage differential to rise by an amount that is assumed to be proportional to the size of the inclusion. The technique is generally thought to be limited to inclusions approximately in the range 10–100 μm, presuming the inclusions to be particles. Although once widely used for the casting of wrought alloys, the author regrets that that technique has to be viewed with great reservation. As we have mentioned, the key inclusions in light alloys are not particles but (double) films, and although often extremely thin, can be up to 10 mm in diameter. Such inclusions sometimes succeed to find their way into the LiMCA tube, where they tend to hang in the metal stream, caught up at the mouth of the tube and rotate into spirals like a flag tied to the mast by only one corner. Asbjornsonn (2001) has reported piles of helical oxides in the bottom of the LiMCA crucible. It is to be regretted that most workers using LiMCA have been unaware of these serious problems. However, the general disquiet about the appropriateness of this technique has finally caused the LiMCA device to be dropped by the manufacturer. It seems unlikely to be used significantly in the future.

Ultrasonic reflections have been used from time to time to investigate the quality of melts. The early work by Mountford and Calvert (1959) is admirable, making recommended reading, and has been followed up by considerable development efforts in Al alloys (Mansfield, 1984), Ni alloys and steels (Mountford et al., 1992). Ultrasound is efficiently reflected from oxide films because the films are double, and the elastic wave cannot cross the intermediate layer of air and thus is reflected with mirror-like efficiency. However, the reflections may not give an accurate idea of the size of the defects because the irregular, crumpled form of such defects, and their tumbling action in the melt. The tiny mirror-like facets of large defects reflect back to the source only when the facets happen to rotate to face the beam. The result is a general scintillation effect, apparently from many minute and separate particles. It is not easy to discern whether the images correspond to many small or a few large bifilms.

Neither LiMCA nor the various ultrasonic probes can distinguish any information on the types of inclusions that they detect. In contrast, the inclusions collected by pressurised filtration can be studied in some detail. In aluminium alloys many different inclusions can be found. Table 1.1 lists some of the principal types.

Nearly all of these foreign materials will be deleterious to products intended for such products as thin foil or computer discs. However, for engineering castings, those inclusions such as carbides and borides are probably not harmful at all (although they would be unpopular with machinists). This is because, having been precipitated from elements in solution in the melt, they would be expected to be in excellent atomic contact with the matrix alloy. These non-metallic phases enjoy well-bonded interfaces and are thereby unable to act as void-type initiators or volume defects such as pores and cracks. On the contrary, they may act as grain refiners. Furthermore, their continued good bonding with the solid matrix is expected to confer on them a minor or negligible deleterious influence on mechanical properties. They may strengthen the matrix to some degree.

Generally, therefore, this book concentrates on those inclusions that have a major detrimental influence on mechanical properties, because of their action to initiate other serious problems such as pores, cracks and localised corrosion. Thus the attention will centre on *entrained surface films, known as bifilms*. Usually, these inclusions will be oxides. However, carbon films are also common, and occasionally nitrides, sulphides and other substances. These entrained films exhibit a unique structure, with outer faces that are in atomic contact with the melt, but have an inner interface that is unbonded, and thus lead to the spectrum of problems of pores and cracks all too familiar to foundry people.

The pressurised filtration tests can find some of these entrained solids, and the analysis of the inclusions present on the filter can help to identify the source of many inclusions in a melting and casting operation. However, films that are newly entrained into the melt as a result of surface turbulence remain undetectable but can be enormously important. These films are commonly entrained during the pouring of castings, and so, perhaps, are not normally subjected to detection or assessment in a melting and distribution operation at this late stage. They are typically only 20 nm thick, and so remain invisible under an optical microscope, especially in a ceramic filter because they will be difficult to discern when draped around a piece of the filter that, when sectioned, appears many thousands of times thicker.

The only fairly reliable detection technique for such inclusions is the lowly reduced pressure test (RPT). This test opens the films (because they are always double and contain residual entrapped air or other gases such as hydrogen) so that they can be seen by eye on a polished section or on a radiograph. The radiography of the cast test pieces reveals the size, shape and numbers of such important inclusions, as has been shown in studies by Fox and Campbell (2000) and Dispinar and Campbell (2006). A central slice of the small cylindrical RPT casting to yield a parallel section gives an improved radiographic result (Figure 10.1). Viewing this work in retrospect the approach using X-rays now appears somewhat over the top, and probably unnecessary. Simply viewing a polished section by unaided eye has subsequently proved almost equally effective, and vastly cheaper and quicker.

Figure 10.2 shows how a quality map can be constructed to show the total length and total number of bifilms on a polished RPT section. The large central square might form a suitable quality window for an operation making 'space filling' components for which filling and feeding are not critical as a result of relatively low requirements from the casting. The '50 × 50' regime might be a minimum requirement for parts requiring some strength and toughness. The Cosworth foundry used a window '3 × 3' during my time and quite often achieved '0 × 0', giving castings of exceptional soundness and high properties.

However, even the RPT technique probably only reveals the more extensive bifilms, between perhaps 0.1 and 10 mm diameter. This is because the effective tensile opening stress applied to the bifilm can only be a maximum of 1 atm (0.1 MPa). In a tensile test of a solidified metal, the stress can easily reach 1000 times greater than this and so is capable of opening much smaller bifilms. For instance, in some Al-Si alloys cracked Si particles at the base of many ductile dimples indicate the failure of the particle by cracking. If we assume this failure can only occur because of the presence of a bifilm, then a true density of defects might approach 10^{15} m^{-3} (corresponding to 10^{6} mm^{-3}) at a diameter of between 1 and 10 μm. This is a huge density of defects, but is consistent with the direct ultrasonic observation of a melt by Mountford and Calvert (1959), in which they describe a dirty melt appearing as a fog. After the inclusions were allowed to settle (aided by precipitating heavy grain-refining titanium-rich compounds on them), the melt cleared

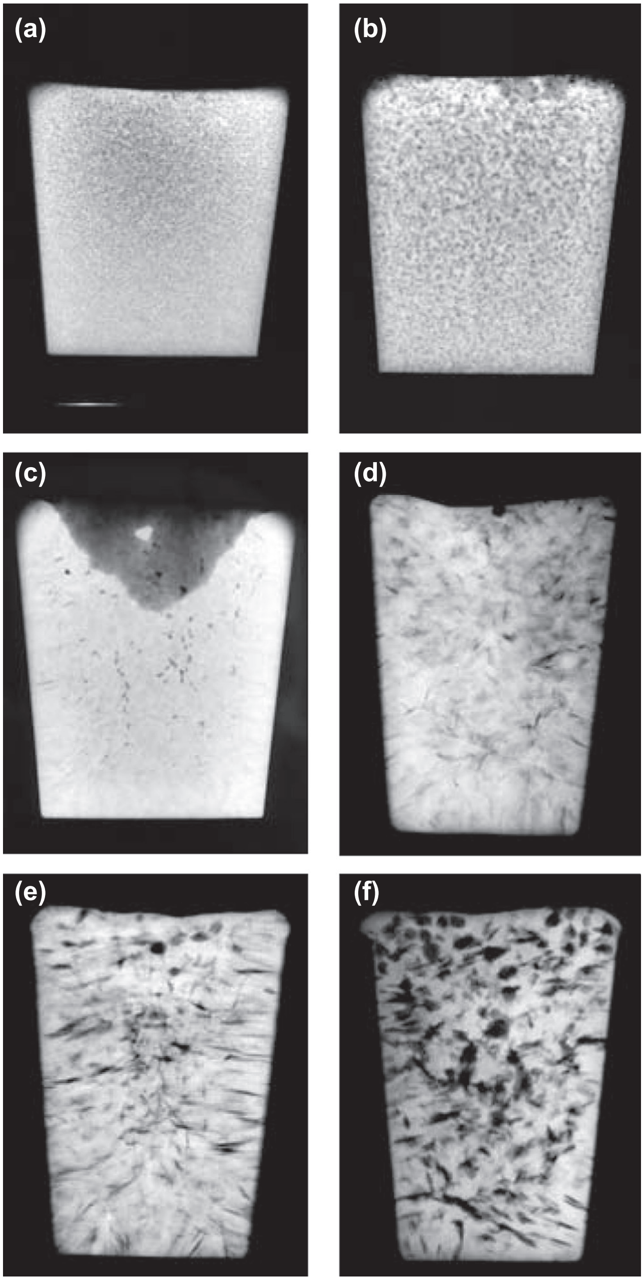

FIGURE 10.1

Radiographs of RPT samples of Al-7Si-0.4Mg alloy illustrating different bifilm populations. (For description of each part (a to f) see text).

Courtesy S. Fox.

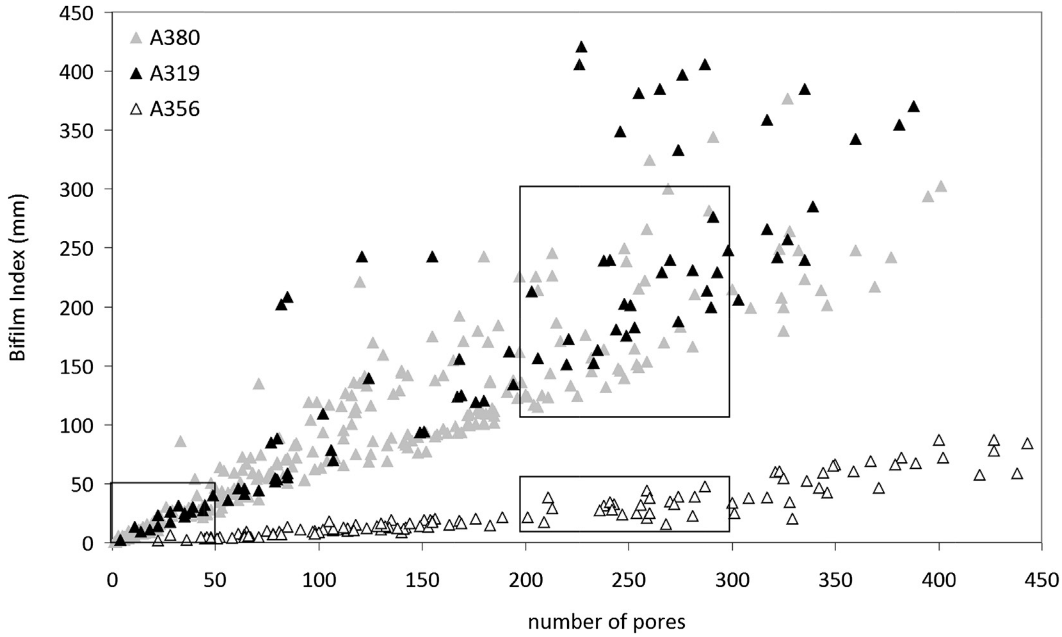

FIGURE 10.2

RPT results for three casting alloys, showing number and total length of bifilms on the polished section (Dispinar and Campbell, 2011).

completely so that a clear back-wall echo could be seen. They were able to repeat this phenomenon simply by stirring up the melt to recreate the fog, and watch the weighed-down oxides clearing once again to build up a layer of bottom sediment.

It is unfortunate that many melts start life with poor, sometimes grossly poor, quality in terms of its content of suspended bifilms. The 'fog' persists in most alloys because the bifilms are usually neutrally buoyant. Figure 10.1 gives several examples of different poor qualities of liquid aluminium alloy. The figures show results from RPT samples observed (somewhat unnecessarily as has already been noted!) by X-ray radiography. Because the samples are solidified under only one tenth of an atmosphere (76 mm residual pressure compared with the 760 mm of full atmospheric pressure), any gas-containing defects, such as bubbles, or bifilms with air occluded in the centres of their sandwich structures, will be expanded by 10 times. Thus rather small defects can become visible for the first time.

We shall assume that pores are always initiated by bifilms, giving initially crack-like or irregular pores. The formation of rounded pores simply occurs as a result of the bifilm being opened beyond this initial condition by excess precipitation of gas, finally achieving a pore diameter greater than the original dimensions of the bifilm. Thus the RPT is an admirably simple device for assessing (1) the number of bifilms; (2) their average size (even though this might be somewhat of an overestimate if much hydrogen is present); and (3) gas content is assessed by the degree of lowering the average density of the cast sample by the opening of the bifilms from thin crack-like forms to fairly spherical pores.

If the melt contained no gas-containing defects, the cut and polished section (or in this case the radiographs) of the RPT would be clear.

However, as we can see immediately, and without any benefit of complex or expensive equipment, the melts recorded in Figure 10.1 are far from this desirable condition. Figure 10.1(a) shows a melt with small rounded pores that indicates that the bifilms that initiated these defects were particularly small, of the order of 0.1 mm or less. The density of these defects, however, was high, between 10 and 100 defects per cubic cm. Figure 10.1(b) has a similar defect distribution, but with slightly higher hydrogen content. Figure 10.1(c) illustrates a melt that displayed a deep shrinkage pipe, normally interpreted to mean good quality, but showing that it contained a scattering of larger pores, probably as a result of fewer bifilms opened mainly by shrinkage rather than by gas. Figure 10.1(d) has considerably larger bifilms, of size in the region of 5 mm in length, and in a concentration of approximately 1/cm^3. Figure 10.1(e) and (f) show similar samples but with increasing gas contents that have inflated these larger bifilms to reasonably equiaxed pores.

Naturally, it would be of little use for the casting engineer to go to great lengths to adopt the best designs of filling and feeding systems if the original melt was so poor that a good casting could not be made from it.

Thus this section deals with some of the aspects of obtaining a good-quality melt.

In some circumstances, it may not be necessary to reduce both bifilms and bifilm-opening agents. An interesting possibility for future specifications for aluminium alloy castings (where residual gas in supersaturated solution does not appear to be harmful) is that a double requirement may be made consisting of (1) the content of dissolved gas in the melt to be high, but (2) the percentage of gas porosity to be low. The meeting of this interesting double requirement will ensure to the customer that bifilms are not present. Thus these damaging but undetectable defects will, if present, be effectively labelled and made visible on X-ray radiographs and polished sections by the precipitation of dissolved gas. It is appreciated that such a stringent specification might be viewed with dismay by present suppliers. However, at the present time we have mainly only rather poor technology, making such quality levels out of reach. We shall not necessarily suffer such backward processing for ever.

The possible future production of Al alloys for aerospace, with high hydrogen content but low porosity, is a fascinating challenge. As our technology improves, such castings may be found not only to be manufacturable, but offer a guaranteed reliability of fatigue life, and therefore command a premium price.

The prospect of producing ultra-clean Al alloys that can be *demonstrated* in this way to be actually extremely clean, raises the issue of contamination of the liquid alloy from the normal metallurgical additions such as the various master alloys, and grain refiners, modifiers, etc. It may be that for superclean material, normal metallurgical additions to achieve refinement of various kinds will be found unnecessary, and possibly even counter-productive because of the unavoidable contamination by primary oxide skins from the outer surfaces of the master alloy additions.

The improvement of Al alloys by melting with dry hearth furnaces to eliminate the primary oxide skins of the charge, simply by scraping these off the hearth through a side door, is to be strongly recommended. The subsequent treatment of the melt with Ti-rich grain refiners is also recommended, not necessarily to grain refine, but to precipitate heavy Ti-rich inclusions on oxide bifilms to sediment these from the melt. It seems likely that extremely clean Al might be obtained in this way. These and other sedimentation techniques are discussed in more detail under the section on melting.

For steels, the content of hydrogen may be a more serious matter, especially if the section thickness of the casting is large. In some steel castings of section thickness above about 100 mm up to 1000 mm or more, the hydrogen cannot escape by diffusion during the time available for cooling or during the time of any subsequent heat treatments. Thus the high hydrogen content retained in these heavy sections may lead to hydrogen embrittlement and catastrophic failure of the section by cracking. Although Ren et al. (2008) claim that hydrogen might nucleate voids with the aid of vacancies, it is difficult to avoid what appears to be the more probable speculation that the hydrogen cracking might initiate by hydrogen precipitating into, and forcing open, bifilms. To avert hydrogen cracking problems when making very large steel castings, such as backup rolls for rolling mills, it was traditional practice to induce a carbon boil to reduce the hydrogen content but now is reduced by modern, rather costly techniques such as treatment by argon oxygen degassing. Having done this, it is necessary to work quickly to cast the metal because the hydrogen content tends to rise rapidly once again, re-establishing its equilibrium with its environment such as damp atmospheres and/or damp refractories in ladles etc.

For ductile iron production, the massive amounts of turbulence that accompanies the addition of magnesium in some form, such as magnesium ferro-silicon, are almost certainly highly damaging to the liquid metal. It is expected that immediately after such nodularisation treatment, the melt will be massively dirty. It will be useful therefore to ensure that the melt can dwell for sufficient time for the entrained magnesium oxide–rich bifilms to float out. The situation is analogous to the treatment of cast iron with CaSi to effect inoculation (i.e. to achieve a uniform distribution of graphite of desirable form). In this case the volumes of calcium-oxide-rich films are well known, so that the CaSi treatment is known as a 'dirty' treatment compared to FeSi inoculation. The author is unsure about in-mould treatments therefore with such oxidisable elements as Ca and Mg. Clearly these treatments spherodise the graphite as they are designed to do, but it is not clear whether the matrix suffers from additional bifilms which would lower properties? Work to clarify this question would be valuable.

For nickel-based superalloys melted and cast in vacuum, it is with regret that the material is, despite its apparently clean melting environment, found to be sometimes as crammed with oxides (and/or nitrides) as an aluminium alloy (Rashid and Campbell, 2004). This is because the main alloying element in such alloys is aluminium, and the high temperature favours rapid formation and thickening of the surface film on the liquid. This problem occurs even when casting in vacuum, because the vacuum is, of course, only dilute air. If the film becomes entrained, the liquid and any subsequent casting is damaged. The process for the production of the alloys designed for remelting to make castings in foundries involves, unfortunately, melting and pouring processes that leave much to be desired:

The melting and alloying process in the manufacture of Ni-based alloys starts in an induction crucible furnace, and is poured under gravity via a series of sloping launders, falling several times, and finally falling 1–2 m or more into steel tubes that act as moulds. This awfully turbulent primary production process for the alloy impairs all the downstream cast products. More recent moves to increase productivity using larger diameter tubes to make larger diameter alloy billets have only made the bad situation worse. An alternative recent move towards the production of Ni-base alloy bar by horizontal continuous casting is to be welcomed as the first step towards a more appropriate production technique for these key ingredients of our modern aircraft turbines. Production of improved material is currently limited, but deserves to be the subject of demand from customers.

Even so, the subsequent melting and gravity casting operations in the investment casting foundries to produce Ni- and Co-base turbine blade castings are also currently extremely disappointing, so that good quality starting material would at this time probably be a waste. The aircraft industry has been trapped within its rigid procedures designed to ensure safety, but which have inhibited the rational engineering development of improved casting techniques that could deliver far safer products at reduced costs.

Melting systems and melt treatments designed to provide improved melt quality are dealt with in more detail in Volume 3 'Melting'.

10.2 RULE 2: AVOID TURBULENT ENTRAINMENT (THE CRITICAL VELOCITY REQUIREMENT)

10.2.1 INTRODUCTION

The avoidance of surface turbulence is probably the most complex and difficult rule to fulfil when dealing with gravity pouring systems.

The requirement is all the more difficult to appreciate by many in the industry because everyone working in this field has always emphasised the importance of working with 'turbulence-free' filling systems for castings. Unfortunately, despite all the worthy intentions, all the textbooks, all the systems and all the talk, so far as the author can discover, it seems that no one appears to have achieved this target so far. In fact, in travelling around the casting industry, it is quite clear that the majority (at least 80%) of all defects are directly caused by turbulence. Thus the problem is massive; far more serious than suspected by most of us in the industry.

In fact, Johnson and Baker concluded from their experiments in 1948 that 'gating systems did not function as commonly supposed' and 'no gating system prevented turbulence'. It seems we have all been warned for a long time.

To understand the fundamental root of the problem, it is clear in Section 2.1 that any fall greater than the height of the sessile drop (of the order of 10 mm) causes the metal to exceed its critical velocity, and so introduces the danger of defects in the casting. Because most falls are in fact at least 10 or 100 times greater than this and because the damage is likely to be proportional to the energy involved (i.e. proportional to the square of the velocity), the damage so created will usually be expected to be 100–10,000 times greater. Thus in the great majority of castings that are poured simply under the influence of gravity, there is a major problem to ensure its integrity. In fact, the situation is so bad that the best outcome of many of the filling system design solutions proposed in this book are merely damage limitation exercises. Effectively, it has to be admitted that at this time it seem impossible to guarantee the avoidance of some damage when pouring liquid metals.

This somewhat depressing conclusion needs to be tempered by several factors.

1. Expectations. The world has come to accept castings as they are. Thus any improvement will be welcome. This book described techniques that will create very encouraging improvements.
2. Continuing development. This book is merely a summary of what has been discovered so far in the development of filling system design. Better designs are to be expected now that the design parameters and filling system concepts (such as critical velocity, critical fall height, entrainment, bifilms, etc.) are defined and understood.
3. There *are* filling systems that can yield, in principle, perfect results.

Of necessity, such perfection is achieved by fulfilling rule 2 by avoiding the transfer of the melt by pouring. Thus consider the three modes of filling a mould:

1. Downhill pouring under gravity;
2. Horizontal transfer into the mould achieved by (a) tilt casting in which the tilt conditions are accurately controlled to achieve horizontal transfer, or (b) by the 'level pour' or 'level transfer' techniques;
3. Uphill (counter-gravity) casting in which the melt is never poured, but caused to fill the mould in only an uphill mode.

Only the last two processes have the potential to deliver castings of near perfect quality. In my experience, I have found that in practice it is often difficult to make a good casting by gravity, whereas by a good counter-gravity process (i.e. a process observing all the 10 rules) it has been difficult to make a bad casting. The jury is still out on horizontal transfer by tilting. This approach has great potential, but is applicable only to certain limited shapes of castings, and requires a dedicated effort to achieve the correct conditions. Horizontal transfer has great potential but unfortunately has fallen into disuse.

Thus in summary, filling of moulds can be carried out down, along, or up. Only the 'along' and 'up' modes totally fulfil the non-surface-turbulence condition.

However, despite all its problems, it seems more than likely that the *downward* mode, *gravity casting*, will continue to be with us for the foreseeable future. Thus in this book we shall devote some considerable length to the damage limitation exercises that can offer considerably improved products, even if, unfortunately, those products cannot be ultimately claimed as perfect.

Most will shed no tears over this conclusion. Although potential perfection in the *along* and *up* modes is attractive, the casting business is all about making *adequate* products; products that meet a specification and at a price a buyer can afford.

The question of cost is interesting; perhaps the most interesting. Of course the costs have to be right, and often gravity casting is acceptable and sufficiently economical. However, more often than might be expected, high quality and low cost can go together. An improved gravity system, and certainly one of the better counter-gravity systems, can be surprisingly economical and productive. Such opportunities are often overlooked. It is useful to be alert to such benefits.

10.2.2 MAXIMUM VELOCITY REQUIREMENT

Some years ago, I was seated in the X-ray radiograph viewing room using an illuminated viewing screen to study a series of radiographic films of cylinder head castings made by our recently developed counter-gravity casting system at the Cosworth foundry. Each radiograph in turn was beautiful, having a clear, 'wine glass' perfection of which every founder dreams. I was at peace with the world. However, suddenly, a radiograph appeared on the screen that was a total disaster.

It had gas bubbles, shrinkage porosity, hot tears, cracks and sand inclusions. I was shocked, but sensed immediately what had happened. I shot out of the viewing room into the foundry to query Trevor, our man on the casting station. 'What happened to this casting?' He admitted instantly, 'Sorry governor, I put the metal in too fast'.

This was a lesson that remained with me for years. This chance experiment by counter-gravity, using an electromagnetic pump allowing independent control of the ingate velocity, had kept all the other casting variables constant (including temperature, metal quality, alloy content, mould geometry, mould aggregate type, binder type etc.), showing them to be of negligible importance. Clearly, the ingate velocity was dominant. By only changing the speed of entry of metal we could move from total success to total failure. This was a fundamental lesson. Clearly, it applies to all casting processes.

Thinking further about this lesson, common sense tells us all that there is an optimum velocity at which a liquid metal should enter a mould. The concept is outlined in Figure 2.18. At a velocity of zero, the melt is particularly safe (Figure 2.18(a)), being free from any danger of damage. Regrettably, this stationary condition for the melt is not helpful for the filling of moulds. In contrast, at extremely high velocities the melt will enter the mould like a jet of water from a fire-fighter's hosepipe (Figure 2.18(c)), and is clearly damaging to both metal and mould. At a certain intermediate velocity, the melt rises to just that height that can be supported by surface tension around the periphery of the spreading drop (Figure 2.18(b)). The theoretical background to these concepts is dealt with at length in Chapter 2. For nearly all liquid metals, this critical velocity is close to 0.5 ms^{-1}. This value is of central importance in the casting of liquid metals, and will be referred to repeatedly in this section and when designing filling systems for castings.

Japanese workers optimising the filling of their design of vertical stroke pressure die-casting machine using both experiment and computer simulation (Itamura et al., 1995) confirm the critical velocity of 0.5 ms^{-1} for their Al alloy, finding at velocities above this value that air bubbles have a chance to become entrained. (These workers go further to define the amount of liquid that needs to be in the mould cavity above the ingate to suppress entrainment at higher ingate velocities.)

Looking a little more closely at the detail of critical velocities for different liquids, it is close to 0.4 ms^{-1} for dense alloys such as irons, steels and bronzes and about 0.5 ms^{-1} for liquid aluminium alloys. The value is 0.55–0.6 ms^{-1} for Mg and its alloys. Taking an average of about 0.5 ms^{-1} for all liquid metals is usually good enough for most purposes related to the design of filling systems for castings, and will be generally used throughout this book.

Returning to Figure 2.18(b) showing the liquid metal emerging close to its critical velocity, and spreading slowly from the ingate. The shape of the drop is closely in equilibrium, its surface tension holding its compact shape, and just balancing the head of pressure due to its density that would tend to spread the drop out infinitely thin. As the ingate steadily supplies metal at close to 0.5 ms^{-1} the steadily growing drop is closely resembles the shape of a *sessile drop* (from the Latin word for '*sitting*'. The word contrasts with *glissile drop*, meaning a gliding or sliding drop). A sessile drop of Al sitting on a non-wetted substrate is always approximately 12.5 mm high. Corresponding values for other liquids are Fe 10 mm, Cu 8 mm, Zn 7 mm, Pb 4 and water 5 mm.

Recent research has demonstrated that if the liquid velocity exceeds the critical velocity, there is a danger that melt will overshoot the height supportable by surface tension, so that on falling back again the surface of the liquid metal may be folded over. This *entrainment* of the liquid surface I have called *surface turbulence*. At risk of overly repeating this important phenomenon, this entrainment of the surface can occur if there is sufficient energy in the form of velocity in the bulk liquid to perturb the surface against the smoothing action of surface tension. In addition, notice that damage is not *necessarily* created by the falling back of the metal. The falling is likely to be chaotic, so that any folding action may or may not occur. The significance of the critical velocity is clear therefore: below the critical velocity the melt is safe from entrainment problems; above the critical velocity there is the danger, but not the necessity, of surface entrainment leading to defect creation.

To be more precise about the entrainment action, actually any disturbance of the surface of the liquid irreversibly extends the area of oxide on the surface, with the result that some entrainment of the additional oxide area is unavoidable. This is because the surface oxide film forms almost instantly, but once formed, cannot reduce its area without crumpling in some way, leading to the folding in of the excess surface oxide (see, for instance, Figure 2.2(b)). Thus even below the critical velocity some entrainment may occur, but, clearly, above the critical velocity the rate of entrainment suddenly increases.

Thus we see there is a chance that if the speed of the liquid exceeds about 0.5 ms^{-1}, its surface film may be folded into the bulk of the liquid. This folding action is an *entrainment* event. It leads to a variety of problems in the liquid that we can collectively call *entrainment defects*. The major entrainment defects are.

1. Bubbles as, of course, air bubbles (but widely misinterpreted as reaction products of solidification such as hydrogen porosity, or reaction products between oxide slag and graphite in cast iron to create CO bubbles).
2. Bifilms, as doubled-over oxide films. The author has named these folded-in films '*bifilms*' to emphasise their double, folded-over nature. Because the films are necessarily folded dry side to dry side, there is little or no bonding between their dry interfaces, so that the double films act as cracks. The cracks (alias bifilms) become frozen into the casting, lowering the strength and fatigue resistance of the metal. Bifilms may also create leak paths, causing leakage failures, and provide the ingress of corrodants leading to pitting and other varieties of corrosion.

The folding-in of the oxide is a random process, leading to scatter and unreliability in the properties on a casting-to-casting, day-to-day and month-to-month basis during a production run.

The different qualities of metals arriving in the foundry will also be expected to contain populations of bifilms that will differ and type and quantity from batch to batch. Thus the performance of the foundry will suffer additional variation. The foundry needs to have procedures in place to smooth variations of its incoming raw material so far as possible so as to fulfil rule 1.

The maximum velocity condition effectively forbids top gating of castings (i.e. the planting of a gate in the top of the mould cavity, causing the metal to fall freely inside the mould cavity). This is because liquid aluminium reaches its critical velocity of about 0.5 ms^{-1} after falling only 12.5 mm under gravity. The critical velocity of liquid iron or steel is exceeded after a fall of only about 10 mm. (These are, of course, the heights of the sessile drops.) Naturally, such short fall distances are always exceeded in practice in top gated castings, leading to the danger of the incorporation of the bifilms, and consequent porosity, leakage and crack defects.

Castings that are made in which velocities everywhere in the mould never exceed the critical velocity are consistently strong, with high fatigue resistance, and are leak-tight (we shall assume they are properly fed, of course, so as to be free from shrinkage porosity).

Experiments on the casting of aluminium have demonstrated that the strength of castings may be reduced by as much as 90% or more if the critical velocity is exceeded (Figure 2.19). The corresponding defects in the castings are not always detected by conventional non-destructive testing such as X-ray radiography or dye penetrant because, despite their large area, the folded oxide films are sometimes extremely thin, and do not necessarily give rise to any significant surface or X-ray indications.

The speed requirement automatically excludes conventional pressure die-castings as having significant potential for reliability because the filling speeds are usually 10–100 times greater than the critical velocity. Even so, over recent years there have been welcome moves, introducing some special developments of high-pressure technology that are capable of meeting this requirement. These include the vertical injection squeeze casting machine, and the shot control techniques. Such techniques can, in principle, be operated to fill the cavity through large gates at low speeds, and without the ingress of air into the liquid metal. Such castings require to be sawn, rather than broken, from their filling systems of course. Unfortunately, the castings remain somewhat impaired by the action of pouring into the shot sleeve. Even here, these problems are now being addressed by some manufacturers, with consequent benefits to the integrity of the castings, but involving technical, maintenance and cost challenges that are not easily overcome.

Other uphill filling techniques such as *low-pressure* filling systems are capable of meeting rule 2. Even so, it is regrettable that the critical velocity is practically always exceeded during the filling of the low-pressure furnace itself because of the severe fall of the metal as it is transferred into the pressure vessel, damaging the melt prior to casting. In addition, many low-pressure die casting machines are in fact so poorly controlled on flow rate that the speed of entry into the die can greatly exceed the critical velocity, thus negating one of the most important potential benefits of the low-pressure system. Processes such as Cosworth Process avoid these problems by never allowing the melt to fall at any stage of processing, and control its upward speed into the mould by electromagnetic pump.

These processes are to be discussed in greater detail in the chapter 16 ‘Casting’. The conventional rammed lining in such low-pressure furnaces is a further source of major defects as a result of the bubbles which emerge from the lining each time the furnace is depressurised. This serious but little-known problem is dealt with in detail in Section 16.31.

Metals that can also suffer from entrained surface films include the zinc-aluminium alloys and ductile irons. For a few materials, particularly alloys based on the Cu-10Al types (aluminium and manganese bronzes), the critical velocities were originally thought to be much lower, in the region of only 0.075 ms^{-1}. However, from recent work at Birmingham (Halvaee and Campbell, 1997), this low velocity seems to have been a mistake probably resulting from the confusion caused by bubbles entrained in the early part of the filling system. With well-designed filling systems, the aluminium bronzes accurately fulfil the theoretically predicted 0.4 ms^{-1} value for a critical ingate velocity.

Some ferrous alloys have similarly dry oxide films that give classical, unbonded bifilm defects, although in certain irons and steels the entrained bifilms agglomerate as a result of being partially molten and therefore somewhat sticky. These sticky masses of oxide therefore remain more compact and float out more easily to form surface imperfections in the form of slag macroinclusions on the surface. The subject of entrainment is described more extensively in the metallurgy of cast steels (Section 6.6).

Because of the central importance of the concept of critical velocity, the reader will forgive a re-statement of some aspects in this summary.

1. Even if the melt does jump higher than the height of a sessile drop, when it falls back into the surface there is no certainty that it will enfold its surface film. These tumbling motions in the liquid can be chaotic, random events. Sometimes the surface will fold badly and sometimes not at all. This is the character of surface turbulence; it is not predictable in detail. The key aspect of the critical velocity is that at velocities less than the critical velocity the surface is always safe. Above the critical velocity, there is always the *danger* of entrainment damage. The critical velocity criterion is therefore seen to be a necessary but not sufficient condition for entrainment damage.
2. If the whole, extensive surface of a liquid were moving upwards at a uniform speed, but exceeding the critical velocity, clearly no entrainment would occur. Thus the surface disturbance that can lead to entrainment should more accurately described not merely as a velocity but in reality a velocity difference. It might therefore be defined more accurately as a critical velocity gradient measured across the liquid surface. For those of a theoretical bent, the critical gradient might be defined as the velocity difference achieving the critical velocity along a distance in the surface of the order of the sessile drop radius (approximately half its height) in the liquid surface. To achieve reasonable accuracy, this approach requires allowing for the reduction in drop height with velocity. Hirt (2003) solves this problem with a delightful and novel approach, modelling the surface disturbances as arrays of turbulent eddies, and achieves convincing solutions for the simulation of entrainment at hydraulic jumps and plunging jets. Such niceties are neglected here. The problem does not arise when considering the velocity of the melt when emerging from a vertical ingate into a mould cavity. In that situation, the ingate velocity and its relation to the critical velocity is clear.
3. If the melt is travelling at a high speed, but is constrained between narrowly enclosing walls, it does not have the room to fold-over its advancing meniscus. Thus no damage is suffered by the liquid despite its high speed and despite the high risk involved. This is one of the basic reasons underlying the design of extremely narrow channels for filling systems that are proposed in this book. Very narrow filling channels have the great advantage of filling the running system in *one pass*, the melt filling the channel with its meniscus operating as a piston to push the air ahead. In this way, despite its high speed, it retains its quality. (This contrasts with oversize channels in which there is sufficient room for the melt to jet and splash, rebounding from the end of the channels to flow counter to the incoming melt, churning with the air. The high shearing speeds fragment the bubbles and oxides into numerous relatively small defects prior to entering the mould cavity. However, on those occasions when the melt does enter the mould cavity too quickly, any resulting damage tends to be even more serious. This is because such defects are then formed in the casting itself, and at the somewhat lower speeds than in the runner, so tending to avoid fragmentation, resulting in relatively few but massive defects.)

10.2.3 THE NO-FALL REQUIREMENT

It is quickly shown that if liquid aluminium is allowed to fall more than 12.5 mm, then it exceeds the critical speed 0.5 m/s. The critical fall height can be seen to be a kind of re-statement of the critical velocity condition. Similar critical velocities and critical fall heights can be defined for other liquid metals. The critical fall heights for all liquid metals are in the 3–15 mm range .

It follows immediately that practically all pouring of metals is bad. Figure 10.3 shows an all too common and lamentably efficient destruction of the melt quality. Clearly, neither the employee, manager nor designer of the plant knew they were engaged in such a destruction of quality. In addition of course, this destruction of metal simply forms oxide in the form of dross. The fall shown in Figure 10.3 would probably have caused the loss of 2–3% of metal; a large and avoidable cost.

It also follows similarly that *top gating* of moulds, almost without exception, will lead to a violation of the critical velocity requirement. In addition, for many forms of gating that enter the mould cavity at the mould joint, if any significant part of the cavity is below the joint, these will also violate this requirement.

In fact, for conventional sand and gravity die casting, it has to be accepted that some fall of the metal is necessary. Thus it has been accepted that the best option is for a single fall, concentrating the total fall of the liquid at the very beginning of the filling system, and causing the fall to be surrounded by a close-fitting, tapered conduit known as a *sprue*, or *down-runner*. This conduit brings the melt to the lowest point of the mould. The damage potential of the fall

FIGURE 10.3

A typical melt transfer, seriously degrading the melt.

is limited if the sprue is as narrow as possible, so as to constrain the melt, giving it little chance to damage itself by folding in its oxide surface. Thus the sprue is designed to have a cross-section only just large enough to take the required flow rate to fill the casting in the target fill time. The distribution system from the point that the melt hits the base of the sprue, consisting of runners and gates, should progress only *horizontally* or preferably *uphill*. The metal should never suffer a further drop.

Considering the mould cavity itself, the no-fall requirement effectively rules that all gates into the mould cavity enter at the bottom level, known as *bottom-gating*. The siting of gates into the mould cavity at the top *(top-gating)* or at the joint *(gating at the joint line)* are not options if safety from surface turbulence is required.

Also excluded are any filling methods that cause waterfall effects in the mould cavity (Figure 2.30). This requirement dictates the siting of a separate ingate at every isolated low point on the casting.

Even so, the concept of the critical fall distance does require some qualification. If the critical limit is exceeded, it does not mean that defects will necessarily occur. It simply means that there is a risk that they may occur. This is because for falls greater than the critical fall distance the energy of the liquid is sufficiently high that the melt is potentially able to enfold in its own surface. Whether a defect occurs or not is now a matter of chance.

There is, however, further qualification that needs to be applied to the critical fall distance. This is because the critical value quoted previously has been worked out for a liquid, neglecting the presence of any oxide film. In practice, it seems that for some liquid alloys, the surface oxide has a certain amount of strength and rigidity, so that the falling stream is contained in its oxide tube and so is enabled to better resist the conditions that might enfold its surface. This behaviour has been investigated for aluminium alloys (Din et al., 2003). It seems that although the original fall distance limit of 12.5 mm continues to be the safest option, fall heights of up to about 100 mm might be allowable in some instances, corresponding to a speed of 1.4 m/s. In the unusual circumstances of a ceramic foam filter being inserted into the flow, the fall distance downstream of the filter might be increased to approximately 200 mm without undue damage to the melt as explained in Section 11.2. However, falls greater than 200 mm definitely entrain defects; the velocity of the melt in this case is over 2 ms^{-1}. Entrainment seems unavoidable at this and higher speeds. Also, of course, other alloys may not enjoy the benefits of the support of a tube of oxide around the falling jet. This benefit requires to be investigated in other alloy systems to test what values beyond the theoretical limits may be used in practice.

The initial fall down the sprue in gravity-filled systems does necessarily introduce some oxide damage into the metal. The damage is concentrated during the first few seconds during which the sprue is priming. This period of fall of the metal can be very messy. The period can be shortened by contact pouring or by the use of a stopper in an offset step-pouring basin to ensure the basin is filled before the sprue being allowed to fill. Even so, it seems reasonable to conclude that gravity poured castings will never attain the degree of reliability that can be provided by counter-gravity and other systems that can totally avoid surface turbulence.

Of necessity therefore, it has to be accepted that the no-fall requirement applies to the design of the filling system beyond the base of the sprue. The damage encountered in the fall down the sprue has to be accepted; although with a good sprue, good contact pouring or good pouring basin and stopper this initial fall damage can be reduced to a minimum.

It is a matter of good luck that it seems that for some alloys much of the oxide introduced by the initial turbulence in the filling system does not appear to find its way through and into the mould cavity. It seems that much of it remains attached to the walls of the running system. This fortunate effect is clearly seen in many top-gated castings, where most of the oxide damage (and particularly any random leakage problem) is confined to the area of the casting under the point of pouring, where the metal is falling, as is seen in Figure 9.45. Severe damage does not seem to extend into those regions of the casting where the speed of the metal front decreases, and where the front travels uphill, but there does appear to be some carryover of defects. Thus the provision of a filter immediately after the completion of the fall is valuable. It is to be noted however that the filter will not take out all of the damage.

The requirement that the filling system should cause the melt to progress only uphill after the base of the sprue appears to force the decision that the runner must be in the drag and the gates must be in the cope for a horizontal single jointed mould. However, it has to be admitted that this preferred arrangement is not always possible, and in any case the velocities dictated by the acceleration from gravity ensures that the melt takes little notice of such niceties; the metal simply overrides this small difference in height.

The no-fall requirement may also exclude some of those filling methods in which the metal slides down a face inside the mould cavity, such as some tilt-casting type operations. This undesirable effect is discussed in more detail in the section devoted to tilt casting. As we shall see, sliding downhill is not a necessary filling mode for tilt casting; if feasible, horizontal transfer of the melt is strongly recommended for the tilt-casting process as explained in Section 16.2.2.

It is noteworthy that these precautions to avoid the entrainment of oxide films also apply to casting in inert gas or even in vacuum. This is because the oxides of Al and Mg (as in Al alloys, ductile irons and high temperature Ni-base alloys for instance) form so readily that they effectively ‘getter’ the residual oxygen in any conventional industrial vacuum, and form strong oxide films on the surface of the liquid.

Rule 2 in its no-fall embodiment applies to ‘normal’ castings with walls of thickness over 3–4 mm. For channels that are sufficiently narrow, having dimensions approaching a millimetre, the curvature of the meniscus at the liquid front should keep the liquid front from disintegration, the effect of surface tension becoming ever more powerful. Thus the action of surface tension in narrow filling system geometries is valuable to conserve the liquid as a coherent mass, and so acting to push the air out of the system ahead of the liquid. The filling systems therefore fill in one pass. (Unfortunately, in some conditions, this highly desirable mode of filling degenerates into micro-jetting as mentioned again later. Possibly this damaging form of flow applies only to certain alloys or certain environmental conditions as mentioned in Section 2.2.7.)

A good filling action, pushing the air ahead of the liquid front as a piston in a cylinder, is a critically valuable action. Such systems deserve a special name such as perhaps ‘*One pass filling designs*’. Although I do not usually care for such jargon, the special name emphasises the special action. It contrasts with the turbulent and scattered filling often observed in systems that are over-generously designed, in which the melt can be travelling in two directions at once along a single channel; a fast jet travels under the return wave that rolls over its top, rolling in air and oxides at the boundary between the opposing streams.

For a wide, shallow horizontal channel, any effect of surface tension is clearly limited to channels that have dimensions smaller than the sessile drop height for that alloy. Thus for Al alloys, the maximum channel height would be 12.5 mm, although even this height would exert little influence on the melt because the roof would just touch the liquid, exerting no pressure on it. Similarly, taking account that the effect of surface tension is doubled if the curvature of the liquid front is doubled by a second component of the curvature at right angles, a channel of square section could be 25 mm square, and be just contained by surface tension. In practice, however, for any useful restraint from the walls of channels, these dimensions require to be at least halved, effectively compressing the meniscus, as a mechanical spring, into the tight confines of the channel.

For very thin-walled castings, of section thickness less than 2 mm, the effect of surface tension in controlling filling becomes dominant. The walls are so much closer than the natural curvature of a sessile drop that the meniscus is tightly compressed and requires the application of pressure to force it into such narrow gaps. The liquid surface is now so constrained that it is not easy to break the surface i.e. once again there is no longer room for splashing or droplet formation. Thus the critical velocity is higher, and metal speeds can be raised by approximately a factor of two without danger.

In very thin-walled castings, with walls less than 2 mm thickness, the tight curvature of the meniscus becomes so important that filling can sometimes be without regard to gravity (i.e. can be uphill or downhill) because the effect of gravity is overpowered by surface tension. This makes even the uphill filling of such thin sections problematic because the effective surface tension exceeds the effect of gravity. Instabilities therefore occur, whereby the moving parts of the meniscus continue to move ahead in spite of gravity because of the reduced thickness of the oxide skin at that point. Conversely, other parts of the meniscus that drag back are further suppressed in their advance by the thickening oxide, so that a runaway instability condition occurs. This dendritic advance of the liquid front is no longer controlled by gravity in very thin castings, making the filling of extensive sections, whether horizontal or vertical, a major filling problem.

The problem of the filling of thin walls occurs because the flow happens, by chance, to avoid filling some areas because of random meandering. Such chance avoidance, if prolonged, leads to the development of strong oxide films or even freezing of the liquid front. Thus the final advance of the liquid to fill such regions is hindered or prevented altogether.

The dangers of a random filling pattern problem are relieved by the presence of regularly spaced *ribs* or other geometrical features that assist to organise the distribution of liquid. Random meandering is thereby discouraged and replaced by regular and frequent penetration of the area, so that the liquid front has a better chance to remain 'live' i.e. it keeps moving so that a thick restraining oxide is given less chance to form.

The further complicating effect of the microscopic break-up of the front known as micro-jetting observed in casting sections of 2 mm and less in sand and plaster moulds is not yet understood. The effect has not yet been systematically researched, and may not occur at all in the dry mould conditions such as are found in gravity die (permanent mould) casting. What little is known about this mysterious phenomenon is discussed briefly in Chapter 2.

10.3 RULE 3: AVOID LAMINAR ENTRAINMENT OF THE SURFACE FILM (THE NON-STOPPING, NON-REVERSING CONDITION)

10.3.1 CONTINUOUS EXPANSION OF THE MENISCUS

If the liquid metal front continues to *advance* at all points on its surface, effectively, continuing to *expand* at all points on its surface, like a progressively inflating balloon, then all will be well. This is the ideal mode of advance of the liquid front. In general, this is the mode of advance of a counter-gravity filling system, which is why counter-gravity filling of castings is such a powerful technique for the production of good castings.

In fact, we can go further with this interesting concept of the requirement for continuous surface expansion. There is a sense in that if surface is lost (i.e. if any part of the surface experiences contraction) then, because the surface oxide cannot contract, some entrainment of the surface oxide necessarily occurs. Thus this can be seen to be an all-embracing and powerful definition of the condition for entrainment of defects, simply that surface area must not be lost. Clearly, surface area is effectively lost by being *enfolded* (in the sense that the fold now disappears inside the liquid, as has been the central issue described under rule 2), or by simply *shrinking* i.e. being lost (necessarily leading to folding) as described later. Thus, in a way, this condition 'avoid loss of surface' can be seen to supersede the conditions of critical velocity or Weber number. It promises to be a useful condition that could be recognised in numerical simulation and thus be useful for computer prediction of entrainment.

In practice, however, an uphill advance of the liquid front, if it can be arranged in a mould cavity, is usually a great help to keep the liquid front as 'alive' as possible, i.e. keeping the meniscus moving, and so expanding and continuously creating new oxide.

It seems possible that the condition for exceeding the critical velocity may lead to more severe entrainment than simple surface contractions, and that surface contractions might lead to entrained film that remains closely attached to the oxide on the surface, instead of becoming displaced into the deep interior of the casting as happens with surface turbulence. These distinctions, if they are real, remain to be clarified by research.

Although the surface is being continuously expanded when filling the mould uphill, the casting has the benefit that the older, thicker oxide is continuously being displaced to the walls of the mould where it becomes the skin of the casting. Thus very old and very thick oxide does not normally have chance to form and become entrained. In fact, one of the great benefits of a good filling system is to ensure that the older oxides on the surface of the ladle or pouring basin etc. do not enter the mould cavity. For instance, in the tilt pouring of aluminium alloy castings, the filling of a new casting can be checked by dropping a fragment of paper on the metal surface as the tilt commences. This marker should stay in place, indicating that the old skin on the metal was being retained by the runner, so that only clean metal underneath could flow into the mould cavity. If the paper disappears into the runner, the runner is not doing its job. In a way, the use of the teapot pouring ladle and bottom-pour ladles common in the steel casting industry are in response to the acute form of these same problems, in which the high rates of reaction with the environment at such high temperatures encourages the surface oxide to grow from microscopic to macroscopic thickness, to constitute the familiar slag layer.

10.3.2 ARREST OF FORWARD MOTION OF THE MENISCUS

Problems arise if the front becomes pinned by the rate of advance of the metal front being too slow, or if it stops or reverses.

If the liquid front stops, a thick surface film has chance to form. Figure 10.4(a) shows a common way in which this can happen—by the abrupt enlargement of area at some stage during the filling of the mould. The stationary surface film may become so thick and strong that if pressure increases later to encourage flow to re-start, the front finds itself pinned in place, no longer able to move. As the advance re-starts elsewhere in the mould, the melt may overflow and submerge the thick stationary film, forming an entrainment defect (Figure 2.25). As the new metal rolls over the old film, a new fresh oxide is laid down over the old thick oxide, forming our familiar double film, with dry side facing dry side so that the double layer forms an unbonded interface, as a crack. On this occasion, the bifilm is a thick/thin asymmetric variety. This mode of creation of a bifilm can constitute a large geometrical defect, sometimes in the form of a horizontal crack extending across the whole casting, or as a horizontal lap around its complete perimeter.

A good name for this defect is an *oxide lap*. (It is to be distinguished from the solidification defect that often has a superficially similar appearance, called here a *cold lap* caused not necessarily by the strength of the oxide film but by the freezing in place of the liquid front. The method of treating these defects is quite different. A cold lap can be cured by increasing the casting temperature, whereas increasing the temperature of an oxide lap is likely to make it worse.)

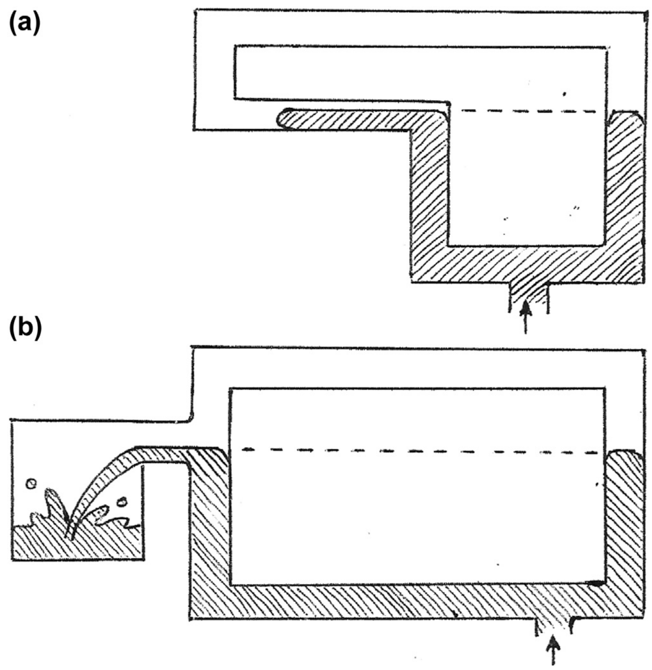

FIGURE 10.4

Two common filling situations in which the general advance of the melt is stopped, introducing the danger of a lap defect. (a) An enlarged area, and (b) a ‘waterfall’ effect to halt the advance of the liquid front.

10.3.3 WATERFALL FLOW: THE OXIDE FLOW TUBE

Instead of a large horizontal defect, a curious but major geometrical defect in the form of a cylindrical tube that I call an *oxide flow tube* can form in several ways.

If the liquid falls vertically, as a plunging jet, the falling stream is surrounded by a tube of oxide (Figures 2.30 and 10.4(b)). Despite the high velocity of the falling metal inside, the oxide tube remains stationary, thickening with time, until finally surrounded by the rising level of the metal in the mould cavity. This rising metal rolls up against the oxide tube, forming a double oxide crack (once again of a highly asymmetrical thick/thin form). Notice the curious cylindrical form of this crack and its largely vertical orientation. The arrest of the advance of the front in this case occurred by the curious phenomenon that although the metal was travelling at a high speed parallel to the jet, its transverse velocity, i.e. its velocity at right angles to its surface, was zero. It is the zero velocity component of a front that allows the opportunity for a thick oxide skin to develop.

These oxide flow tubes are often seen around the falling streams of many liquid metals and alloys as they are poured. The defects are also commonly seen in castings. Although occasionally located deep inside the casting where they are not easily found, they are often clearly visible if formed against the casting surface.

A patent dating from 1928 (Beck) describes how liquid magnesium can be transferred from a ladle into a mould by arranging for the pouring lip of the ladle to be as close as possible to the pouring cup of the mould, and to be in a relatively fixed position so that the semi-rigid oxide flow tube which forms automatically around the jet is maintained unbroken, thus protecting the metal from contact with the air (Figure 2.23(a)). A similar phenomenon is seen in the pouring of aluminium alloys and other metals such as aluminium bronze.

A large Ni-base superalloy turbine blade, cast in vacuum, and destined for a land-based turbine, was placed upside down and top-poured through its root. The root had almost certainly created a cluster of vertically oriented oxide flow tubes as the melt poured in a series of separate streams through the root, to start its long fall towards the blade tip. When the transverse grooves of the fir tree root were machined by grinding, vertical cracks were observed which were argued to be grinding cracks from over-enthusiastic grinding. However, their origin as oxide flow tubes was corroborated by the observation under the microscope of a series of sulphide precipitates with an alignment that was characteristic of occupying one side of a vertical oxide film. Micrographs could not be released for security reasons, so a sketch to illustrate the effect is seen in Figure 10.5(a). The interpretation of the micrograph is shown in Figure 10.5(b). The presence of the bifilm in the form of an oxide flow tube explains both the vertical cracking

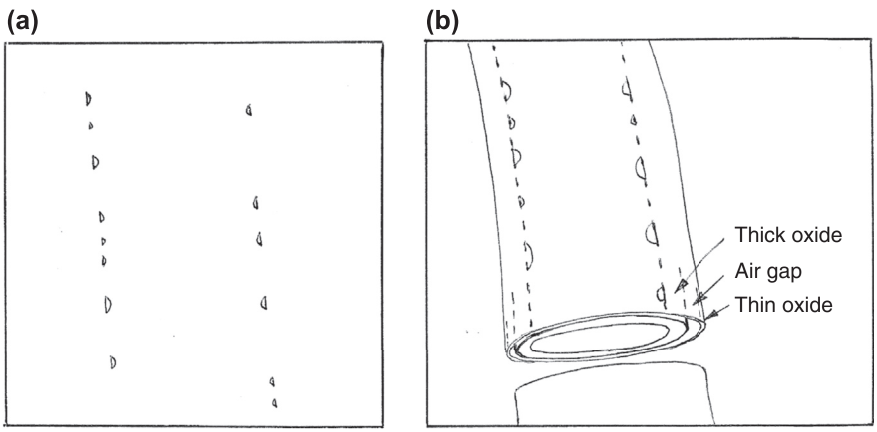

FIGURE 10.5

(a) Vertical lines of sulphide inclusions seen on a polished section of a vacuum top-poured Ni-base superalloy.
(b) Interpretation in terms of precipitation on an oxide flow tube (oxide thicknesses and air gap are greatly exaggerated for clarity).

during grinding, and the vertical alignment of the sulphides which have clearly grown only on one side of an interface, which in this case appears to be the inner wetted interface of the original flow tube. Precipitates have been reported to form on only once side of an asymmetric bifilm, as seen in a confluence weld (Garcia-Garcia et al., 2007). (The air gap between the two oxide films shown in Figure 10.5(b) is exaggerated for clarity. Clearly in practice the two films will be in contact, and the air gap will be comprised only of the air trapped in the microscopic roughness of the two contacting surfaces.)

The stream does not need to fall vertically. Streams can be seen that have slid-down gradients in such processes as tilt casting when carried out under poor control. Part of the associated flow tube is often visible on the surface of the casting, as a witness to the original presence of the metal stream. Alternatively, a wandering horizontal stream can define the flow tube, as is commonly seen in the spread of liquid across a horizontal surface. Figure 2.24 shows how, in a thin horizontal section, the banks of the flowing stream remain stationary whilst the melt continues to flow. When the flow finally fills the section, coming to rest against the now-rigid oxide forming the banks of the stream, the banks will constitute long meandering asymmetrical bifilms as cracks, following the original line of the flow.

The jets of flow in pressure die castings can sometimes be seen to be leave permanent legacies as oxide tubes as seen in Figure 2.31.

Even with the best design of gravity-poured system, the rate of fill of the mould may be far from optimum at certain stages during the fill. For instance, Figure 10.4 shows two common geometrical features in castings that cause the advance of the liquid metal to come to a stop. The heavy section filled downhill will cause the metal front to stop, possibly causing a lap-type defect at points on the casting well away from the real cause of the problem. Until the recess is filled by the pouring metal, the remainder of the liquid front cannot advance. There are several reasons for avoiding any ‘waterfall’ action of the metal during the filling of the mould.

1. A cylindrical oxide flow tube forms around the falling stream. If the fall is from a reasonable height, the tube is shed from time to time, and plunges into the melt where it will certainly contribute to severe random defects. The periodic shedding of oxide flow tubes into the melt is a common sight during the pouring of castings. Several square metres of oxide area can be clearly seen to be introduced in this way during the filling of the mould.
2. The plunging jet is likely to exceed the critical velocity. Thus the metal that has suffered the fall is likely to be impaired by the addition of randomly entrained bifilms generated near the point of impact.
3. As the melt rises around the tube, supporting it to some degree and reducing the height of the fall, the flow tube remains in place and simply thickens. As the general level of the melt rises around the tube, the new oxide rolls up against the surface of the cylinder, forming the curious cylindrical bifilm that acts as a major cylindrical crack around a substantially vertical axis.
4. During the period of the waterfall action, the general rise of the metal in the rest of the mould will be interrupted, causing an oxide to form across the whole of the stationary level surface. If this has time to thicken significantly, it may be too strong to move when the levels equalise. Thus rising melt will overflow the film, as the flooding over an ice sheet, rolling a second thin film into place on top of the thick stationary layer. Thus a major horizontal lap defect may be created as a horizontal bifilm. The bifilm will be highly asymmetrical, with one thick and one thin film facing each other, as described earlier.

Defects (1) and (2) are the usual fragmented and chaotic type of bifilm. Defects from (3) and (4) are the major geometrical bifilms.

Waterfall problems are usually easily avoided by the provision of a gate into the mould cavity at every low point in the cavity. Occasionally, deep recesses can be linked by channels through the mould or core assembly; the links being removed during the subsequent dressing of the casting.

10.3.4 HORIZONTAL STREAM FLOW

If the melt is allowed to spread without constraint across a horizontal thin-walled surface, gravity can play no part in persuading the flow to propagate on a broad stable front, as would happen naturally in a vertical or sloping plate (Figure 2.24). The front propagates unstably in the form of a river bounded by river banks composed of thickening oxide.

A fascinating example of a flow tube can be quoted from observations of flow up a sloping open channel driven by a travelling magnetic field from a linear motor sited under the channel. When used to drive liquid aluminium alloy uphill, out of a furnace and into a higher-level receiver, the travelling melt is seen to flow inside its oxide tube. When the magnetic field is switched off, the melt drains out of the oxide tube and back into the furnace, with the tube collapsing flat on the bottom of the channel. However, when the field is switched on again, the same oxide tube magically refills and continues to pass metal as before. Clearly, the tube has considerable strength and resilience. It is sobering to think that such features can be built into our castings, but remain unsuspected and almost certainly undetectable. Clearly, the casting methods engineer requires vigilance to ensure that such defects will not form.

The vertical oxide flow tube is probably more common than any of us suspect. The example given later is simply one of many that could be described.

Figure 10.6 illustrates the bronze bell, sometimes known as the 'Freedom Bell' as a larger-sized replica of the Liberty Bell, hung outside the railway station in Washington, DC. Horizontal weld repairs, etched for enhanced clarity by rain water, record for all time the fatal hesitations in the pouring process that led to the horizontal oxide bifilms that would have appeared as horizontal laps. Perhaps these were the points at which the pouring ladles were changed. The vertical weld repairs record the passage of the falling stream that created the vertical flow tube that led to cracks through the thickness of the casting. These vertical cracks were opened by the hoop stress created by the casting contracting onto its core. This is a common source of failure for bells, nearly all of which are top-poured through the crown.

The renowned Liberty Bell (the only survivor of three attempts, all of which cracked) reveals a magnificent example of a flow tube defect that starts at the crown, curves sinuously around and over the shoulder and finally falls vertically down the skirt to the mouth of the bell. Although there are many examples of bells that exhibit these long cracks, it is perhaps all the more surprising that any bells survive the top-pouring process. It seems likely that in the majority of cases of bells of thicker section the oxide flow tube is not trapped between the walls of the mould to create a through-thickness pair of parallel cracks. In such thicker sections, the tube is more likely to be detached and carried away, crumpling into a somewhat smaller defect that can be accommodated elsewhere. It is to be hoped that the new resting place of the defect will not pose any serious future threat to the product. Clearly, the top pouring of castings is a risky manufacturing technique.

FIGURE 10.6

Inspection of the Freedom Bell outside the railway station in Washington, DC, unfortunately spoiled by welds from an attempt to repair cracks caused by vertical oxide flow tubes from top pouring and horizontal oxide bifilms from filling hesitations.

Photograph by Sheila.

Oxide flow tubes are common defects seen in a wide variety of castings that have been filled across horizontal sections or down sloping downhill sections. The deleterious oxide flow tube structures described previously that form when filling *downwards* or *horizontally* cannot form when filling *vertically upwards* i.e. in a *counter-gravity* mode. The requirement for reliable castings that the meniscus only travels uphill is sacred.

Although moulds can be filled substantially without risk only by *counter-gravity filling,* even in this case vigilance is still required to avoid other filling defects. Even in this most favourable mode of filling, a related *oxide lap* defect, or even a *cold lap* defect, can still occur if the advance of the meniscus is stopped at any time as we have seen.

In all cases, it will be noticed that such interruptions to flow, where, for any reason, the surface of the liquid locally stops its advance, a large, asymmetric double film defect is created. These defects are always large, and always have a recognisable, predictable geometrical form (i.e. they are cylinders, planes, meandering streams etc.). They are quite different to the double films formed by surface turbulence, which are random in size and shape, often extremely thin or of variable thickness, but essentially completely unpredictable as a result of their chaotic origin.

10.3.5 HESITATION AND REVERSAL

If the meniscus stops at any time, it is common for it to undergo a slight reversal. Minor reversals to the front occur for a variety of reasons.

1. A reversal will practically always occur when a waterfall is initiated. This occurs because at the point of overflow, the liquid will be at a level slightly above the overflow, dictated by the curvature of its meniscus, i.e. for a liquid Al alloy it will be about 12.5 mm above the height of the overflow because this is the height of the sessile drop. However, immediately after the overflow starts, the general liquid level drops, no longer supported by the surface tension of the meniscus. In the case of liquid aluminium alloy this fall in general level of the liquid will be perhaps about 6 mm, just enough to flatten the more distant parts of the meniscus against the rest of the mould walls.
2. Hesitations to an advancing flow will often be accompanied by slight reversals because of inertial effects of the flow. Momentum perturbations during filling will cause slight gravity waves, the surface therefore experiencing minor slopping and surging motion, oscillating gently up and down.

These minute reversals of flow flatten the oxidised surface of the meniscus. When advancing, the meniscus adopts a rounded form, but when flattened, the oxidised surface now occupies a smaller area. A fold necessarily develops, wrinkling the surface, endangering the melt with the possibility that the entrainment of this excess oxide is permanent; the entrainment folding action is not reversed once the melt is able to continue its advance. The folding in of a small crack attached to the surface of the casting is illustrated in Figure 10.7. Such shallow surface cracks occurring as a result of hesitation and/or reversal of the front are common in aluminium alloys, and are revealed by dye penetrant testing.

It is instructive to estimate the maximum depth that such oxide folds might have. In Figure 10.7, if the front of the liquid in Figure 10.7(a) is a cylinder of radius r, the perimeter of the quarter of a cylinder is $\pi r/2$, so that the maximum length of excess surface if the melt level now drops a distance r to become horizontal is approximately $\pi r/2 - r = r((\pi/2) - 1) = r/2$. The radius of the meniscus r is approximately 6 mm for liquid aluminium (as a result of the total height of a sessile drop being approximately 12 mm), giving the excess length 3 mm. If this is folded just once to create a bifilm, its potential depth is therefore found to be approximately 1.5 mm. This may not seem a lot, but an appearance of such a crack during the NDT examination of a casting by dye penetrant will almost certainly fail any safety critical casting. Clearly, the depth of the crack cannot easily be known during the dye penetrant inspection, and in any case might genuinely be serious for a highly stressed part experiencing fatigue.

If the melt continues its downward oscillation, the defect can be straightened out as shown in Figure 10.7(c). Alternatively, if the bifilm created in this way holds itself closed as in Figure 10.7(b), possibly because of viscous adhesion (i.e. the trapped liquid metal takes time to escape from between the films) or possibly as the result of other forces such as Van der Waals forces, then there is the danger that the fold is not reversible, and additional folds may be created on each oscillation cycle.

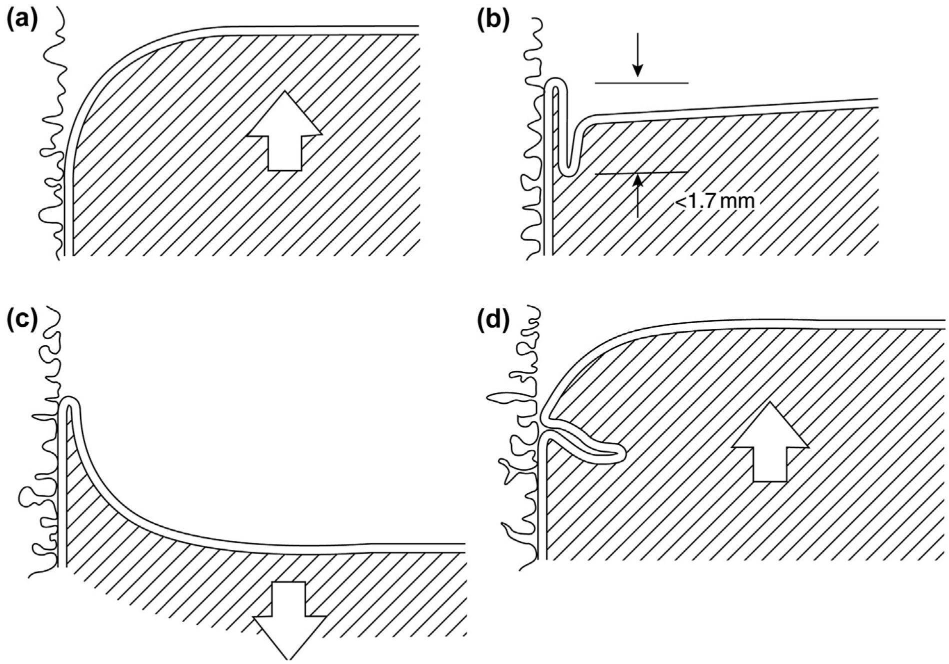

FIGURE 10.7

The creation of bifilm cracks of the order of millimetres deep by the reversal of the front, causing the meniscus to flatten and enfold the excess surface area of film. If (a) the advancing front suffers (b) a small reversal, or (c) a somewhat larger reversal, then (d) the re-starting of the flow may generate a bifilm crack from the enfolding of the excess (stretched) area of oxide film.

In fact, many of these defects are not as deep as the maximum estimate of 1.5 mm for several reasons. (1) The melt surface may not drop the full distance r; (2) the film may be folded more than once, creating a greater number of shallower folds; and (3) the fold-like crack may hinge to lie flat against the surface of the casting. The action of internal forces as a result of flow of the liquid, surging through the mould sections, may be helpful in this respect. For these reasons, such defects are usually only a fraction of a millimetre deep, so that they can often be removed by grit blasting. Only relatively rarely do they reach the maximum possible depth approaching 1.5 mm. Even so, for castings requiring total integrity, requiring resistance to high stress or fatigue, these minor oscillations of the front are very real threats that are best avoided.

The ultimate solution, as we have emphasised here, is that the melt should be designed to be kept on the move, advancing steadily forwards at all times. All we foundry people should be deeply grateful for the simplicity of this solution.

10.3.6 OXIDE LAP DEFECTS

The flooding of the melt over a large oxide film that has grown as a result of a hesitation in the vertical rise of the liquid, and the problem of joining of flows if one has stopped, are essentially equivalent. The first is usually a horizontal double film, where the second is usually vertical. Both are highly asymmetrical, consisting of a thick film (grown on the stationary interface) and a thin film (grown on the moving front).

These very deleterious defects, as major cracks, can be avoided by increasing the rate of filling of the mould. Care is needed of course to avoid casting at too high a rate at which surface turbulence may become an issue. However, providentially, there is usually a comfortably wide operational window in which the fill rate can meet all the requirements to avoid defects. This is another simple solution for which we are all grateful.

10.4 RULE 4: AVOID BUBBLE DAMAGE

Entrainment defects are caused by the folding action of the (oxidised) liquid surface or by the impact of splashes or droplets, all of which meet dry side to dry side. Sometimes only oxides are entrained, as doubled-over film defects, called bifilms. Sometimes the bifilms themselves contain small pockets of accidentally enfolded air, so that the bifilm is decorated by arrays of trapped bubbles. Much, if not all, of the microporosity observed in castings either is, or has originated from, a bifilm.

Sometimes, however, the folded-in packet of air is so large that the buoyancy of the newly created bubble confers on it a life of its own. This oxide-wrapped piece of air that we call a bubble is a massive entrainment defect. It can be sufficiently buoyant to power its way through the liquid and sometimes through the dendrites. In this way it develops its own distinctive damage pattern in the casting.

10.4.1 THE BUBBLE TRAIL

The passage of a single bubble through an oxidisable melt is likely to result in the creation of a bubble trail as a double oxide crack, a kind of long bifilm, in the liquid (Figure 2.32). A bubble trail is the name coined (JC, 1991) to describe the defect that was predicted to remain in a film-forming alloy after the passage of a bubble through the melt. Thus, even though the bubble may be lost by bursting at the liquid surface, the trail remains as permanent damage in the casting.

The bubble trail occurs because the trail is nearly always attached to the point where the bubble was first entrained in the liquid. The enclosing shroud of oxide film covering the crown of the bubble attempts to hinder its motion. This restrain is so large that bubbles having relatively little buoyancy, of diameter less than about 5 mm, are held back, unable to rise. However, if the bubble is larger, its upward buoyancy force will split this restraining cover. Immediately, of course, the oxide on the crown re-forms, and splits and re-forms repeatedly. The expanding region of oxide film on the crown effectively slides around the surface of the bubble, continuing to expand until the equator of the bubble is reached. At this point the area of the film is a maximum. Figure 10.8 shows residual fragments on such a bubble in a Ti alloy (also shown in this figure is some shrinkage porosity that has grown from the underside of the bubble presumably because the casting had been poorly fed. The bubble trail is just visible below this region). Although the film was able to expand by splitting and re-forming, it is not able to contract because it is a solid. Further progress of the bubble causes the film to continue sliding around the bubble, gathering together in a mass of longitudinal pleats under the bubble, forming a trail that leads back to the point at which the bubble was first entrained as a packet of gas. Any spiralling motion of the bubble will twist and additionally tighten this rope-like tether that continues to lengthen as the bubble rises.

The structure of the trail is a kind of collapsed tube. It is star-like in section but with a central portion that has resisted complete collapse because of the small residual rigidity of the oxide film (Figure 2.32). This is expected to form an excellent leak path if it joins opposing surfaces of the casting or if cut into by machining. In addition, of course, the coming together of the opposite skins of the bubble during the formation of the trail ensure that the films make contact dry side to dry side, and so constitute our familiar classical bifilm crack.

Poor designs of filling systems can result in the entrainment of much air into the liquid stream during its travel through the filling basin, during its fall down the sprue and during its journey along the runner. In this way, dozens or even hundreds of bubbles can be introduced into the mould cavity.

10.4.2 BUBBLE DAMAGE

Bubble damage is usually a mixture of the residue of bubbles and bubble trails. It is a typical entrainment mess.

When many bubbles are involved, the later bubbles have problems to rise through the morass of bubble trails that remain after the passage of the first bubbles. Figure 10.9 shows counts of the number of bubbles per cubic centimetre

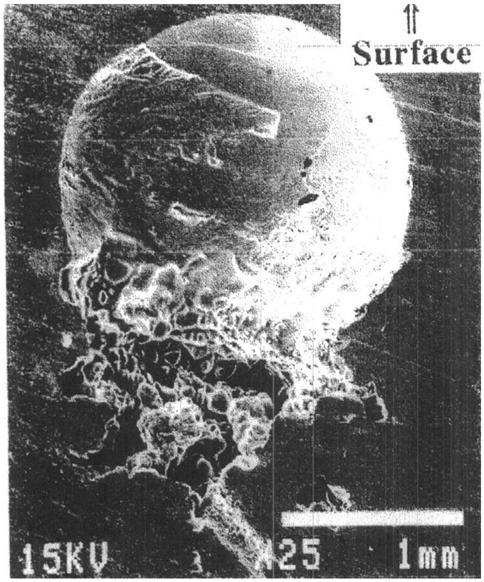

FIGURE 10.8

A bubble in a centrifugally cast Ti-6Al-4V alloy by Suzuki et al. (1996), showing oxide slipping around its surface, and some shrinkage porosity clustered around the start of its bubble trail in the bottom of the image.

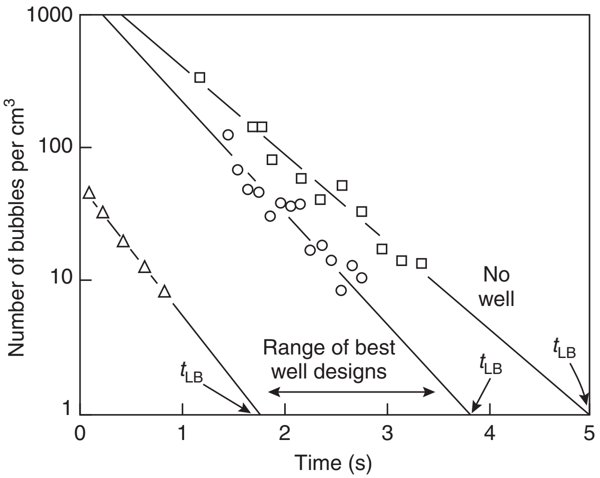

FIGURE 10.9

Water model of bubbles entrained by surface turbulence in a well (Isawa and Campbell, 1994) for runners twice the area of the sprue, showing their decrease with time for different well designs, extrapolated to the time for the last bubble t_{LB}.

arising in a small experimental casting as a result of the provision of a ‘well’ at the base of the sprue. Clearly, hundreds of bubbles are generated, and only die away after several seconds. Larger castings would be expected to generate many times this number, and experience bubble production persisting for proportionally longer. Furthermore, if this were not bad enough, even more bubbles are created in the pouring basin (requiring the offset step basin to keep many of these out of the filling system) and many more are generated at the base of the sprue and in the running system unless these are designed to be naturally pressurised.

It is easy to understand that the free flotation and escape of late-arriving bubbles is hampered by the accumulation of the tangle of residual bifilms. If the density of films is sufficiently great, fragments of bubbles remain entrapped as permanent features of the casting. This mass of fragments of bifilm trails and bubbles is collectively called ‘bubble damage’ (Figure 2.38). In the experience of the author, bubble damage is probably the most common defect in castings, but up to now has been almost universally unrecognised.

Bubble damage in the form of masses of trails is nearly always mistaken for shrinkage porosity as a result of the irregular form of the air porosity that it encloses, usually with characteristic cusp-like morphology. Another common feature of bubble damage is its non-uniform distribution. Unlike shrinkage porosity, it is often not in areas where shrinkage porosity might be expected. For instance, shrinkage porosity would be expected in hot spots, but oxide damage resembling shrinkage porosity can be found almost anywhere in the casting.

In some stainless steels, the phenomenon is seen under the microscope as a mixture of bubbles and cracks. (A remarkable combination! Without the concept of the bifilm, such a combination would be extremely difficult to explain; with bubbles that would naturally expand to relieve and eliminate stresses, and cracks that therefore could not form because their formation would need stresses.) In these strong materials, the high cooling strain leads to high stresses that open up the double-oxide bubble trails.

10.4.3 BUBBLE DAMAGE IN GRAVITY FILLING SYSTEMS

In gravity-filled running systems, the requirement to reduce bubbles in the liquid stream during the filling of the casting calls for offset stepped basins or another non-entraining filling system such as contact pouring, for instance. (The conventional conical or funnel-shaped pouring basin is the worst possible design of filling device; it concentrates at least 50%, but sometimes hundreds of percent, of air in the metal.)

The sprue is required to be tapered, the taper calculated to match, or very slightly compress, the natural form of the falling stream; the stream naturally narrows during its fall because of its acceleration under the action of gravity. By tailoring the shape of the sprue to the natural shape of the stream, the melt has the best chance to avoid the entrainment of air.

Parallel or reversed taper sprues are not recommended. Elliott and Mezoff (1948) show that their parallel slot sprues entrain air bubbles, finding that at low casting temperatures, their Mg alloy castings exhibit bubbles, whereas at high casting temperatures, the castings exhibit mainly oxide films. Although the authors conclude that the air is consumed by the Mg alloy, which will certainly be partly if not wholly true, it seems also likely that many larger bubbles would have time to escape at high temperatures before the formation of a solidified skin. The authors are clearly not aware of the seriousness of these residual oxides.

Special precautions are needed to reduce the damage introduced if a parallel or reverse tapered sprue design has to be used. Such features include the radial choke or a filter at the entrance to the sprue.

At one time, it was mandatory that each sprue had a sprue well at its base. The designs of filling systems recommended in this book explicitly avoid wells. This departure from tradition is possible only because the new filling systems are characterised by runners of approximately the same area as that of the sprue exit. It is to be noted that the traditional choice of sprue exit/runner/gate ratios of 1:2:2 and 1:2:4 etc. are automatically bad. The runner is too large to fill completely, regrettably entraining air, and so ensuring bubble damage problems.

An additional beneficial consequence of the avoidance of a well is the addition friction to the liquid provided by the additional solid surface of the mould at the point impacted by the metal as it turns the corner, therefore slowing the velocity of the liquid. This reduction in velocity is usually not great, in the region of 20%, but any reduction in speed is to be welcomed. If the sprue/runner junction is nicely formed, bubbles are formed for precisely 0 s. This awkward way of making a simple statement that no bubbles are formed is deliberate. It emphasises the contrast with filling systems that have been accepted as conventional up to now (for instance as in the study for Figure 10.9). Nowadays it is not necessary to accept a design that introduces any bubbles at all.

It is mandatory that no interruption to the pour occurs that leads to the lowering of the melt in the pouring basin below the minimum design level. If the sprue entrance is unpressurised in this way air will enter the running system. In the worst instance of this kind, if the basin level drops to the point that the sprue entrance becomes uncovered this has to be viewed as a disaster. A provision must be made for the foundry to reject automatically any castings that have suffered an interrupted pour, or slow pour that has allowed the basin to empty to a level below the designed minimum level.

To be safe, if a basin is to be used, it is worth ensuring that it has up to twice the required minimum depth to keep the sprue filled, and ensuring that the pourer keeps the basin topped up as highly as possible, well above the minimum level. In this way the casting may run a little faster, but air will be excluded and bubble damage avoided.

The best option is the avoidance of any kind of pouring basin by using contact pouring. This not only eliminates bubbles but bifilms too (basins have been designed to eliminate bubbles, but remain great manufacturers of bifilms).

10.4.4 BUBBLE DAMAGE IN COUNTER-GRAVITY SYSTEMS

Pumped systems such as Cosworth Process, or low pressure castings systems into sand moulds or dies, are highly favoured as having the potential to avoid the entrainment of bubbles if, and only if, the processes are carried out under proper control. The reader needs to be aware that good control of a *potentially* good process should not be assumed; it requires to be demonstrated.

Although low pressure filling systems can in principle satisfy the requirement for the complete avoidance of bubbles in the metal, a leaking riser tube in a low-pressure casting machine can lead to a serious violation (Figure 10.10(a)). The

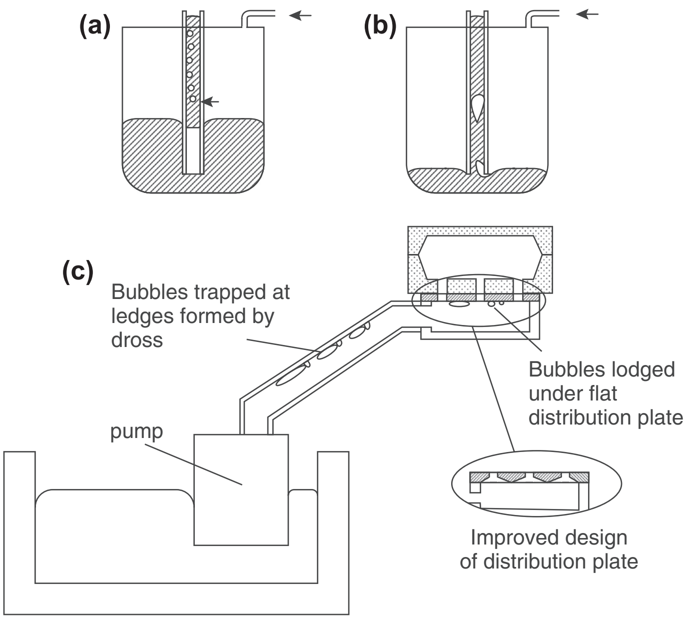

FIGURE 10.10

Bubbles introduced by defective counter-gravity systems. (a) Leak in a riser tube of a low pressure casting unit; (b) the dangerous ingestion of massive bubbles if the melt level is too low; (c) bubbles entrapped by dross and poor design features of some pumped systems, creating bubbles that are released erratically.

stream of bubbles from a leak in a defective riser tube will float directly up the tube and enter the casting. Unfortunately, this problem is not rare. Thus regular checks for such leakage, and the rejection of castings subjected to such consequent bubble damage, will be required. If the melt in the pressurised vessel becomes too low a huge amount of gas can erupt up the riser tube (Figure 10.10(b)) ejecting the liquid metal ahead.

For pumped systems, bubbles can be released erratically from the interior wall of a sloping tube launder system, especially if it is not cleaned with regular maintenance. Bubbles can become trapped and escape from time to time from the underside of a badly designed distribution plate used in some counter-gravity systems such as the early variants of the low-pressure sand system (Figure 10.10(c)).

A particularly insidious and little known source of a copious supply of bubbles in a low pressure casting system can come from a refractory lining in the pressurised furnace. This extremely important problem is dealt with in detail in Section 16.3.1.

All counter-gravity filling systems need to be designed to avoid the ingress of air or the trapping of air bubbles that might be randomly released.

10.5 RULE 5: AVOID CORE BLOWS

10.5.1 BACKGROUND

Gas bubbles can be forced into the liquid metal from a mould of core if their internal pressure from outgassing reactions exceeds the pressure in the melt. This section deals with an attempt to quantify the problem and the ways in which this problem can be avoided. However, there are additional dangers from gases 'blown' or 'fired' into the liquid metal from other sources which can be dealt with first.

Blows from the explosive volatilisation of core adhesives are well known if excess adhesive has been squeezed out from a sand joint so that it is brought into direct contact with the liquid metal.

Less well known is that considerable volumes of water vapour are given off from clay-based core repair and mould repair pastes. This is because the clay contains water of crystallisation, so that even after thoroughly drying the core repair at 100 or 200°C, the water bound in the structure of the clay remains unchanged, only being released at a high temperature, in the region of perhaps 600°C. Thus the water is released only when the clay contacts the liquid metal. This is particularly unfortunate, because the clay is composed of such fine particles that it is substantially impermeable, preventing the escape of the water into the core or mould, so that the water is forced to boil off through the metal. Repair of cores with clay-based pastes therefore generally leads automatically to blow defects. The wide use of core repair pastes illustrates that this danger is little known.

Unless you know from independent tests that your core repair technique does not result in core blows, the clear lesson is 'Do not repair cores.'

The generation of blows off metal chills is the result of an almost identical process. When a block metal chill is placed in a bonded aggregate mould, the pouring of the metal causes a rapid outgassing of the volatiles in the aggregate/binder mixture. The volatiles, particularly water vapour, are driven ahead of the spreading liquid metal and condense on any cold surface, such as a metal chill. When the liquid metal finally arrives and overruns the chill the condensates boil off. Because the chill is impermeable, the vapour is forced to bubble through the melt (Figure 10.11).

The prevention of blows from condensation on chills is widely known and generally well applied. The chill should normally be coated with a ceramic wash or spray that is afterwards thoroughly dried to give an inert, permeable and non-wetted surface layer. If this coat has residual surface roughness, so much the better. The effective permeability of the surface can be further enhanced by providing deep V grooves in a criss-cross pattern. The grooves are bridged to some degree by the action of the surface tension of the melt, so that the bottoms of the grooves act as surface vents, tunnelling the expanding vapours to freedom ahead of the advancing melt. Additionally, the V grooves are thought to enhance the effectiveness of the chilling action by increasing the contact between the casting and the chill, particularly during the cooling and contraction of the surface of the casting, forcing the grooved cast surface sideways against the walls of the grooves on the chill.

FIGURE 10.11

Blows off a damp steel chill in a steel sand casting. (In addition, it is to be noted that bubble damage would be expected in a vertical zone above the chill).

Courtesy S. Scholes.

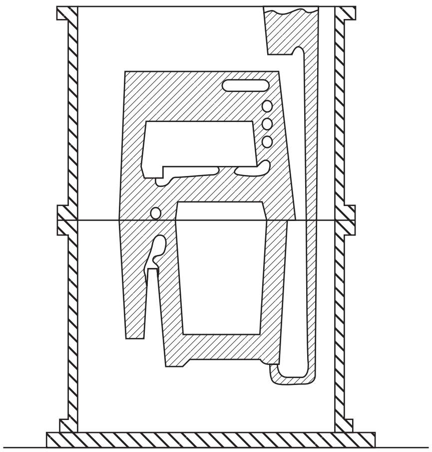

FIGURE 10.12

Casting in a close-fitting steel mould box on an unvented flat steel plate, showing blows from an upwardly oriented feature on the lower part of the mould.

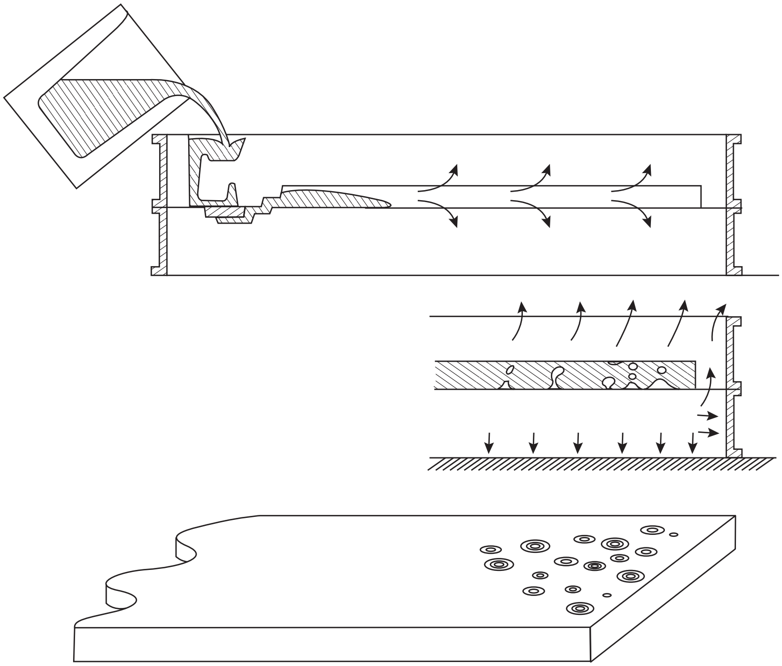

FIGURE 10.13

A large flat plate casting with an enclosed drag. Volatiles are driven ahead and condense in the cooler distant mould, exacerbating defects at the far end.

To demonstrate that a chill, a core, or assembly of cores, does not produce blows may require a procedure such as the removal of all or part of the cope or overlying cores, and taking a video recording of the filling of the mould. A small (probably sacrificial!) videocamera built into a completed mould might provide a better and economical test. If there are any such problems, the eruption of core gases will be clearly observable and will be seen to result in a boiling action, creating a froth of surface dross that would of course normally be entrapped inside the upper walls of the casting. A series of video recordings might be found to be necessary, showing the steady development of solutions to a core-blowing problem and recording how individual remedies resulted in progressive elimination of the problem. The video recording requires being retained by the foundry for inspection by the customer for the life of the component. Any change to the filling rate of the casting, or core design or the core repair procedure should necessitate a repeat of this exercise.

It should be noted that the danger of forming blows (bubbles in the melt) only arises if the liquid metal is subjected to the pressure in the mould (or core) and the pressurised gas has no option but to escape via the melt.

For instance, the general buildup of pressure in the mould cavity during filling is not usually a problem from the point of view of creating bubbles in the melt. In fact, for a fast-filling mould, the buildup of backpressure inside the mould cavity can be useful in slowing the final stages of filling, reducing the mould penetration that can occur because of the impact at the final instant of filling. Naturally, the converse is true for some difficult-to-fill castings, where mould backpressure can be sufficient to prevent the complete filling of moulds. The escape of gases entrapped in the mould cavity is made more difficult by the application of mould coats, so that pressures can easily be doubled (Ohnaka, 2003). Thus the internal pressure in the mould cavity may affect the filling of the mould positively or negatively, but will never usually lead to the creation of bubbles.

The buildup of pressure within the body of the mould *can* lead to problems. It can be severe in moulds that are enclosed in steel boxes and that are placed on a steel plate or on a concrete floor. The gases are relatively free to escape from the cope, but gases attempting to escape from the drag are sealed in by the overlying liquid in the mould cavity (Figures 10.12 and 10.13). The problem is enhanced if the casting is a tight fit in the moulding box, as is usually the case of course because the casting engineer is always trying to get as much value as possible out of each box. In fact the buildup of pressure inside sand moulds crammed into tight-fitting steel boxes has, in the author's experience, contributed to several spectacularly defective castings, and in one instance, to a casting that persistently refused to fill its mould because the backpressure of gases rose so high.

Fortunately, mould backpressure is easily dealt with by the provision of one or more whistlers. These are narrow, pencil-shaped vents connecting the mould cavity to the outside world via the cope.

Finally, it is worth commenting on the curiously misguided but common provision of whistler vents through the top of the mould in an effort to eliminate 'gas' porosity in the form of bubbles from some source. Clearly, this action is of no use. A moment's reflection reveals the self-evident fact that the pores (i.e. the entrained air bubbles) are already in the metal, and the metal itself would have to rise up the vent to eliminate the porosity from the casting. The error in thinking arises because of the confusion between gas entrapped in the mould cavity and gas entrained in the melt. When these are separated into their logically separate categories, confusion disappears and the correct remedy can be identified.

10.5.2 OUTGASSING PRESSURE IN CORES

The sudden heat from the liquid metal causes the volatile materials in the mould to evaporate fiercely. In greensand moulds and many other binder systems, the main component of this volatilisation is water. Even in so-called dry-binder systems, there is usually enough water to constitute a major contribution to the total volume of liberated gas. On contact with the hot metal, much of the water is decomposed to surface oxide films and to hydrogen. The high hydrogen contents of analysed mould gases are clearly seen in Figure 4.2.

In the case of the mould, the generation of copious volumes of gas is usually not a problem. The gas has plenty of opportunity to diffuse away through the bulk of the mould. The pressure buildup in a greensand cavity during mould filling is normally only of the order of a 100 mm water gauge (0.01 atm) according to measurements by Locke and Ashbrook (1972). This corresponds to merely 10 mm or so head pressure of liquid iron or steel. However, even this rather modest pressure might have been overestimated because their experimental arrangement corresponded to a closely fitting steel moulding box, and escape for mould gases only occurs via the cope. Even so, in greensand systems where the percentage of fines and clay and other constituents is high, the permeability of the mould falls to levels at which the ability of the mould volatiles to escape becomes a source of concern. The venting of the cope by needling with wires of about 3 mm diameter is a time-honoured method of re-introducing useful levels of permeability.

Chemically bonded moulds are usually of no concern from the point of view of generating a backpressure during the filling of the mould. This is because the sand is usually bought in as ready washed, cleaned and graded into closely similar sizes (a 'three pan sand'). In addition, usually only a 1 or 2 volume percent of binder is used, leaving an open, highly permeable bonded mould. A single measurement by the author using a water manometer showed a pressure rise during the filling of a cylinder head mould of less than 1 mm water gauge. Even this completely negligible rise seemed to decay to nothing within a second or less.

In the case of cores, however, once the core is covered by liquid metal, the escape of the core gases is limited to the area of the core prints, if the metal is not to be damaged by the passage of bubbles through it. Furthermore, the rate of heating of the core is often greater than that of the mould because it is usually surrounded on all sides by hot metal, and the volume of the core is, of course, much less. All these factors contribute to the internal pressure within the core rising rapidly to high values.

Many authors have attempted to provide solutions to the pressure generated within cores. However, there has until recently been no agreed method for monitoring the rate or quantity of evolved gases that corresponds with any accuracy to the conditions of casting. A result of one method by Naro and Pelfrey (1983) is shown in Figure 10.14. (This method is an improvement on earlier methods in which the water and other volatiles would condense in the pipework of the measuring apparatus, reducing the apparent volume of gases, and thereby invalidating the experimental results.)

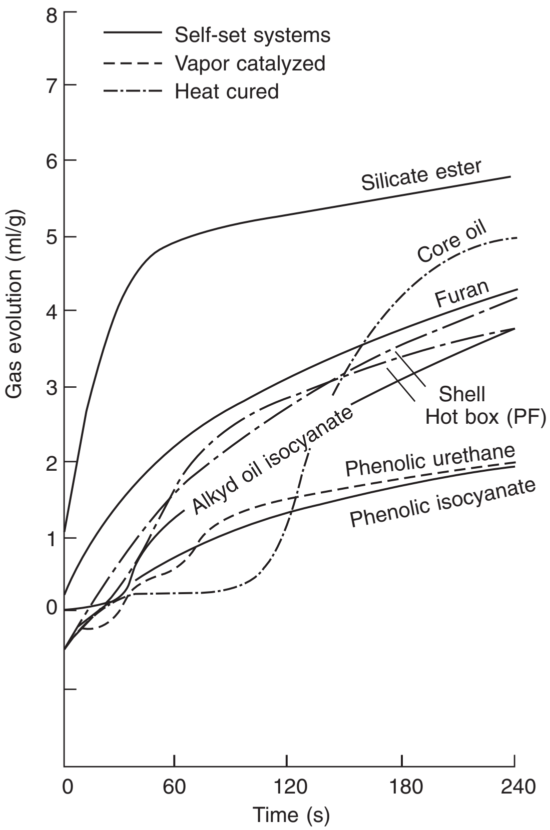

FIGURE 10.14

Gas evolution from carious binder systems using an improved test procedure that includes the contribution from water and other volatiles (Naro and Pelfrey, 1983).

The really important quantity given by these curves is the *rate* of evolution of gas. The rates, of course, are equal to the slope of the curves in Figure 10.14, and are presented in Figure 10.15. Only a few of their results are presented for clarity.

It is sufficient to note that the rates of outgassing are very different for different chemical binder systems. The initial high rate for the silicate is the result of its water content evolved at relatively low temperature during the early stages of the heating of the core. Other organic binders evolve their carbonaceous gases at a variety of temperatures, and apart from core oil, generally show much gentler rates that are more easily vented.

Anyway, taking these recent estimates of Q, the rate of volume of gas generated from a given weight of core in $mLg^{-1}s^{-1}$ (or preferably the identical-sized value in the equivalent unit $Lkg^{-1}s^{-1}$) as being of tolerable relevance to the real situation in castings, we can construct a core outgassing model. We shall roughly follow the method originally pioneered by Worman and Nieman (1973).

We first need to define the concept of permeability. This is a measure of the ease with which a fluid (the core gas in our case) can flow through a porous material (the core in our case). Permeability P_e is defined as the rate of gas flow Q (as a volume per unit time) through a permeable material of area A and length L and driven by a pressure difference ΔP:

$$P_e = QL/A\Delta P$$

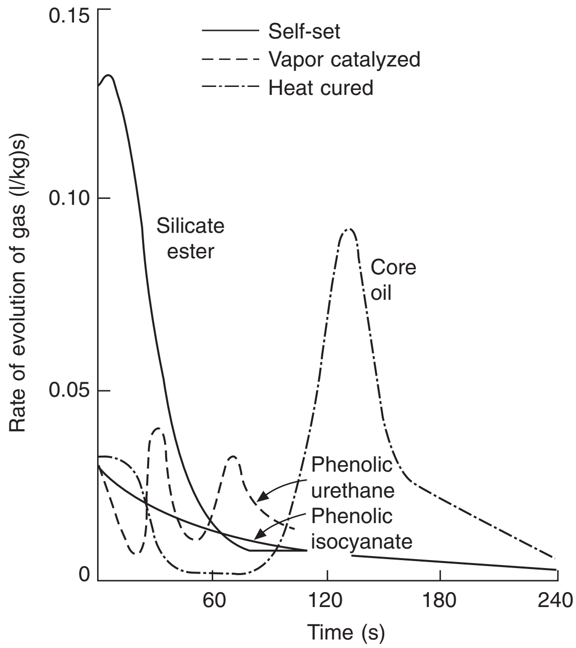

FIGURE 10.15

Rates of gas evolution from various sand binders based on the slopes of the curves shown in Figure 10.13.

The SI units of P_e are quickly seen to be:

$$[P_e \text{ units}] = [\text{litre}/s][m]/[m^2]\ [Pa]$$
$$= Ls^{-1}m^{-1}Pa^{-1}$$

Consider now our simple model of a core shown in Figure 10.16. The measured volume of gas evolved per second from a kilogram of core material is Q. If we allow for that this will have been measured at temperature T_1, usually above 100°C (373 K) to avoid condensation of moisture, and the temperature in the core at the point of generation is T_2, then the volume of gas produced in the core is actually QT_2/T_1 where T is measured in degrees K. For the casting of light alloys, the temperature ratio T_2/T_1 (in K remember) is about 3, whereas for steels it is nearly 6.

If we multiply this by the weight of sand heated by the liquid metal, then we obtain the total volume of gas evolved per second from the core. Thus if the heated layer is depth d, the core area A_c and density ρ, then the volume of gas evolved per second is $QdA_c\rho T_2/T_1$. If the core is surrounded by hot metal, this volume of gas has to diffuse to the print and force its way through the length L of the print of area A_p. We shall assume that the pressure drop experienced by the gas in diffusing through the bulk of the core is negligible in comparison with the difficulty of diffusing through the print (this is of course not always accurate depending on the shape of the core). Considering then the permeability definition for only the pressure drop along the print, we obtain the pressure in the core (above the ambient pressure at the outside tip of the print):

$$P = QdA_c\rho \cdot LT_2/A_pT_1P_e$$

This simple model emphasises the direct role of permeability P_e and of Q, the rate of gas evolution on the generation of pressure in the core. It is to be noted that the high casting temperature for steels is seen to result in values for P

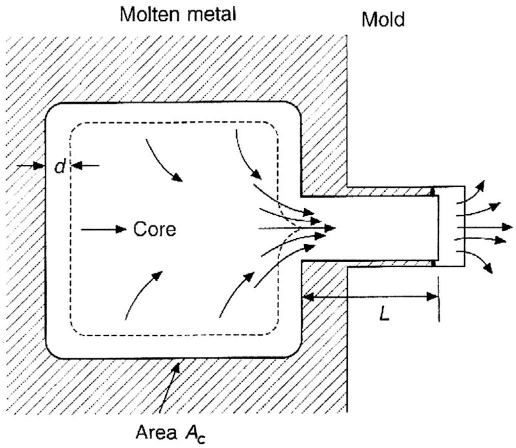

FIGURE 10.16

Core model, showing heated layer thickness *d* outgassing via its print. In this particular case, metal flash along the sides of the print force gas to exit only from the end area *A*.

approximately twice that for aluminium alloys. Thus cores in steel castings will be twice as likely to create blows than cores in aluminium alloy castings. For this reason, an enclosed core that would give no problems in an aluminium alloy casting may cause blows when the same pattern is used to make an equivalent bronze or iron casting.

Our model also highlights the various geometrical factors of importance. In particular, the area ratio of the core and the print, A_c/A_p, is a powerful multiplier effect, and might multiply the pressure by anything between 10 and 100 times for different core shapes. Also emphasised is the length L of the core print. If the print is a poor fit, then L may be unnecessarily lengthened by the flashing of the metal into the print so as to enclose the flow path in an even longer tunnel. If the liquid metal completely surrounds the end of the print too, then, of course, all venting of gases is prevented. Gases are then forced to escape through the molten metal, with consequential bubble damage to the casting.

The provision of a vent such as a drilled hole along the length of the print will effectively reduce L to zero; the model predicts that the internal pressure in the core will then be eliminated (the only remaining pressure will, of course, be the much smaller pressure to overcome the resistance to flow through the core itself). The value of vents in reducing blowing from cores has been emphasised by many workers. Caine and Toepke (1966), in particular, estimate that a vent will reduce the pressure inside a core by a large factor, perhaps 5 or 10. This is an important effect, far outweighing all other methods of reducing outgassing pressure in cores.

Vents can be moulded into the core, formed from waxed string. The core is heated to melt out the wax, and the string can then be withdrawn prior to casting. This traditional practice was often questioned as possibly being counter-productive, because of the extra volatiles from the wax that, on melting, soaks into the core. Such fears are seen to be happily unfounded. The technique is completely satisfactory because the presence of the vent completely overrides the effect of the extra volatile content of the core.

A final prediction from the model is the effect of temperature. In theory, a lowering of the casting temperature will lower the internal core pressure. However, this is quickly seen to be a negligible effect within the normal practical limits of casting temperatures. For instance, a large change of 100 K in the casting temperature of an aluminium alloy will change the pressure by a factor of approximately 100/900. This is only 11%. For irons and steels, the effect is smaller still. It and can therefore be abandoned as a useful control measure.

We shall now move on to some further general points.

Cores are almost never made from greensand because the volatile content (particularly water, of course) is too high and the permeability is too low (because of its 5–10% clay and other fines). In addition, the cores would be weak and unable to support themselves on small prints; they would simply sag. If greensand is used at all, then it is usually dried in an oven, producing ‘dry sand’ cores (their name should perhaps use the past tense and so be more accurately ‘dried sand’ cores). These are relatively free of volatiles and are mechanically strong. But because they retain the poor permeability of the original greensand they therefore usually require good venting. This is usually time-consuming and labour-intensive.

Sand cores are therefore not nowadays made from dried sand, but generally chemically bonded from clean, washed and dried silica sand that is closely graded in size to maintain as high a level of permeability as is possible. The limit to the size of sand grains and the permeability is set by the requirements of the casting to avoid (1) penetration by the metal and (2) the production of internal surfaces of the casting that are unacceptably rough.

These cores are bonded with a chemical binder that is cured by heat or chemical reaction to produce a rigid, easily handled shape. The numerous different systems in use all have different responses to the heat of the casting process, and produce gases of different kinds, in different amounts, at different times and at different rates, as illustrated in Figures 10.14 and 10.15. These results are not to be taken as absolute in any sense. Manufacturers’ products are changing all the time for a variety of reasons: health and safety; economics; commercial; changes in world markets and supplies of raw materials etc. Thus binder formulations change and new systems are being developed all the time. At present, the phenolic urethane systems are among the lowest overall producers of volatiles, which explains their current wide use for intricate cores, for instance in the case of water jackets for automobile cylinder heads and blocks.

Part of the reason for the historical success enjoyed by the phenolic urethane binders is their high strength, which means that the addition levels needed to achieve an easily handled core are low. This is one of the important factors in explaining their position near the bottom of Figure 10.14; the volume of gas evolved is, of course, proportional to the amount of binder present. This self-evident fact is clearly substantiated in the work of Scott et al. (1978), shown in Figure 10.17. (If allowance is made for the fact that these workers used a core sample size of 150 ml, corresponding to a weight of approximately 225 g, then the rate of evolution measurements converted to 1 $kg^{-1}s^{-1}$ agree closely with those presented in Figure 10.15. This is despite the significant differences in the techniques. The data in Figure 10.15 may therefore be of more universal application than is apparent at first sight.)

Renewed interest these days is being taken in inorganic binders based on salts. Preferably the salts should contain no water, particularly water bound up in the crystalline structure that could be evolved at the high temperature required for casting. It will repay us all to monitor these developments closely.

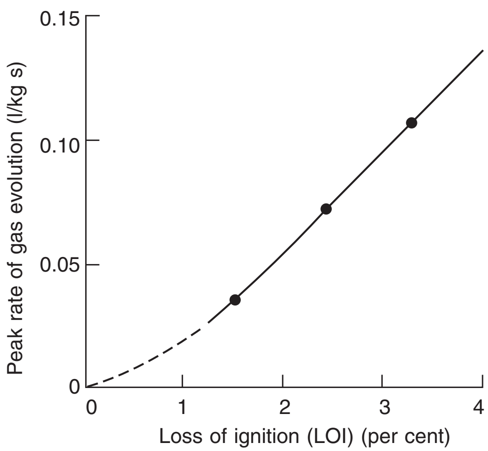

FIGURE 10.17

Increase in the peak rate of outgassing as loss on ignition (LOI) increases.

Data recalculated from Scott et al. (1978).

10.5.3 CORE BLOW MODEL STUDY

Campbell (1950) (not the author) devised a useful test that assessed the pressure in cores compared with the pressure in the liquid metal. A modified version of his test is shown in Figure 10.18; the following is a modification and further development of his explanation.

If the mould is filled quickly, then the hydrostatic pressure resulting from depth in the liquid metal is built up more rapidly than the pressure of gas in the core. This dominance of metallostatic pressure can persist throughout filling and solidification. The higher metal pressure effectively suppresses any bubbling of gas through the core at point A as illustrated in Figure 10.18(a). Bubbles will form at A only if the core pressure reaches the metal pressure because either the venting of the core is poor (Figure 10.18(b)) or the core is covered too slowly (Figure 10.18(c)).

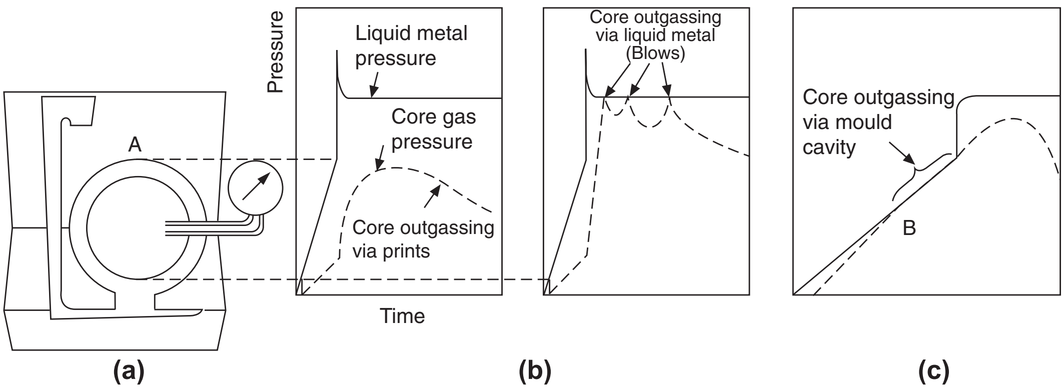

FIGURE 10.18

Effect of fill rate on core outgassing. (a) Fast fill with well-vented core; (b) fast fill but badly vented core; (c) slow fill.

Adapted from Campbell (1950).

If the mould is filled slowly, the gas pressure in the core exceeds the metal pressure at B and remains higher during most of the filling of the mould. Gas escapes through the top of the core during the late stages of the filling, so that, although the metal pressure in Figure 10.18(c) finally exceeds the gas pressure in this case, it is doubtful whether the gas will stop flowing. This is because, in practice, when the liquid metal reaches the top of the casting at A, it is prevented from closing and joining together by the constant evolution of gas. The migration of volatiles ahead of the slowly rising metal front also concentrates the gas source near A, and the creation of a well-established, well-oxidised bubble trail will further hinder the welding and closing together of the meniscus at this point. Gas issuing from the top of a submerged core can be seen to cause the melt to tremble and flutter against the surface of the core, the melt attempting to close together to stop the flow, but repeatedly failing in this attempt as drossy surface films thicken.

Figure 10.18(b) illustrates the case in which the pressure in the core rises only just above the metal pressure, with the result that a limited number of bubbles escape. If the casting has started to solidify, then although the bubble may be forced into the melt, it is unlikely to be able to escape from the upper surface of the casting.

10.5.4 PREVENTION OF BLOWS

By far the best solution to the evolution of gases from cores is the use of a sand binder for the core that has little or no evolution of gas as the core becomes hot. This would represent a perfect solution. The best hopes here are the inorganic binders that contain no water of crystallisation. However, the few binders that have so far been developed to meet this criterion are usually not satisfactory in other ways. At the time of writing the perfect core binder has yet to be developed!

In the meantime, one of the best actions to avoid blows from cores (or more occasionally from moulds) is to increase the permeability of the core by the use of a coarser aggregate and/or by the use of venting.

Because the core print is usually the area where all the escaping gas has to concentrate, a simple hole through the length of the print makes a huge impact on the problem, as has been shown previously. For some castings, this can be a complete solution. However, of course, if the vent hole can be continued to the centre of the core this is even better. The further provision of easy escape for gases through the mould and out to the atmosphere is necessary for copper-based and iron and steel castings where the outgassing problem becomes severe because of the higher temperatures. In iron and steel foundries, many readers will have seen the vigorous jets of burning carbon monoxide issuing from vents in moulds for many minutes after pouring. Figure 10.19 illustrates a succession of improved venting techniques.

For low-volume production involving the making of cores by hand, a vent can be provided along a curved path through a core by laying a waxed rope inside whilst it is being made. The core is subsequently heated to melt the wax so that the rope can be pulled out. As we have noted earlier, the additional volatiles from the wax remaining in the core are easily accommodated by the provision of the vent; the vent is an overwhelming benefit.

For more delicate low volume work, the author has witnessed long curved cores for aerospace castings being drilled by hand, using a drill bit fashioned from a length of piano wire held in a three-jaw rotary chuck driven by a small motor. The tip of the high-carbon-steel wire is hammered flat and ground to a sharp point shaped like an arrowhead. The core is drilled by hand in a series of straight lengths, the piano wire drill buzzing quickly through the core. Each hole is targeted to intersect the previous hole, the straight holes emerging on the bends, where the openings are subsequently plugged by a minute wipe of refractory cement (Figure 10.20). The complete vent is checked to ensure that it is continuous, and free from leaks, by blowing smoke in to one of the vent openings, and watching for the smoke to emerge from the far opening. Only when the smoke emerges freely at the far end, and from no other location, is the core accepted for use. It is then stored in readiness for mould assembly.

Occasionally, instead of an opening to the atmosphere, it is necessary to link the outside opening to a vacuum line. This is relatively common practice in gravity die (permanent mould) casting, increasing the efficiency of the extraction of gases from a resin-bonded sand core. However, the evolution of volatiles from the binder creates problems by condensing as sticky resins and tars in the vacuum line, so that, for long production runs, regular attention is required to avoid blockage, often dictating the timing of the withdrawal of the tooling for maintenance.

The reader is advised caution with regard to the application of a vacuum line to aid venting. The author once tried this on an extensive thin-section core with a single small area print around which was poured liquid stainless steel at 1600°C. The resulting rapid buildup of pressure was so dramatic that it blew off the vacuum connection with a bang! However, the discerning reader will notice the extreme circumstances described here, and rightly conclude that in this case the author was testing the patience of Providence.

The application of vacuum connections to core prints is widely practised in the gravity die casting of Al alloy automotive cylinder heads. The water jacket core is particularly difficult to vent without such extreme measures. However, because of the wide use of resin-based binders for this core, the vacuum line quickly becomes blocked by condensing resins and tars. Thus after perhaps only a dozen castings the die may have to be withdrawn from service to clean out the vacuum lines.

If there is an option, it is far better to arrange that the core vents through prints that are directed vertically upwards (Figure 10.21). This is because as the melt rises in the mould, the volatiles migrate through the core ahead of the metal, concentrating in the last part of the core. If the core is vented at its base, this is a potential disaster. The volatiles are too far from the print and will continue to be pushed ahead, finally being pushed into the form of an eruption of bubbles from the top of the core. This problem can be reduced by covering the core with liquid metal as quickly as possible. Venting from the base is then given its best chance.

Even so, a print allowing outgassing from the top of the core is ideal. If a vent cannot be provided up the centre of the top print, a top print is still valuable, even though it may contain no central vent because the volatiles will travel up the core surface. This can be seen on core prints that emerge from the tops of aluminium alloy castings. The melt is seen to

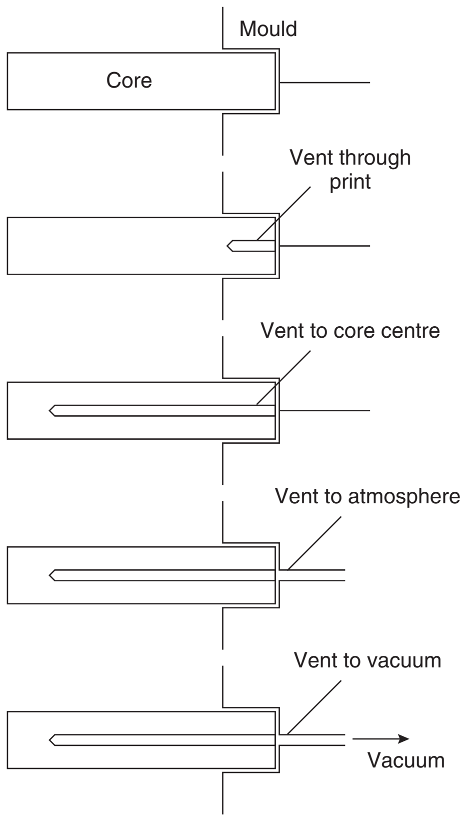

FIGURE 10.19

Venting of a core, illustrating progressively more effective techniques.

flutter, trembling against the side of the core print as gas rushes up in the form of minute, high-frequency waves, causing ripples to radiate out across the surface of the melt.

The provision of soft ceramic paper gaskets, preferably with a central hole, shaped like a washer, placed on the end of core prints is an excellent provision for the escape of gases. This simple remedy prevents the melt from flashing over the end of the print to block the vent (Figure 10.21). The compressible washer allows for the sealing of the core print against ingress by liquid metal, but allows closure of the mould without danger of the crushing of the core.

Finally, if the core can be covered quickly with liquid metal, and the pressure in the metal quickly raised to be at all times greater than the internal pressure generated inside the core, then bubble formation will be suppressed. Thus simply filling the mould faster is often a quick and complete solution to a core blow problem. The provision of an additional top feeder to increase hydrostatic pressure needs care because if the feeder has large volume the delay in the rise of pressure to fill it may be counter-productive. If feeding of the casting is not really required, the sprue and pouring basin can provide the early pressurisation that is needed, it would be better to leave well alone and not be tempted to add the top feeder.

Eventually, it is hoped, we can look forward to the day when computer simulation will provide an accurate description of each core and mould, allowing in detail for the effect of outgassing of cores of intricate geometries and the complicating effect of rate of filling. A welcome start has been made by Maeda et al. (2002), who demonstrated a computer simulation of the flow of gas through an aggregate core, and Starobin et al. (2010), who investigated water

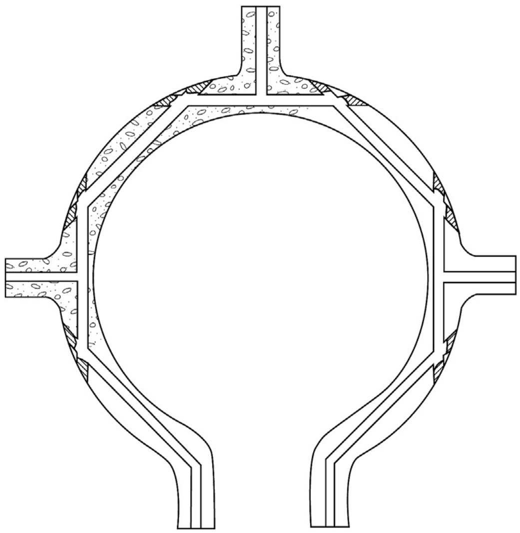

FIGURE 10.20

Drilled and plugged holes to vent a narrow ring-shaped core.

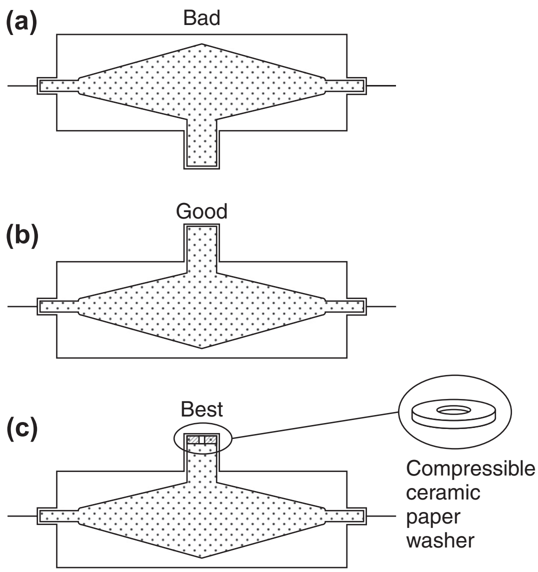

FIGURE 10.21

(a) Downwards venting core vents poorly, compared with (b) an upward vented core. (c) The cored compressible washer prevents the closure of the vent by ingress of metal. A further benefit would be a central hole though the core print, preferably extending into the centre of the core.

jacket cores for cylinder blocks. Perhaps we can now look forward to such studies becoming a commonplace feature of the design of a new casting.

10.5.5 SUMMARY OF BLOW PREVENTION

The various actions that can be taken to eliminate blow holes are:

1. Use a core binder that has minimal outgassing (i.e. minimise the binder so far as possible) or delayed outgassing (change to a late outgassing material).
2. Improve the permeability of the core aggregate.
3. Provide vents, particularly through the print.
4. If possible, vent the core from its top-most point.
5. Apply a vacuum to the core prints.
6. Redesign the core (or invert the casting) to avoid upward-pointing features.
7. Fill quickly to cover the core with liquid metal as soon as possible, before its internal pressure rises high enough to force a bubble into the melt.
8. Use a high hydrostatic head of liquid metal to suppress the evolution of gas from the surface of the core.
9. Raise the casting temperature. This is a last resort. The technique is quite commonly applied. It has the effect of giving more time for the core to outgas through the liquid metal before a solidified shell can form, so that bubbles will not be trapped by the growth of the solid skin of the casting. The approach can only be recommended with some reluctance, perhaps justifiable in an emergency, or perhaps only for grey iron cast with dry sand cores. In most situations, the passage of the bubbles through the melt will, even if the bubbles escape, result in damage to the casting in the form of the creation of bubble trails, leading possibly to both leakage and dross defects.

10.6 RULE 6: AVOID SHRINKAGE DAMAGE

10.6.1 DEFINITIONS AND BACKGROUND

Before getting launched into this section, we need to define some terms.

There is widespread confusion in parts of the casting industry, particularly in investment casting, between the concepts of *filling* and *feeding* of castings. It is essential to separate these two concepts.

Filling is self-evidently the short period during the pour and refers to the filling of the filling channels themselves (sometimes called the *priming* of the running system) and the filling of the mould cavity. This may only last seconds or minutes.

Feeding is the long, slow process that is required during the contraction of the liquid that takes place on freezing. This process takes minutes or hours depending on the size of the casting. It is made necessary as a result of the solid occupying less volume than the liquid, so the difference has to be provided from somewhere. This contraction on solidification is a necessary consequence of the liquid being a structure resembling a random close-packed array of atoms, compared to the solid, which has the denser regular close packing in a structure known as a crystal lattice.

Chapter 7 is required reading. It reminds the reader that three separate stages of contraction accompany the cooling of a liquid metal down to room temperature, but the linear contraction of the liquid and of the solid are of little consequence for feeding; we shall concentrate here on the solidification contraction. This contraction gives a volume deficit that can result in problems for the casting.

The volume contraction during freezing can be fed by several mechanisms, but mainly by liquid feeding from a feeder. If for some reason liquid metal to feed the contraction cannot easily be supplied, the contraction starts to act on the surrounding solid, drawing it in by plastic deformation (solid feeding) and at the same time stretching the liquid metal. This elastic expansion of the liquid is naturally accompanied by the development of a hydrostatic tension, a negative pressure which can provide the driving force for the nucleation and growth of porosity. These concepts of negative pressure are not easily understood. The reader is strongly recommended to read Chapter 7, where this phenomenon is described in detail.

10.6.2 FEEDING TO AVOID SHRINKAGE PROBLEMS

To allow ourselves the luxury of some repeated emphasis and further definitions: To provide for the fact that extra metal needs to be fed to the solidifying casting to compensate for the contraction on freezing, it is normal to provide a separate reservoir of metal. We shall call this reservoir a *feeder* because its action is to *feed* the casting i.e. to compensate for the solidification shrinkage (obvious really!).

In much casting literature the reservoir is known, less than obviously, as a *riser*, and worse still, may be confused with other channels that communicate with the top of the mould, such as vents, or whistlers because metal *rises* up these openings too. The author reserves the name *riser* for (1) the special kind of feeder that is connected to the side of the casting via a slot gate, and in which metal rises up at the same time as it rises in the mould cavity, and (2) the 'surge riser' designed to avoid the surge of first metal into the ingate.

If a feeder is *not* placed on a chunky casting that definitely requires feed metal, then we can expect a significant shrinkage problem. Figure 10.22 illustrates for an Al casting the absence of a feeder would result in a pore volume of up to 7%. That is a big porosity problem. However, as the feeder size increases in proportion to the casting, the resulting porosity reduces, eventually reaching a minimum at a feeder-to-casting modulus ratio of about 1.25. The porosity in the casting at this minimum depends on the gas content; at zero gas, the casting will be sound. At higher gas levels, the porosity rises, but it is clear that gas porosity does not raise the total porosity level so significantly as an undersized feeder. Although there is much discussion concerning the dangers of overfeeding of castings, the figure illustrates that an oversize feeder is far less damaging than an undersized feeder.

It is most important to be clear that the filling system (sometimes called the running system) is not normally required to provide any significant feeding. The filling system and the feeding system have two quite distinct roles: one fills the casting so its role is completed quickly, and the other feeds the shrinkage during solidification, whose role takes the much

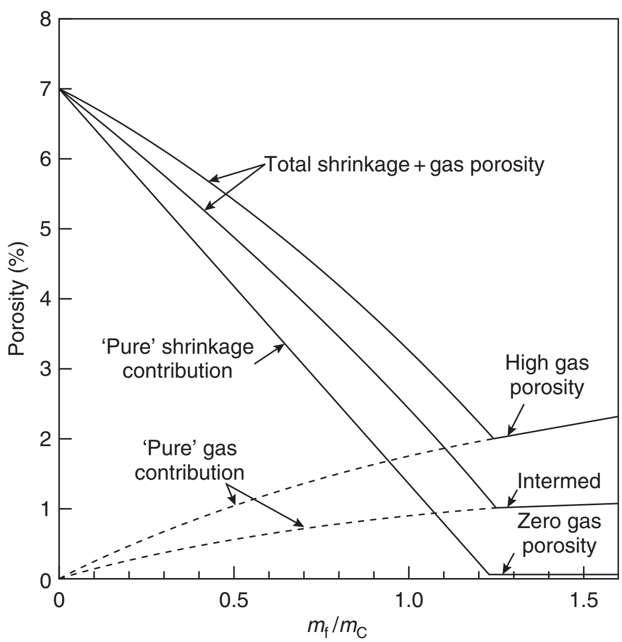

FIGURE 10.22

Generalised relation between gas and shrinkage as feeder size is increased in terms of the modulus ratio, based on particular results by Rao et al. (1975), shown in Figure 7.5.

longer time that the casting requires to freeze. (On occasions it is possible and valuable to carry out some feeding via the filling system, but this is somewhat unusual and requires special precautions. We shall deal with this later.)

Where computer modelling is not carried out, the following of the *seven feeding rules* by the author is strongly recommended. Even when computer simulation is available, the seven rules will be found to be good guidelines. For the computer itself, the following of Tiryakioglu's reduced rules constitutes the most powerful logic and is recommended, although the same rules also constitute a useful check for those of us who determine feeder sizes by pen and paper.

In addition to observing all the requirements of the rules for feeding, the use of all *five mechanisms* for feeding (as opposed to only liquid feeding) should also be used to advantage. These, once again, are described at length in Chapter 7. They will be found to be especially useful when attempting to achieve soundness in an isolated boss or heavy section where the provision of feed metal by conventional techniques may be impossible. However, a reminder of the attendant dangers of the use of solid feeding is presented later.

10.6.3 THE SEVEN FEEDING RULES

The *feeding rules* have already been briefly outlined in volume one, but are presented here in greater detail.

It is essential to understand that all of the rules must be fulfilled if castings are to be produced that require soundness, accuracy and high mechanical properties. The reader must not underestimate the scale of this problem. The breaking of only one rule may result in ineffective feeding and a defective casting. The wide prevalence of porosity in castings is a sobering reminder that solutions are often not straightforward. Because the calculation of the optimum feeder size is therefore so fraught with complications, it is dangerous if calculated wrongly, and the casting engineer is strongly recommended to consider whether a feeder is really necessary at all. (For instance, many modest feeding problems are easily dealt with by the provision of a chill or cooling fin rather than a feeder. Many thin-walled castings do not even require this because thin sections in general act good naturedly as though self-feeding.)

Feeding rule 1: Do not feed (unless necessary)

This first and most important question relating to the provision of a feeder on a casting is 'Should we have a feeder at all?' It is a question well worth asking.

For instance, most castings I see being produced in foundries are covered in feeders when it is evident there is no real shrinkage problem. The so-called 'shrinkage' being actually a mass of bifilms and bubbles because of a poor filling system.

Another practice which deserves a separate feeding rule of its own is the topping up of feeders with hot metal after the mould has been filled. This is a seriously damaging practice; the minimal aid to feeding temperature gradient by this addition of hot metal is vastly countered by the serious damage done to the casting by the surface oxide on the feeder being dragged down by the falling metal. This entrained oxide is carried down through the feeder and enters the casting, producing serious defects that *appear* to be shrinkage just below the feeder. Thus if a feeder is to be provided to a casting, we should adopt the additional feeding rule 'Never top up feeders'.

The avoidance of feeding is to be greatly encouraged, in contrast to the teaching in most traditional foundry texts. There are several reasons to avoid the placing of feeders on castings. The obvious one is cost. They cost money to put on and money to take off. In addition, many castings are actually impaired by the inappropriate placing of a feeder. This is especially true for thin-walled castings, where the filling of the feeder diverts metal from the filling of the casting, with the creation of a misrun casting.

Probably half of the small- and medium-sized castings made today do not need to be fed. This is especially true as modern castings are being designed with progressively thinner walls. In fact, as we have already mentioned, the siting of heavy feeders on the top of thin-walled castings is positively unhelpful for the filling of the casting because the slow filling of the feeder delays the filling of the thin sections at the top of the casting, with consequent misruns. Sometimes, as a result of the presence of a feeder, the casting suffers delayed cooling, impaired properties, and even segregation problems. Finally, it is easy to make an error in the estimation of the appropriate feeder size, with the result that the casting can be more defective than if no feeder were used at all.

As a general rule, therefore, it is best to avoid the placing of feeders on thin-walled castings. The low feed requirement of thin walls can be partly understood by assuming that of the total of 7% solidification shrinkage in an aluminium casting, 6% is easily provided along the relatively open pathway through the growing dendrites. Only about the last 1% of the volume deficit is difficult to provide. Thus if this final percentage of contraction on freezing has to be provided by solid feeding, moving the walls of the casting inwards, this becomes, at worst, 0.5% per face, which on a 4 mm thick wall is only 20 μm. This small movement is effectively not measurable because it is less than the surface roughness. If this deficit does appear as internal porosity, then it is in any case rather limited, and normally of little consequence in commercial castings. (It may require some attention in castings for safety-critical and aerospace applications.)

The other feature of thin-walled castings is that considerable solidification will often take place during pouring. Thus the casting is effectively being fed via the filling system. The extent to which this occurs will, of course, vary considerably with section thickness and pouring rate. If the section thickness (or rather, modulus; see later) of the filling system is similar to that of the casting, then feeding via the filling system might be a valuable simplification and cost saving. This important and welcome benefit to cost reduction is strongly recommended.

Because the calculation of the optimum feeder size is so fraught with complications, is dangerous if calculated wrongly, and costs money and effort, the casting engineer is strongly recommended to consider whether a feeder is really necessary at all. Rule 1 is probably the most valuable rule.

Of the remainder of castings that do suffer some feeding demand, many could avoid the use of a feeder by the judicious application of chills or cooling fins. The general faster freezing of the casting might then allow the provision of sufficient residual feeding via the filling system as indicated previously. Minor revisions, opening up restrictions to the feed path along the length of the filling system may provide valuable (and effectively 'free') feeding from the pouring basin.

However, this still leaves a reasonable number of castings that have heavy sections, isolated heavy bosses, or other features which cannot easily be chilled and thus *do* need to be fed with the correct size of feeder. The remainder of the rules are devoted to getting these castings right.

Feeding rule 2: The heat-transfer requirement

Chvorinov's heat transfer requirement for successful feeding can be stated as follows: *The feeder must solidify at the same time as, or later than, the casting.*

Nowadays, this problem can be solved by computer simulation of solidification of the casting. Nevertheless, it is useful for the reader to have a good understanding of the physics of feeding, so that computer predictions can be checked. We need to keep in mind that the computer simulation may not be especially accurate because much of the basic input data are sometimes not well known or wrongly inputted. Also, of course, computer time could be usefully avoided in sufficiently simple cases. In this chapter we shall concentrate on approaches which do not require a computer.

The freezing time of any solidifying body is approximately controlled by its ratio (volume)/(cooling surface area) known as its modulus, m. Thus the problem of ensuring that the feeder has a longer solidification time than that of the casting is simply to ensure that the feeder modulus m_f is larger than the casting modulus m_c. To allow a factor of safety, particularly in view of the potential for errors of nearly 20% when converting from modulus to freezing time, it is normal to increase the freezing time of the feeder by 20% i.e. by a factor of 1.2. Thus the heat-transfer condition becomes simply

$$m_f > 1.2\, m_c \tag{10.1}$$

It is important to notice that the modulus has dimensions of length. Using SI units, it is appropriate to use millimetres. (Take care to note that in French literature, the normal units are centimetres, and in the United States at the present time, to the despair of all those promoters of the welcome logic of SI units (the *Systeme International*), a confusing mixture of millimetres, centimetres, inches and feet. It is essential therefore to quote the length units in which you are working.)

Figure 10.22 confirms the optimum feeder modulus of factor approximately 1.2 times the casting modulus. The danger of insufficient feeding is clear, whereas overfeeding is clearly preferable. The contribution of gas porosity to the total unsoundness of the casting is also seen superimposed on the feeding response.

The modulus of a feeder can be artificially increased by the use of an insulating or exothermic sleeve. It can be further increased by an insulating or exothermic powder applied to its open top surface after casting. Recent developments in such exothermic additions have attempted to ensure that after the exothermic reaction is over, the spent exothermic material continues in place as a reasonable thermal insulator. These products are constantly being further developed, so the manufacturer's catalogue should be consulted when working out minimum feeder sizes when using such aids. However, as a guide as to what can be achieved at the present time, a cylindrical feeder in an insulating material is only 0.63*D* in diameter compared with the diameter *D* in sand. This particular insulated feeder therefore has only 40% of the volume of the sand feeder. Useful saving can therefore be made, but have, of course, to be weighed against the cost of the insulating sleeve and the organisational effort to purchase, store and schedule it etc. However, a further benefit that is easily overlooked from the use of a more compact feeder is the faster pressurisation (1) of thin sections that may aid filling and so reduce losses from occasional incomplete filling of mould cavities and (2) of cores to reduce the chance of blows.

When working out the modulus of the casting, it is necessary to consider which parts are in good thermal communication. These regions should then be treated as a whole, characterised by a single modulus value. Parts of the casting that are not in good thermal communication can be treated as separate castings.

For instance, castings of high thermal conductivity such as those of aluminium- and copper-based alloys can nearly always be treated as a whole because when extensive thin sections cool attached thicker sections and bosses, the thin sections act as cooling fins for the thicker sections. Conversely, of course, the thick sections help to maintain the temperature of thinner sections. The effect of thin sections acting as cooling fins extends for up to approximately 10 times the thickness of the thin section.

However, for castings of low thermal conductivity materials such as steel and nickel-based alloys (and surprisingly, the copper-based Al-bronze), practically every part of the casting can be treated as separate from every other. Thus a complex product can be dealt with as an assembly of primitive shapes: plates, cubes, cylinders etc. (making allowance, of course, for their common mating faces, which do not count as cooling area in the modulus estimate).

Table 5.3 lists some common primitive shapes. Familiarisation with these will greatly assist the estimation of appropriate feeder modulus requirements.

Feeder rule 3: Mass transfer (volume) requirement

The second and widely understood and well-used rule, usually known as the volume criterion is as follows: *The feeder must contain sufficient liquid to meet the volume-contraction requirements of the casting*.

At first sight, it may seem surprising that when the heat transfer requirement discussed previously is satisfied then the volume requirement is not automatically satisfied also. However, this is definitely not the case. Although we may have provided a feeder of such a size that it would theoretically contain liquid until after the casting is solid, in fact it may still be too small to deliver the volume of feed liquid that the casting demands. Thus it will be prematurely sucked dry, and the resulting shrinkage cavity will extend into the casting.

Different metals and alloys need significantly different amounts of feed metal. Figure 10.23(a) illustrates a section of a feeder on a plate casting in which the required shrinkage volume is just nicely concentrated in the feeder. This is the success we all hope for. However, success is not always easily achieved, and Figure 10.23(b),(c) and (d) show the complication posed by the different shrinkage behaviour of different alloys. The pure Al and the Al-12Si alloy are both short freezing range, and contrast with the Al-5Mg alloy which is a long freezing range material.

Some additional points of complexity in the operation of feeders in real life need to be emphasised.

1. The Mg-containing alloy in Figure 10.23(d) will almost certainly contain some fine, scattered microporosity that will have acted to reduce the apparent shrinkage cavity. Thus in practice the demand from the feeder will vary with the gas content of the metal.
2. The complicated form of the pipe in Al-12Si alloy (Figure 10.23(c) and (e)) almost certainly reflects the presence of large oxide films that were introduced by the pouring of the castings. These large planar defects fragment both the heat flow and the mass flow in the feeder, and the short freezing range and surface tension conspire to round off the cavities in the separated volumes of liquid isolated by the planes of oxide. In addition, the oxide, together with the solidifying crust on the top surface of the feeder also has some strength and rigidity, again complicating the collapse of the feeder top and influencing the shape of the shrinkage pipe as it, and its associated oxide skin, gradually expands downwards. These effects are additional reasons for the 20% safety factor often used for the calculation of feeder sizes. Feeders often do not have the simple carrot-shaped shrinkage pipe predicted by the

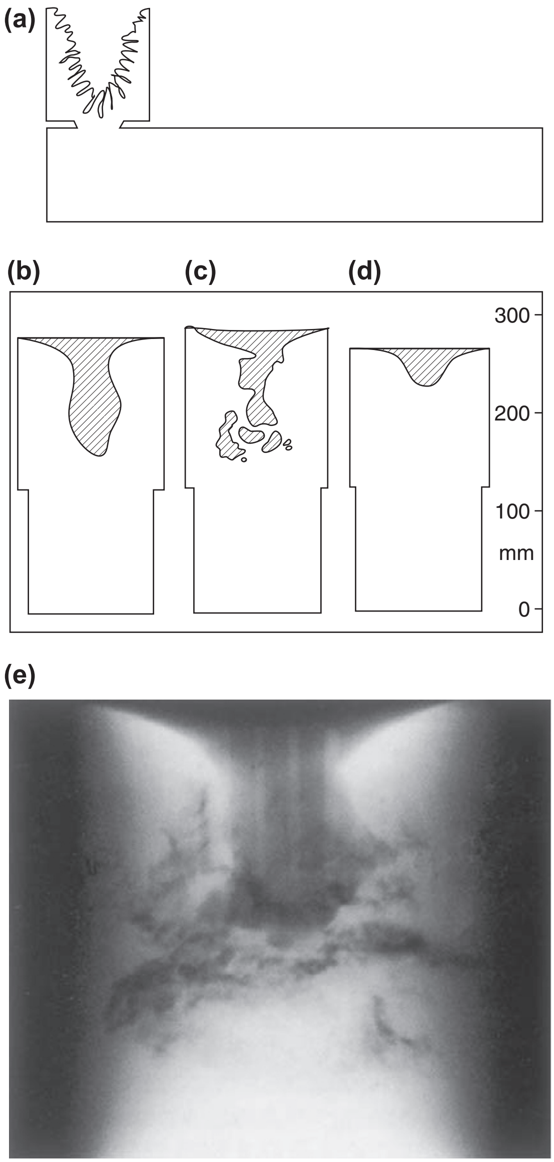

FIGURE 10.23

Cross-section of (a) simple plate casting, nicely fed, with all its shrinkage porosity concentrated in the feeder; (b) 99.5Al; (c) Al-12Si; (d) Al-5Mg; (e) radiograph of Al-12Si alloy feeder.

Courtesy Bird (1989).

computer. The radiograph of the feeder in Figure 10.23(e) is an excellent example of the action of large bifilms fracturing and diverting the melt draining from a feeder.

Figure 10.23 illustrates that normal feeders are relatively inefficient in the amount of feed metal that they are able to provide. This is because they are themselves freezing at the same time as the casting, depleting the liquid reserves of the reservoir. Effectively, the feeder has to feed both itself and the casting. We can allow for this in the following way. If we denote the efficiency e of the feeder as the ratio (volume of available feed metal)/(volume of feeder, V_f) then the volume of feed metal is, of course, eV_f. Because the liquid contracts by an amount α during freezing, then the feed demand from both the feeder and casting together is $\alpha(V_f + V_c)$, and hence:

$$eV_f = \alpha(V_f + V_c) \tag{10.2}$$

or:

$$V_f = (e - \alpha)V_c \tag{10.3}$$

For aluminium where $\alpha = 7\%$ approximately (see Table 7.2 for values of α for other metals), and for a normal cylindrical feeder of $H = 1.5D$ where $e = 14\%$, we find:

$$V_f = V_c \tag{10.4}$$

That is, there is as much metal in the feeder as in the casting! This is partly why the yield (measured as the weight of metal going into a foundry divided by the weight of good castings delivered) in many aluminium foundries is rarely above 50%. Metal in the running system, and scrap allowance will reduce the overall yield of good castings even further so that overall yields are often closer to 45%. (In comparison, the economic benefits of higher-yield casting processes such as counter-gravity casting, in which metallic yields of 80–90% are common, appear compellingly attractive, especially for high-volume foundries.)

For steels, the value of α lies between 3 and 4%, depending on whether solidification is to the body-centred cubic or face-centred cubic structures. For pure Fe-C steels the face-centred cubic structure applies above 0.1% carbon where the melt solidifies to austenite. For $\alpha = 4$ and $e = 14\%$, Eqn (10.3) gives:

$$V_f = 0.40V_c$$

and for steel that freezes to the body-centred cubic structure (delta ferrite) with $\alpha = 3$, and using a feeder of 14% efficiency we have:

$$V_f = 0.27V_c$$

Thus, compared with Al alloys, the smaller solidification shrinkage of ferrous metals reduces the volume requirement of the feeder considerably. For graphitic cast irons, the value reduces even further of course, becoming approximately zero in the region of 3.6–4.0% carbon equivalent. Curiously, a feeder may still be required because of the difference in *timing* between feed demand and graphite expansion, as will be described later.

The interesting reverse tapered feeder (Figure 10.24(c)) has been promoted for many years (Heine, 1982; Creese and Xia, 1991) and is currently widely used for ductile iron castings. Even so, the reader needs to be aware that in the opinion of the author, Figure 10.24 may not be as accurate we would like. At this time, the extent of the uncertainties is not known following the recent work of Sun and Campbell (2003). This investigation of the effect of positive and negative tapers on the efficiencies of feeders, found that the reverse tapered feeder (Figure 10.24(c)) appeared to be less efficient than parallel sided cylindrical feeders, or even feeders with a slight positive taper. These doubts are an unwelcome sign of the extent of our ignorance of the best feeder designs at this time.

Whether the size of the feeder is dictated by the thermal or volume requirement is related to the geometry of the casting. Figure 10.25 shows a theoretical example, calculated neglecting non-cooling interfaces for simplicity. Curve A is the minimum feeder volume needed to satisfy the thermal condition $m_f = 1.2m_c$; and curve B is the minimum feeder volume needed to satisfy the feed demand criterion based on 4% volume shrinkage and 14% metal utilisation from the feeder. Figure 10.25 reveals that chunky steel plates up to an aspect ratio of about 6 or 7 length to thickness are properly

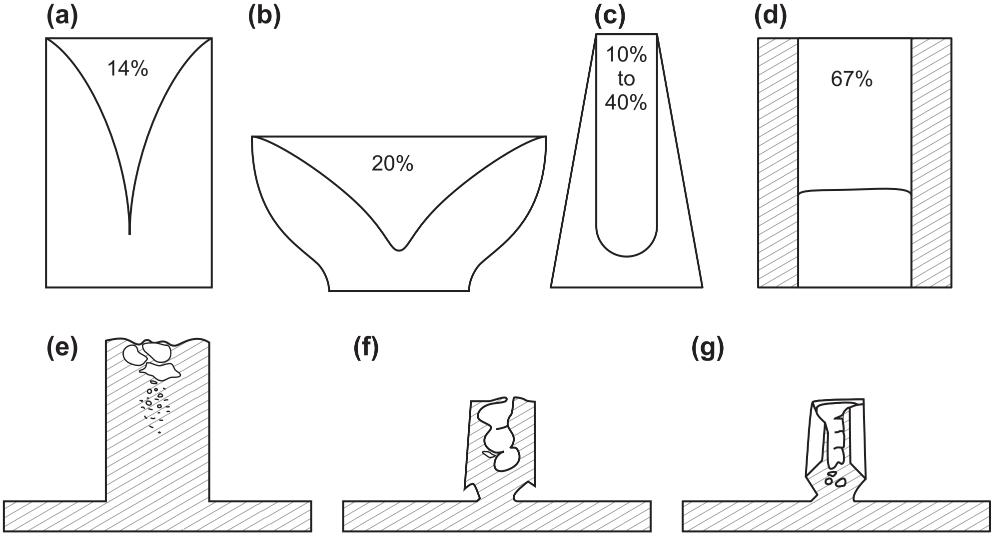

FIGURE 10.24

Metal utilisation of feeders of various forms moulded in sand. The (a) cylindrical and (b) hemispherical heads have been treated with normal feeding compounds; (c) efficiency of reverse tapered feeders depends on detailed geometry (Heine, 1982; Heine and Uicker 1983); (d) exothermic sleeve (Beeley, 1972); metal utilisation for ductile iron plates with (e) cylindrical sand feeder; (f) insulating feeder; and (g) cruciform exothermic feeder. (*After Foseco, 1988.*)

fed by a feeder dictated by freezing time requirements. However, thin section steel plates above this critical ratio always freeze first, and so require a feeder size dictated by volume requirements.

In fact the shape of the shrinkage pipe in the feeder is likely to be different for each of these conditions. For instance, the feeder efficiencies shown in Figure 10.24 are appropriate for the feeding of chunky castings because the continuing demand of feed metal from the casting until the feeder itself is almost solid naturally creates a long, tapering shrinkage pipe, resembling a carrot.

In the case of the more rapid solidification of thin castings, the relatively large diameter feeder needed to provide the volume requirement will give a shallowly dished shape in the top of the feeder because the feed metal is provided early, before the feeder itself has solidified to any great extent. The efficiency of utilisation of the feeder will therefore be expected to be significantly higher, as confirmed by Figure 10.25.

Research is needed to clarify this point. In the meantime, the casting engineer needs to treat the present data with caution, and conclusions from Figures 10.30 and 10.31, for instance, have to be viewed as illustrative of general principles rather than numerically accurate. Clearly, it is desirable to achieve smaller, more cost-effective feeders. The change of feeder efficiency depending on whether freezing or volume requirements are operating requires more work to clarify this uncertainty. In the meantime, this problem illustrates the power of a good computer simulation to avoid the necessity for simplifying assumptions.

A further use of feeders where the casting engineer requires care is the use of a blind feeder sited low down on the casting. The problems are compounded if such a low-sited blind feeder is used together with an open feeder placed higher. It must be remembered that during the early stages of freezing the top feeder is supplying metal to the blind feeder as well as the casting. The blind feeder has to be treated as though it is an integral part of the casting. The size of the top feeder needs to be enlarged accordingly.

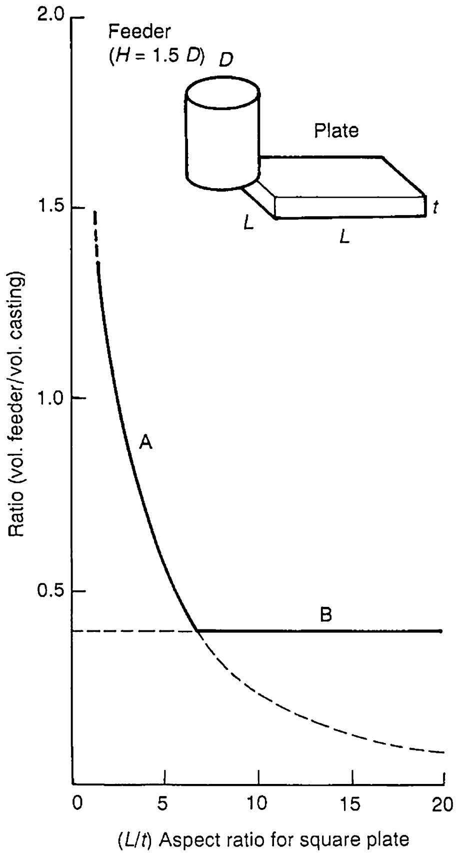

FIGURE 10.25

Feeder volume based on a feeder moulded in sand, and calculated neglecting non-cooling interfaces for simplicity. Curve A is the minimum feeder volume to satisfy $m_f = 1.2m_C$ and curve B is the minimum volume to satisfy the feed demand of 4% volume shrinkage and 14% utilisation of the feeder.

The blind feeder only starts to operate independently when the feed path from the top feeder freezes off. This point occurs when the solidification front has progressed a distance $d/2$, where d is the thickness of the thickest casting section between the top and the blind feeders. Thus the volume of the blind feeder is now reduced by the $d/2$ thickness layer of solid that has already frozen around its inner walls.

If this caution were not already enough, a further pitfall is that the thickness of solid shell inside the blind feeder may now exceed the length of the atmospheric vent core, creeping over its end and sealing it from the atmosphere. The blind feeder is therefore prevented from breathing, and is unable to provide any feed metal. In fact it now works in reverse, sucking metal out of the casting.

There are therefore subtleties in the operation of blind feeders that make success illusive. It is easy to make a mistake in their application, and the correct operation of the atmospheric vent is not always guaranteed, so it is difficult to recommend their use on smaller castings. For larger castings, where the size of the blind feeder is large, the collapse of the top of the feeder is more predictable, so that the action of the blind feeder becomes more reliable.

Whereas the size of feeders for alloys such as those based on alloys that shrink in a conventional fashion on freezing are straightforward to understand and work with, graphitic cast irons are considerably more complicated in their behaviour. They are therefore more complicated to feed, and the estimate of feeder sizes subject to more uncertainty.

The amount of graphite that is precipitated depends strongly on factors that are not easy to control, particularly the efficiency of inoculation. In addition, the expansion of the graphite can lead to an expansion of the casting if the mould and its container are not rigid. Because a mould and a moulding box can never be perfectly rigid, this leads to a larger volume casting that requires more metal to feed it, with the danger that the feeder may now be insufficient to provide this additional volume. Shrinkage porosity as a result of mould dilation is a common feature of iron castings.

One of the ways to reduce (but of course, never to eliminate) this problem is to use very dense, rigid mould in rigid, well-engineered boxes. Furthermore, the expansion of the graphite can be accommodated without swelling the casting by allowing the excess melt to exude out of the casting and back into the feeder. The provision of a small feeder is therefore essential to the production of many geometries of small iron castings. This small feeder provides essential feed metal during the time that normal shrinkage occurs, but also acts as an overflow during graphite expansion. It follows therefore that subsequent examination of the feeder indicates, mysteriously, that no net volume of liquid has flowed out from the feeder. If, as a result of appearing to have provided no feed metal, and concluding that the feeder had served no purpose, the feeder was removed; the result would have been a porous casting!

Ductile cast irons commonly use a reverse-tapered feeder such as that shown in Figure 10.24(c). If a feeder remains full for too long, its top would freeze over, isolating it from the atmosphere and so preventing the delivery of any liquid. It is logical therefore to encourage the feeder to start feeding almost immediately, by tapering the feeder so as to concentrate the action of shrinkage throughout the casting all on the small area at the top of the feeder. In this way, the level of liquid in the top of the feeder falls quickly, becoming surrounded by hot metal, so very soon there is no danger of it freezing over. For this to happen, it is essential that no feeding continues to be provided via the running system. This would keep the feeder full for too long. Thus it is necessary to design the ingate to freeze quickly. The feeder then works well.

This feeding technique, although at this time used exclusively in the ductile iron industry so far as the author is aware, would be expected to be applicable to a wide variety of metals and alloys.

Feeding rule 4: The junction requirement

The junction problem is a pitfall awaiting the unwary. It occurs because the simple act of placing a feeder on a casting creates a junction. As we shall see in Section 12.4.8, a junction with an inappropriate geometry will lead to a hot spot, with the danger that the hot spot may cause a shrinkage cavity that extends into the casting.

We shall see later (in the section on filling system design, dealing with the design of ingates) that is it easy to create a junction, planting either a feeder or a gate on to a casting, only to find that it has created a hot spot, and so leads to a localised feeding problem at the junction. After the feeder or gate is cut off, a shrinkage cavity is found extending into the casting. This occurs when the ratio of modulus of the addition (the feeder or gate) to the modulus of the casting is close to 1:1.

The simplest example illustrating the problem clearly is that of the feeding of a cube. The cube casting has the reputation of being notoriously difficult to feed. This is because the casting technologist, carefully following rule 2, calculates a feeder of 1.0 or perhaps 1.2 times the modulus of the cube. If the cube has side length D, then the feeder of 1:1 height to diameter ratio works out to have a diameter of $1.2D$. Thus the cube appears to require a feeder of rather similar volume sitting on top. However, the cube and its feeder are now a single compact shape that solidifies as a whole, with its thermal centre in the centre of the new total cast shape i.e. approximately in the centre of the junction. The combination therefore develops a shrinkage cavity at the junction, the hot spot between the casting and feeder. When the casting is cut off from the feeder, the porosity that is found is generally called '*under-feeder shrinkage porosity*.' This rather pompous pseudo-technical jargon clouds the clear conclusion that *the feeder is too small*.

The junction rules, developed later in the section on design of ingates, indicate that to avoid creating a hot spot we need to ensure that the feeder actually has twice the modulus of the casting. Thus the cube should have had a feeder of side length $2D$. The shrinkage cavity would then have been concentrated only in the feeder.

The junction problem is a widely overlooked requirement. The use of Chvorinov's equation systematically gives the wrong answer for this reason, so the junction requirement is often found to override Chvorinov. This is an important result that has caused much trouble for methoding engineers over the years and requires us all to update our thinking.

However, in some cases, the junction problem can be avoided. The simplest solution is not to place the feeder directly on the casting so as to create a junction. It happens that this rule is not easily applied to a cube because there is no alternative site for it.

However, in the case of a plate casting, there are options. The feeder should not be placed directly on the plate, but should be placed on an extension of the plate.

The general rules to solve the junction problem are therefore as follows:

1. Appendages such as feeders and ingates should not be planted on the casting so as to create a T- (or an L-) junction (although the L-junction is rather less detrimental than the T-junction). They are best added as extensions to a section, as an elongation to a wall or plate, effectively moving the junction off the casting.
2. If there is no alternative to the placing of the feeder directly on the casting, then to avoid the hot spot in the middle of the junction, the additional requirement that the feeder must meet is, if a T-junction,

$$m_f > 2m_c \tag{10.5}$$

or if an L-junction

$$m_f > 1.33m_c \tag{10.6}$$

The value of the constants is taken from Sciama (1974).

Note that no safety factor of 1.2 has been applied to these feeder sizes. This is because the shrinkage cavity does not occur exactly at the geometric centre of the freezing volume trapped at the thermal centre; the cavity 'floats' to the top, and the feed liquid finds its level at the base of the isolated region. Thus the final shrinkage cavity is naturally displaced above the junction interface, giving a natural 'built-in' safety factor.

Note that we have assumed that the feeder is above the casting to feed downwards under gravity. This is the recommended safe way to use feeders. If the feeder (or large gate) were placed *below* the casting, gravity would now act in reverse, so that any shrinkage cavity caused by the junction would float into the casting (in other words, the residual liquid metal in the casting drains into the feeder). This action illustrates one of the dangers of attempting to feed uphill.

Although it is not a good idea to make the feeder any larger than is really required, if it is only marginally adequate, the root tip of porosity seen in Figures 10.29 and 10.30 may on occasions just enter the casting, and may therefore be unacceptable. This necessitates the application of a safety factor, giving a feeder of larger size on average, but still just acceptable even when all the variables are loaded against it. It is common to use the safety factor 1.2 for the estimation of its modulus.

Feeding rule 5: Feed path requirement (the communication requirement)

There must be a path to allow feed metal to reach those regions that require it. It is clearly no use having feed metal available at one point on the casting, unable to reach a more distant point where it is needed. Clearly there has to be a way through. The reader can see why this criterion has been often overlooked as a separate rule; the communication criterion appears self-evident! Nevertheless, it does have several geometrical implications which are not so self-evident and which will be discussed.

In a valuable insight, Heine (1968) has drawn attention to the highest-modulus regions in a casting are either potential regions for shrinkage porosity if left unfed, or may be feeding paths if connected to feed metal. He recommends the identification of feed paths that will transport feed metal through castings of complex geometry, such as the hot spots at the T-junctions between plates (Figure 10.26). (He also draws attention to the fact that certain locations are never feed paths. These include corners or edges of plates or the ends of bars and cylinders.)

The various ways to help to ensure that feed paths remain open are considered in this section.

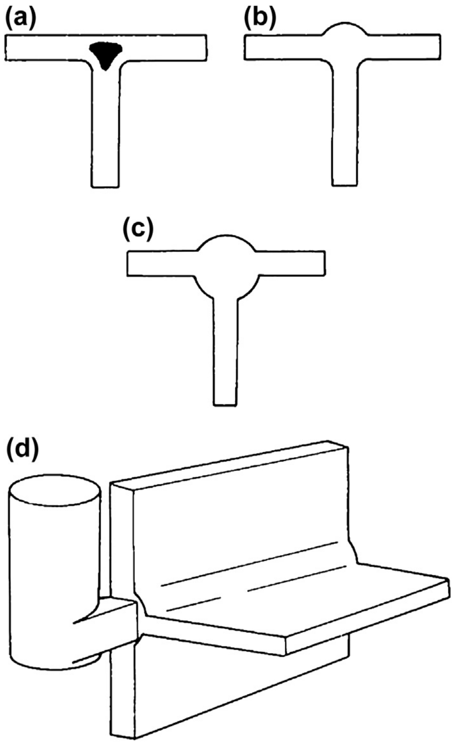

FIGURE 10.26

(a) T-junction with normal concave fillet radius; (b) a marginal improvement to the feed path along the junction; (c) convex fillets plus pad that doubles feeding distance along the junction; and (d) practical utilisation of a T-junction as a feed path (Mertz and Heine, 1973).

Directional solidification towards the feeder

If the feeder can be placed on the thickest section of the casting, with progressively thinner sections extending away, then the condition of progressive solidification towards the feeder can usually be achieved.

A classical method of checking this according to Heuvers can be used in which circles are inscribed inside the casting sections. If the diameters of the circles increase progressively towards the feeder, then the condition is met (Figure 10.27). Lewis and Ransing (1998) draw attention to the fact that Heuvers' technique is only two-dimensional, and that the condition would be more accurately represented in three dimensions by a progressive change in the radius of a sphere, effectively equivalent to the progressive increase in casting modulus towards the feeder. The fundamental reason for tapering the casting in this way is to achieve taper in the liquid flow path (Sullivan et al., 1957). For convenience, we shall call this the modulus gradient technique.

Failure to provide sufficient modulus gradient towards the feeder can be countered in various ways by (1) either re-siting the feeder (2) providing additional feeder(s) or (3) modifying the modulus of the casting. Ransing describes a further option, (4) in which he proposes a change in heat transfer coefficient. The latter technique is a valuable insight because it is easily and economically computed by a geometrical technique, and so contrasts with the considerable computing reserves and effort required by finite element and finite difference methods. If R is the radius of the inscribed sphere, the local solidification time t is proportional to R^2/h where h is the heat transfer coefficient at the metal/mould

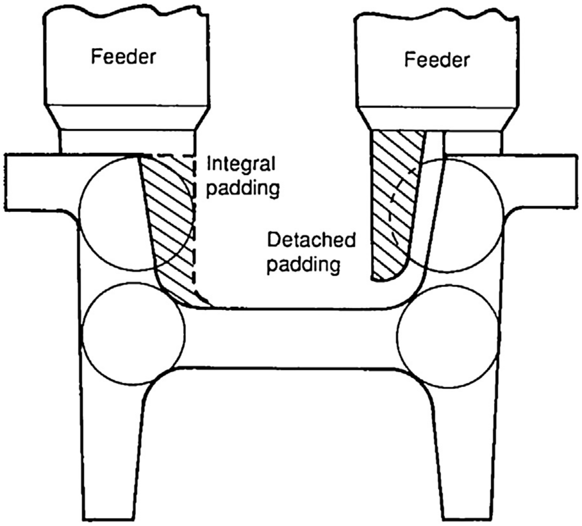

FIGURE 10.27

Use of Heuvers circles to determine the amount of added padding (Beeley, 1972) and the use of detached (or indirect) padding described by Daybell (1953).

interface. Thus at locations 1 and 2 we have $h_2 = h_1(R_1/R_2)^2$. This relation allows an estimate of the change of h that is required to ensure that the freezing time increases steadily towards the feeder. Ransing uses a change of 10% increase of solidification time for each different geometrical section of the casting (i.e. for feeding via a thin section to a thick distant section he increases the freezing time of the thin section over that of the thick by 10%). The technique can be usefully employed in reverse, in the sense that known values of h produced by a chill can quickly be checked for their effectiveness in dealing with an isolated heavy section. In every case, the target is to eliminate the local hot spot and ensure a continuous feed path back to the feeder. This simple technique has elegance, economy and power and is strongly recommended.

The casting modulus can be modified by providing either a chill or cooling fin or pin to speed solidification locally, or by providing extra metal to thicken the section and delay solidification locally. The provision of extra metal on the casting is known as padding. The addition of padding is most usefully carried out with the customer's consent, so that it can be left in position as a permanent feature of the casting design. If consent cannot be obtained, then the caster has to accept the cost penalty of dressing off the padding as an additional operation after casting.

Occasionally, this problem can be avoided by the provision of detached, or indirect, padding as shown in Figure 10.27. Daybell (1953) was probably the first to describe the use of this technique. The author has found it useful in the placement of feeders close to thin adjacent sections of casting, with a view to feeding through the thin section into a remote thicker section.

The principle of progressive increase of modulus towards the feeder, although generally accurate and useful, is occasionally seen to be not quite true. Depending on the conditions, this failure of the principle can be either a problem or a benefit, as shown later. (Even so, the Ransing technique described previously is unusual because it successfully takes this problem into account.)

In a re-entrant section of a casting, the confluence of heat flow into the mould can cause a hot spot, leading to delayed solidification at this point and the danger of local shrinkage porosity in an alloy that shrinks during freezing, such as an Al alloy. Alternatively, in an alloy that expands such as a high carbon-equivalent cast iron, the exudation of residual liquid into the mould as a result of the high internal pressure created during the precipitation of graphite can cause penetration of the aggregate mould material, with unwelcome so-called *burned-on* sand in a particularly inaccessible region. Such a hot spot can occur despite an apparent unbroken increase in modulus through that region towards the feeder. This is because the simple estimation of modulus takes no account of the geometry of heat flow away from cooling surfaces; all surfaces are assumed to be equally effective in cooling the casting. Such a hot spot requires the normal attention such as extra local cooling by chill or fin or additional feed via extra padding or feeders.

The failure of the modulus gradient technique can be used to advantage in the case of feeder necks to reduce the subsequent cutoff problem. Feeders are commonly joined to the main casting via a feeder neck, with the modulus of the neck commonly controlled to be intermediate between that of the casting and the feeder; the moduli of casting, neck and feeder are in the ratio 1.0:1.1:1.2 (Beeley, 1972). However, the neck can be reduced considerably below this apparently logical lower limit because of the hot spot effect and because of the conduction of heat from the neighbouring casting and feeder that helps to keep the neck molten for a longer period than its modulus alone would suggest. This point is well illustrated by Sciama (1975); his results are summarised in Figure 10.28. The results clearly demonstrate that for steel, feeder necks can be reduced to half of the diameter D of the feeder, providing that they are not longer than $0.1D$. The higher thermal conductivity copper- and aluminium-based materials can have necks almost twice as long without problems.

By extrapolation of these results towards smaller neck sizes, it seems that a feeder neck in steel can be only $0.25D$ in diameter, providing it is no more than approximately $0.03D$ in length. Similarly, for copper and aluminium alloys the $0.25D$ diameter neck can be up to $0.06D$ long. These results explain the action of the Washburn core, or breaker core, which is a wafer-thin core with a narrow central hole, and which is placed at the base of a feeder, allowing it to be removed after casting by simply breaking it off. In separate work the dimensions of typical Washburn cores is recommended to be a thickness of $0.1D$ and a central hole diameter of $0.4D$ (work by Wlodawer summarised by Beeley, 1972). The hole size and thickness appear to be very conservative in relation to Sciama's work. However, Sciama may predict optimistic results because he uses a feeder of nearly 1.5 times the modulus of casting which would tend to keep the

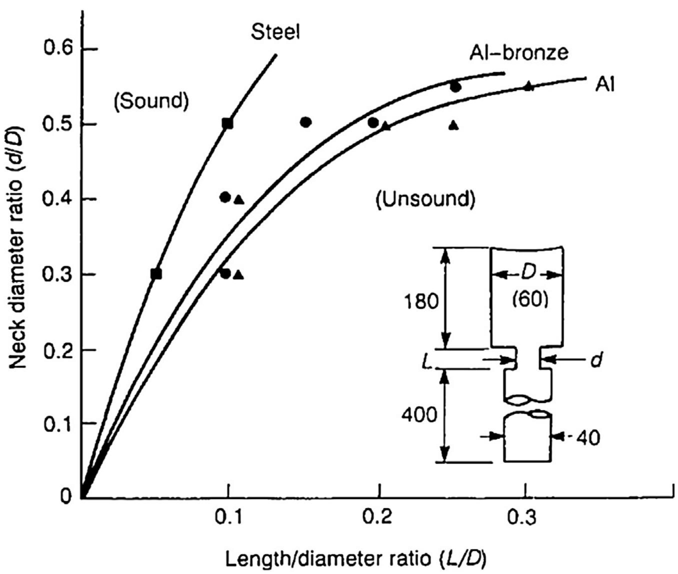

FIGURE 10.28

Effect of a constricted feeder neck on soundness of steel, aluminium bronze, and 99.5Al castings. The experimental points by Sciama (1975) denote marginal conditions.

junction rather hotter than a feeder with a modulus of only 1.2 times that of the casting (it would be valuable to repeat this work using a more economical feeder). Also, of course, conservatism may be justified where feeding conditions are less than optimum for other reasons in the real foundry environment.

The aspect of conservatism because of the real foundry environment is an interesting issue. For instance, the feeder neck could, in the extreme theoretical case, be of zero diameter when the thickness of the feeder neck core was zero. The reductio ad absurdum argument illustrates that an extreme is not worth targeting, especially when sundry debris in suspension, such as rather rigid bifilms, could close off a narrow aperture.

In fact, this danger appears to be very real for metals that commonly suffer a high population of bifilms, such as the light alloys. During the rapid filling of the mould, the bifilms swarm into the feeder, propelled by the momentum of the flow. However, during feeding, the liquid in the feeder drifts into the feeder neck, causing the gradual accumulation of bifilms, piling them up at a narrowed exit. Feeding may then be prevented by the simple mechanical action of the lowest bifilm opening, its low half expanding away into the casting as the metal in the casting attempts to suck down feed liquid. At the same time, the pileup of bifilms above the feeder exit may be sufficiently rigid to support the weight of metal in the feeder, thus preventing feeding. The eventual outcome is the mysterious case of the full feeder with a large shrinkage cavity immediately underneath. The pompous 'underriser shrinkage porosity' jargon now no longer means 'the feeder was too small', but 'the melt was too dirty to allow feeding'.

Minimum temperature gradient requirement

Experiments on cast steels have found that when the temperature gradient at the solidus (i.e. the temperature gradient at which the final residual liquid freezes) falls to below approximately 0.1–1 Kmm^{-1} then porosity is observed even in well-degassed material. Although there is much scatter in other experimental determinations, it seems in general that the corresponding gradients for copper alloys are around 1 Kmm^{-1} and those for aluminium alloys are around 2 Kmm^{-1} (Pillai et al., 1976). It seems therefore that the temperature gradient defines a critical threshold of a non-feeding condition. As the flow channel nears its furthest extent and becomes vanishingly narrow, it will become subject to small random fluctuations in temperature along its length. This kind of temperature 'noise' will occur as a result of small variations in casting thickness, or of density of the mould, thickness of mould coating, blockage or diversion of heat flow direction by random entrained films etc. Thus the channel will not reduce steadily to infinite thinness, but will terminate when its diameter becomes close to the size of the random perturbations.

There has been some discussion about the absolute value of the critical gradient for feeding on the grounds that the degree of degassing, or the standard of soundness, to which the casting was judged, will affect the result. These are certainly very real problems, and do help to explain some of the wide scatter in the results.

Hansen and Sahm (1988) draw attention to a more fundamental objection to the use of temperature gradients as a parameter that might correlate with feeding problems. They indicate that the critical gradient required to avoid shrinkage porosity in a steel bar is five to 10 times higher than that required for a plate, and point to other work in which the critical gradient in a cylindrical steel casting is a function of its diameter. Thus the concept of a single gradient which applies in all conditions seems too simplistic. It seems the rate of flow of the residual feeding liquid may also be important.

Feeding distance

It is easy to appreciate that in normal conditions it is to be expected that there will be a limit to how far feed liquid can be provided along a flow path. Up to this distance from the feeder the casting will be sound. Beyond this distance the casting will be expected to exhibit porosity.

This arises because, along the length of a flow channel, the pressure will fall progressively because of the viscous resistance to flow (Section 7). When the pressure falls to a critical level, which might actually be negative (simply becoming a tensile stress in the liquid), then porosity may form. Such porosity may occur from an internal initiation event (such as the opening of a bifilm) or from the drawing inwards of feed metal from the surface of the casting because this may now represent a shorter and easier flow path than supply from the more distant feeder.

There has been much experimental effort to determine feeding distances. The early work by Pellini and coworkers (summarised by Beeley (1972)) at the US Naval Materials Laboratory is a classic investigation that has influenced the thinking on the concept of feeding distance ever since. They discovered that the feeding distance L_d of plates of carbon steels cast into greensand moulds depended on the section thickness T of the casting: castings could be made sound for a distance from the feeder edge of $4.5T$. Of this total distance, $2.5T$ resulted from the chilling effect of the casting edge; the remaining $2.0T$ was made sound by the feeder. The addition of a chill was found to increase the feeding distance by a fixed 50 mm (Figure 10.29). They found that increasing the feeder size above the optimum required to obtain this feeding distance had no beneficial effects in promoting soundness. The feeding distance rule for their findings is simply:

$$L_d = 4.5T \tag{10.7}$$

Pellini and colleagues went on to speculate that it should be possible to ensure the soundness of a large plate casting by taking care that every point on the casting is within a distance of $2.5T$ from an edge, or $2.0T$ from a feeder.

Note that all the semi-empirical computer programs written since have used this and the associated family of rules, as illustrated in Figure 10.29, to define the spacing of feeders and chills on castings. However, the original data relate only to steel in greensand moulds, and only to rather heavy sections ranging from 50 to 200 mm. Johnson and Loper (1969) have extended the range of the experiments down to section a thickness of 12.5 mm and have re-analysed all the data. They found that for plates, the data, all in units of millimetres, appeared to be more accurately described by the equation:

$$L_d = 72m^{1/2} - 140 \tag{10.8}$$

and for bars:

$$L_d = 80m^{1/2} - 84 \tag{10.9}$$

where m is the modulus of the cast section in millimetres. The revised equations by Johnson and Loper have usually been overlooked in much subsequent work. What is also overlooked is that all the relations apply to cast mild steels in greensand moulds, not necessarily to any other casting alloys in any other kinds of mould.

In their nice theoretical model, Kubo and Pehlke (1985) find support for Pellini's feeding distance rules for steel castings, but it is a concern that no equivalent rule emerged for Al-4.5Cu alloy that they also investigated.

In fact, colleagues of Flinn (1964) found that whereas the short-freezing-range alloys manganese bronze, aluminium bronze and 70/30 cupro-nickel all had feeding distances that behaved like steel, increasing with section thickness, the long-freezing-range alloy tin bronze appeared to react in the opposite sense, giving a reduced feeding distance as section thickness increased. (The nominal composition of this classical long-freezing-range material is 85Cu, 5Sn, 5Zn and 5Pb. It was known among traditional foundrymen as 'ounce metal' because, to make this alloy, they needed to take 1 lb of copper to which 1 oz of tin, 1 oz of zinc and 1 oz of lead was added. This gives, allowing for small losses on addition, the ratios 85:5:5:5.)

Kuyucak (2002) reviews the relations for estimating feeding distance in steel castings, and finds considerable variation in their predictions. This makes sobering reading.

Jacob and Drouzy (1974) found long feeding distances, greater than $15T$, for the relatively long-freezing-range aluminium alloys Al-4Cu and Al-75i-0.5Mg, providing the feeder is correctly sized.

All this confusion regarding feeding distances remains a source of concern. We can surmise that the opposite behaviour of short- and long-freezing-range materials might be understood in terms of the ratio pasty zone/casting section. For short-freezing-range alloys, this ratio is less than 1, so the solidified skin of the alloy is complete, dictating feeding from the feeder, and thus normal feeding distance concepts apply.

For the case of long-freezing-range materials where the pasty zone/casting section ratio is greater than 1, and in fact might be 10 or more, the outer solid portions of the casting are far from solid for much of the period of solidification. The connections of liquid through to the outer surface will allow flow of liquid from the surface to feed solidification shrinkage if flow from the more distant feeder becomes more difficult.

In addition, the higher temperature and lower strength of the liquid/solid mass will allow general collapse of the walls of the casting inwards, making an important contribution to the feeding of the inner regions of the casting by the 'solid feeding' mechanism. It is for this reason that the higher conductivity and lower strength alloys of Al and Cu can be characterised by practically infinite feeding distances, particularly if the alloys are relatively free from bifilms. For clean metal, internal porosity simply does not nucleate, no matter how distant the casting happens to be from the feeder; the outer walls of the casting simply move inwards very slightly.

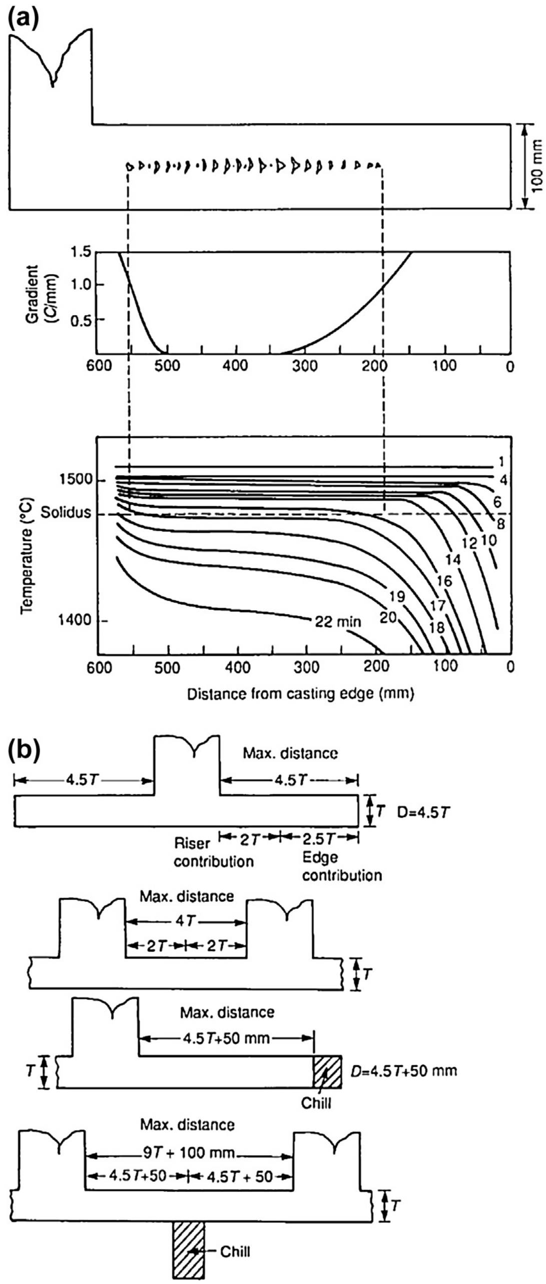

FIGURE 10.29

The famous results by Pellini (1953) for (a) the temperature distribution in a solidifying steel bar and (b) the feeding distances for steel plates cast in greensand.

Thus, although the general concept of feeding distance is probably substantially correct, at least for short-freezing-range alloys, and particularly for stronger materials such as steels, it should be used, if at all, with great caution for non-ferrous metals until it is better understood and quantified. In summary, it is worth noting the following:

1. The data on feeding distances have been derived from extensive work on carbon steels cast in greensand moulds. Relatively little work has been carried out on other metals in other moulds.
2. The definition of feeding distance is sensitive to the level of porosity that can be detected and/or tolerated.
3. It is curious that the feeding distance is defined from the edge of a feeder (not its centreline).
4. The quality of the cast metal in terms of its gas and oxide content would be expected to be crucial. For instance, good quality metal achieved by the use of good degassing and casting technique, possibly using filters (i.e. having the benefit of a low bifilm content) would be expected to yield massive improvements in feeding distance. This has been demonstrated by Romero et al. (1991) for Al-bronze. Berry and Taylor (1999) report a related effect, whilst reviewing the benefit to the feeding distance of pressurising the feeder. This work is straightforwardly understood in terms of the pressure on the liquid acting to suppress the opening of bifilms.

A final note of caution relates to the situation in which the concept of feeding distance probably does apply to an alloy, but has been exceeded. When this happens, it is reported that the sound length is considerably less than it would have been if the feeding distance criterion had just been satisfied. If true, this behaviour may result from the spread of porosity, once initiated, into adjacent regions. The lengths of sound casting in Figure 10.29(a) are considerably shorter than the maximum lengths given by Eqns (10.7) to (10.9), possibly because the feeding distance predicted by these equations has been exceeded and the porosity has spread. Mikkola and Heine (1970) confirm this unwelcome effect in white iron castings.

Criteria functions

For a discussion of the use of criteria functions to assess the difficulty to feed and the propensity of the metal to develop porosity, please refer to the earlier Section 7.1.3.2.

Feeding rule 6: Pressure gradient requirement

Although all of the previous feeding rules may be met, including the provision of feed liquid and a suitable flow path, if the pressure gradient needed to cause the liquid to flow along the path is not available, feed liquid will *not* flow to where it is needed. Internal porosity may therefore occur.

One of the most usual causes of failure to feed is not taking advantage of gravity. As opposed to *filling uphill* (which is, of course, quite correct), *feeding* should only be carried out *downhill* (using the assistance of gravity).

Attempts to feed uphill, although possible in principle, are unreliable in practice, and may lead to randomly occurring defects that have all the appearance of shrinkage porosity. In castings of a modest size, feeding uphill appears in general to be successful using the technique 'active feeding' as will be discussed later. In many castings, particularly larger castings, problems occur when attempting to feed uphill because of the difficulties caused by various effects: (1) adverse pressure gradient as discussed later; (2) danger of upward floating of gas or air bubbles into the casting and (3) adverse density gradient leading to convection as dealt with in rule 7 of the 10 Rules.

A positive pressure gradient from the outside to the inside of the casting will help to ensure that the feed material (either solid or liquid) travels along the flow path into those parts of the casting experiencing a shrinkage condition. The various feeding mechanisms are seen to be driven by the positive pressure such as atmospheric pressure and/or the pressure due to the hydrostatic head of metal in the feeder. The other contributor to the pressure gradient, the driving force for flow, is the reduced or even negative pressure generated within poorly fed regions of some castings. All of these driving forces happen to be additive; the flow of feed metal is caused by being pushed from the outside and by being pulled from the inside.

Figure 10.30 illustrates the feeding problems in a complicated casting. The casting divides effectively into two parts either side of the broken line. The left-hand side has been designed to be fed by an open feeder Fl and a blind feeder F2. The right-hand side was intended to be fed by blind feeder F3.

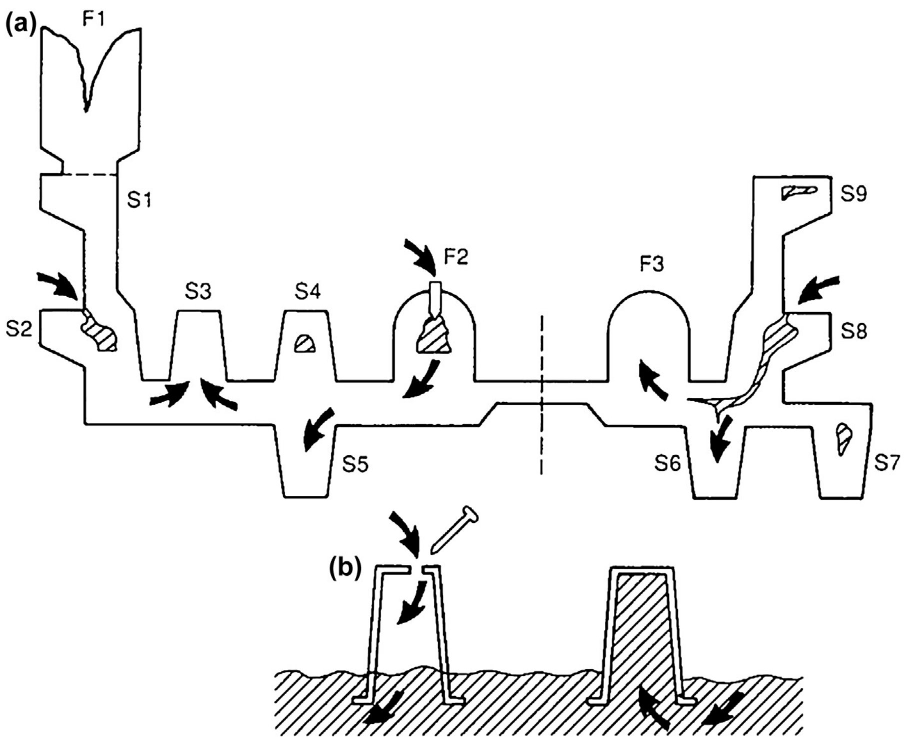

FIGURE 10.30

(a) Castings with blind feeders; F2 is correctly vented but has mixed results on section S3 and S4. Feeder F3 is not vented and therefore does not feed. The unfavourable pressure gradient draws liquid from a fortuitous skin puncture in section S8. See text for further explanation. (b) The plastic coffee cup analogue: the liquid is held in the upturned cup and cannot be released until air is admitted via a puncture. The liquid it is holding is then immediately released.

Feeder Fl successfully feeds the heavy section S1. This feeder is seen to be comparatively large. This is because it is required to provide feed metal to the whole casting during the early stages of freezing, whereas the connecting sections remain open. At this early stage, the top feeder is feeding the whole casting and both blind feeders.

Feeder F2 feeds S5 because it is provided with an atmospheric vent, allowing the liquid to be pressurised by the atmosphere as in the plastic coffee cup experiment illustrated in Figure 10.30, so forcing the metal through into the casting. (The reader is encouraged to try the coffee cup experiment.)

The identical heavy sections S3 and S4 show the unreliability of attempting to feed uphill. In S4, a chance initiation of a pore has created a free liquid surface, and the internal gas pressure within the casting happens to be close to 1 atm. Thus the liquid level in S4 falls, finding its level equal to that in the feeder F2. The surface-initiated pore in S2 has grown from the hot spot that has weakened and broken through the corner of the casting. The pore has grown, equalising its level exactly with that in the feeder F2 because both surfaces are subject to the same atmospheric pressure. In section S3, by good fortune, no pore initiation site is present, so no pore has occurred, with the result that atmospheric pressure via F2 (and unfortunately also via the puncture by the atmosphere at the hot spot in the re-entrant section S2) will feed solidification shrinkage here, causing the section to be perfectly sound.

Turning now to the right-hand part of the casting, although feeder F3 is of adequate size to feed the heavy sections S6, S7 and S8; unfortunately, its atmospheric vent has been forgotten. This is a serious mistake. The plastic coffee cup experiment shows that such an inverted airtight container cannot deliver its liquid contents. The pressure gradient is now reversed, causing the flow to be in the wrong direction, from the casting to the feeder! The detailed reasoning for this is as follows. The pressure in the casting and feeder continues to fall as freezing occurs until a pore initiates, either under the hydrostatic tension, or because of a buildup of gas in solution, or because of the inward rupture of the surface at a weak point such as the re-entrant angle in section S8. The pressure in section S8 is now raised to atmospheric pressure whilst the pressure in the feeder F3 is still low, or even negative. Thus feed liquid is now forced to flow from the casting into the feeder as freezing progresses. A massive pore then develops because feeder F3 has a large feed requirement, and drains section S8 and the surrounding casting. The defect size is worse than that which would have occurred if no feeder had been used at all!

Section S6 remains reasonably sound because it has the advantage of natural drainage of residual liquid into it. Effectively it has been fed from the heavy section S8. The pressure gradient resulting from the combined actions of gravity, shrinkage and the atmosphere from S8 to S6 is positive. The only reason why S6 may display any residual porosity may be that S8 is a rather inadequate feeder in terms of either its thermal requirement or its volume, or because the feed path may be interrupted at a late stage.

Section S7 cannot be fed because there is no continuous feed path to it. S9 is similarly disadvantaged. This has been an oversight in the design of the feeding of this casting. In a sand casting, it is likely that S7 and S9 will therefore suffer porosity. This will be almost certainly true for a steel casting, but less certain if the casting is a medium-freezing-range aluminium alloy. The reason becomes clear when we consider an investment casting, where, if a high mould temperature is chosen, and if the metal is clean, will allow solid feeding to operate, allowing the sections the opportunity to collapse plastically, and so become internally sound, provided that no pore-initiation event interrupts this action. Solid feeding is often seen in aluminium alloy sand castings, but more rarely in steel sand castings because of the greater rigidity of the solidified steel, which successfully resists plastic collapse in cold moulds.

If feeding had taken place with the assistance of gravity, feeder F2 in Figure 10.30 would have successfully fed sections S2 and S4 either if it was taller, or if it had been placed at a higher location, for instance on the top of S4.

It is clear that F3 may not have fed section S8 if the corner puncture occurred, even if it had been provided with an effective vent, because the pressure gradient for flow would have been removed. A provision of an effective vent, and the re-siting of the base of the feeder F3 to the side of S8, would have maintained the soundness of both S6 and S8 and would have prevented the surface puncture at S8. S7 and S9 would still have required separate treatment.

The exercise with the plastic coffee cup shows that the water will hold up indefinitely in the upturned cup until released by the pin causing a hole. The cup will then deliver its contents immediately (but not before!). Blind feeders are therefore often unreliable in practice because the atmospheric vent may not open reliably. Such feeders then act to suck feed metal from the casting, making any porosity worse.

If a blind feeder is provided with an effective atmospheric vent, then the available atmospheric pressure may help it to feed uphill. The atmosphere is capable of holding up several metres of head of metal. For liquid mercury the height is approximately 760 mm, being the height of the old-fashioned atmospheric barometer of course. Equivalent heights for other liquid metals are easily estimated allowing for the density difference. Thus for liquid aluminium of specific gravity 2.4 compared with liquid mercury of 13.9, the atmosphere will hold up about $(13.9/2.4) \times 0.76 = 4$ m of liquid aluminium or $(13.9/7.0) \times 0.76 = 1.5$ m liquid iron.

Summarising the maximum heights supportable by 1 atm for various pure liquids near their melting points:

Mercury	0.760 m (the barometric height 760 mm = 1 atm)
Steel	1.48 m
Zinc	1.58 m
Aluminium	4.36 m
Magnesium	6.54 m
Water	10.40 m

Whilst no pore exists, the tensile strength of the liquid will in fact allow the metal, in principle, to feed to heights of kilometres because in the absence of defects the liquid can withstand tensile stresses of thousands of atmospheres. The liquid can, in principle, hang up in a tube, its great weight stretching its length somewhat. However, the random initiation of a single minute pore will instantly cause the liquid to ‘fracture’, causing the liquid in the tube to fall, finally stabilising at the level at which atmospheric pressure can support the liquid. Thus any height above that supportable by 1 atm is clearly at high risk.

Moreover, there is even worse risk. If when attempting to feed against gravity, there is a leak path to atmosphere, allowing atmospheric pressure to be applied in the liquid metal inside the solidifying casting, the melt will then fall further, the action of gravity tending to equalise levels in the mould and feeder. Thus, if the feeder is sited below the casting, the casting will completely empty of residual liquid. Regrettably, this is an efficient way to cast feeders of excellent soundness, but seriously porous castings.

Clearly, the initiation of a leak path to atmosphere (via a double oxide film or a liquid region in contact with the surface at a hot spot) is rather easy in many castings, making the whole principle of uphill feeding so risky that it should not be attempted in circumstances where porosity cannot be tolerated. It is a pity that the comforting theories of pushing liquid uphill by atmospheric pressure or even hanging it from vast heights using the huge tensile strength of the liquid cannot be relied upon in practice.

For all practical purposes, the only really reliable way to feed is *downhill*, using gravity.

The conclusion to these considerations is: *place feeders high to feed downhill*. This is a general principle of great importance. It is of similar weight to the general principle discussed previously, *place ingates low to fill uphill*. These are fundamental concepts in the production of good castings.

Feeding rule 7: Pressure requirement

The final rule for effective feeding is a necessary requirement like all the others. Sufficient pressure in the residual liquid within the casting is required to suppress both the initiation and the growth of cavities from both internal and external sources.

This is a *hydrostatic* requirement relating to the suppression of porosity and contrasts with the previous pressure gradient requirement that relates to the *hydrodynamic* requirements for flow (especially flow in the *correct* direction!)

A fall in internal pressure may cause a variety of problems:

1. Liquid may be sucked from the surface. This is particularly likely in long-freezing-range alloys, or from re-entrant angles in shorter-freezing-range alloys, resulting in internal porosity initiated from, and connected to, the outside.
2. The internal pressure may fall just sufficiently to unfurl, but not fully open the population of bifilms. The population of bifilms will therefore remain effectively invisible. The result will be an apparently sound casting but, inexplicably, the mechanical properties will be poor, particularly the elongation and fatigue performance. (It is possible that some so-called ‘*diffraction mottle*’ may be noted on X-ray radiographs as a result of the newly appearing population of slightly opened cracks.)
3. The internal pressure may fall sufficiently to open bifilms, so that a distribution of fine and dispersed microporosity will appear. The mechanical properties will be even lower. (For the technically punctilious, the mechanical opening of bifilms is driven not so much by pressure reduction, a *stress* phenomenon, but by the progression of shrinkage, a *strain* phenomenon experienced by the liquid.)
4. The internal shrinkage may cause macro-shrinkage porosity to occur, especially if there are large bifilms present as a result of poor filling of the casting. Properties may now be in disaster mode and/or large holes may appear in the casting. Figure 10.31 illustrates pressure loss situations in castings that can result in shrinkage porosity.
   Figure 10.32 illustrates the common observation in Al alloy castings in which a glass cloth is placed under the feeder to assist the break-off of the feeder after solidification. Bafflingly, it sometimes appears that the cloth prevents the feeder from supplying liquid, so that a large cavity appears under the feeder. The truth is that double oxide films plaster themselves against the underside of the cloth as the feeder fills, keeping the bifilms in the casting, and ensuring that the feeder contains clean metal. The half of the bifilm against the cloth tends to weld to the cloth, possible by a chemical fusing action, or possibly by mechanical wrapping around the fibres of the cloth. Whatever the mechanism, the action is to hold back the liquid above, whereas the lower half of the (unbonded) double film is

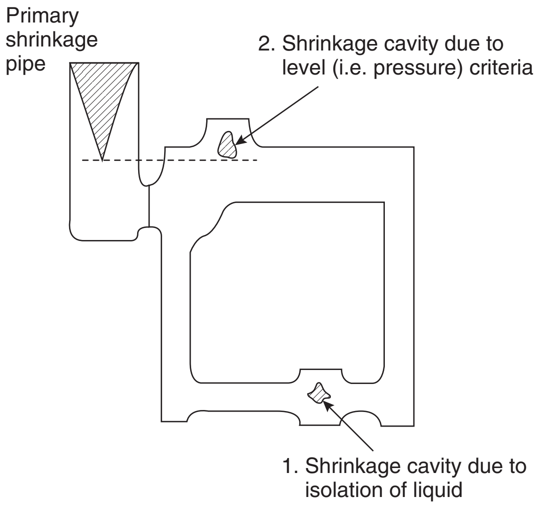

FIGURE 10.31

Pressure loss situations in castings leading to the possibility of shrinkage porosity.

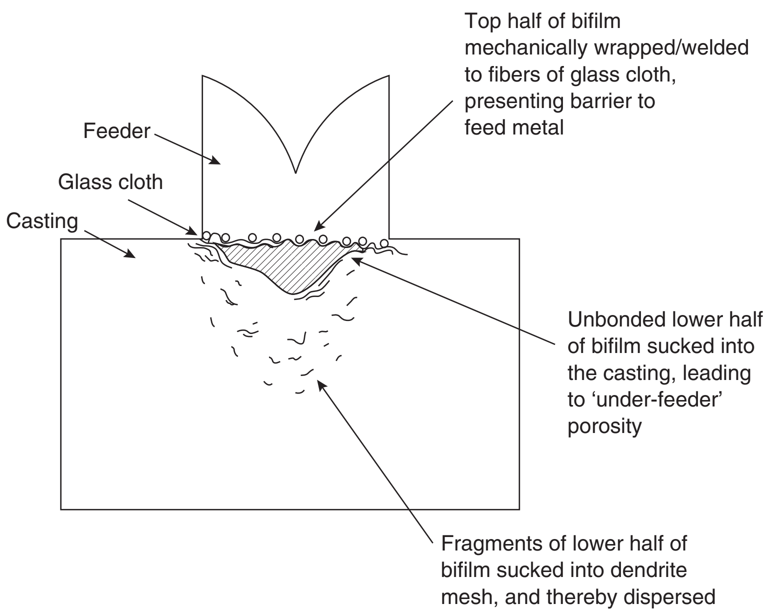

FIGURE 10.32

Apparent blocking of feed metal by a glass cloth strainer in an Al alloy casting because of bifilms collected on the underside of the cloth.

easily pulled away by the contracting liquid, opening the void that was originally the microscopically thin interface of air inside the bifilm. Once again, bifilms are seen to interfere with the action of a feeder, and create the curious phenomenon of shrinkage porosity immediately under a feeder which remains completely full of (clean) metal.

5. If there is insufficient opportunity to open internal defects, the external surface of the casting may sink to accommodate the internal shrinkage. (The occurrence of *surface sinks* is occasionally referred to elsewhere in the literature as '*cavitation*'; a misuse of language to be deplored. Cavitation properly refers to the creation and collapse of minute bubbles, and the consequent erosion of solid surfaces such as those of ships' propellers, although, I admit, it is beginning to have a tolerably respectable use to describe the opening of pores during super plastic forming.)

Often, of course, the distribution of defects observed in practice is a mixture of the previous list. The internal pressure needs to be maintained sufficiently high to avoid all of these defects.

Finally, however, it is worth pointing out that over-zealous application of pressure to reduce the previous problems can result in a new crop of different problems.

For instance, in the case of long-freezing-range materials cast in a sand mould, a high internal pressure, applied for instance to the feeder, will force liquid out of the surface-linked capillaries, making a casting having a 'furry' appearance. Overpressures are not easy to control in low-pressure sand casting processes, and are the reason why these processes often struggle to meet surface-finish requirements. Work at the University of Alabama (Stefanescu et al., 1996) on cast iron castings has shown that the generation of excess internal pressure by graphite precipitation can lead to exudation of the residual liquid via hot spots at the surface of the casting, leading penetration of the sand mould. Later work on steels has shown analogous effects (Hayes et al., 1998).

In short-freezing-range materials the inward flow of solid can be reversed with sufficient internal pressure. Too great a pressure will expand the casting, blowing it up like a balloon, producing unsightly swells on flat surfaces (Figure 10.33).

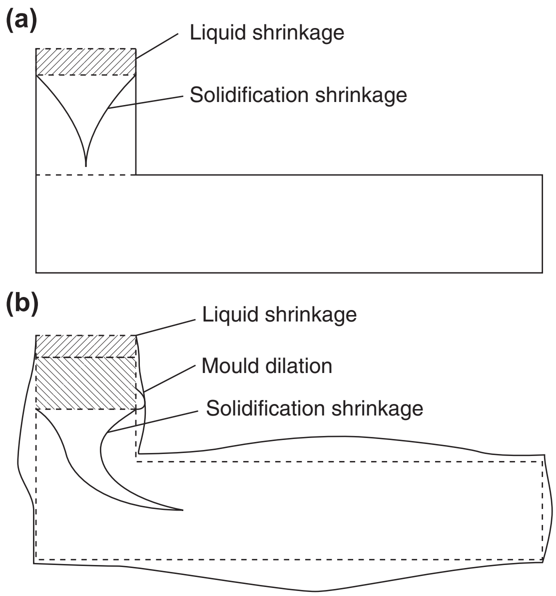

FIGURE 10.33

Comparison between the external size and internal shrinkage porosity in a casting as a result of (a) moderate pressure in the liquid, and adequately rigid mould and (b) too much pressure and/or a weak mould.

Feeding of cast iron castings by control of pressure

The successful feeding of cast irons is perhaps the most complex and challenging feeding task compared to all other casting alloys as a result of the curious and complicating effect of the expansion of graphite during freezing. The effects are most dramatically seen for ductile irons.

The great prophet of the scientific feeding of ductile irons was Stephen Karsay. In a succession of engaging and chattily written books, he outlined the principles that applied to this difficult metal (see, for instance, Karsay, 1992, 2000). He drew attention to the problem of the swelling of the casting in a weak mould as shown in Figure 10.33, in which the valuable expansion of the graphite was lost by enlarging the casting, causing the feeder to be inadequate to feed the increased volume. He promoted the approach of making the mould more rigid, and so better withstanding stress, and at the same time reducing the internal pressure by providing feeders that acted as pressure relief valves. The feeder, after some initial provision of feed metal during the solidification of austenite, would reverse its action, back-filling with residual liquid during the expansion of the solidification of the eutectic graphite, relieving the pressure and thus preventing mould dilation. The final state was that *both* the casting and feeder were substantially sound. (Occasionally, one hears stories that such sound feeders have been declared to be evidently useless, having apparently provided no feed metal. However, their removal would immediately cause all subsequent castings to become porous!)

The reproducibility of the achievement of soundness in ductile iron castings is, of course, highly sensitive to the efficiency of the inoculation treatment because the degree of expansion of graphite is directly affected. This is notoriously difficult to keep under good control, and makes for one of the greatest challenges to the iron founder.

Roedter (1986) introduced a refinement of Karsay's pressure relief technique in which the pressure relief was limited in extent. Some relief was allowed, but total relief was prevented by the premature freezing of the feeder neck. In this way, the casting was slightly pressurised, elastically deforming the very hard sand mould, and the surrounding steel moulding box (if any). The elastic deformation of the mould and its box would store the strain energy. The subsequent relaxation of this deformation would continue to apply pressure to the solidifying casting during the remainder of solidification. Thus soundness of the casting could be achieved, but without the danger of unacceptable swells on extensive flat surfaces. Even so, the use of strain energy is clearly seen to be vulnerable, because it can only reverse only a very small amount of deformation as is explained later.

For somewhat heavier ductile iron castings, however, it has now become common practice to cast completely without feeders. This has been achieved by the use of rigid moulds, now more routinely available from modern greensand moulding units. Naturally, the swelling of the casting still occurs because, ultimately, solids are incompressible. However, as before, the expansion is restrained to the minimum by the elastic yielding of the mould and its container and distributed more uniformly. Thus the whole casting is a few percentages larger. If the total net expansion was 3 volume %, this corresponds to 1 linear % along the three orthogonal axes, so that from a central datum, each point on the surface of the casting would be approximately 0.5% oversize. This uniform and very reproducible degree of oversize is usually negligible. However, of course, it can be compensated, if necessary, by making the pattern 0.5% undersize.

The use of the elastic strains to re-apply pressure is strictly limited because such strains are usually limited to only 0.1 linear % or so. Thus only a total of perhaps 0.3 volume % can be compensated by this means. This is, as we have seen previously, only a fraction of the total volume change that is usual in a graphitic iron, and which permanently affects the size and dimensions of the casting. The judgement of feeder neck sizes to take advantage of such small margins is not easy.

With the steady accumulation of experience in a well-controlled casting facility, the casting engineer can often achieve such an accuracy of feeding that even such a modest gain is considered a valuable asset. Even so, the reader will appreciate that the feeding graphitic irons is still not as exact a science and still not as clearly understood as we all might wish.

10.6.4 THE NEW FEEDING LOGIC

Much of the formal calculation of feeders has been of poor accuracy because of several simplifying assumptions that have been widely used. Tiryakioglu has pioneered a new way of analysing the physics of feeding, having, in addition, the good fortune to have as a critical test the exemplary experimental data on optimum feeder sizes determined by his late father, Ergin Tiryakioglu, many years earlier (1964). This was carried out at the University of

Birmingham, UK, where Ergin and I shared an office for a time while we were both researching for our doctorates. The reader is recommended to the original papers by Murat Tiryakioğlu (1997–2002) for a complete description of his admirable logic. We shall summarise his approach only briefly here, following closely his excellent description (Tiryakioğlu et al., 2002).

As we have seen in rules 2–4, an efficient feeder should (1) remain molten until the portion of the casting being fed has solidified (i.e. the solidification time of the feeder has to be equal to, or exceed, that of the casting), (2) contain sufficient volume of molten metal to meet the feeding demand of that same portion of the casting and (3) not create a hot spot at the junction between feeder and casting. An optimum feeder is then defined as the one with the smallest volume, for its particular shape, to meet these criteria. A feeder that is less compact or that has less volume than the optimum feeder will result in an unsound casting.

The standard approaches to solve these problems have usually been based on the famous rule by Chvorinov (1940) for the solidification time of a casting is:

$$t = (V/A)^n \tag{10.10}$$

where B is the mould constant, V is the casting volume, A is the surface area through which heat is lost, and n is a constant ($n = 2$ in Chvorinov's work for simple shaped castings in silica sand moulds). The V/A ratio is known as the modulus m, and has been used as the basis for several approaches to determining the size of feeders for the production of sound castings, as described in feeding rule 2.

Despite its wide acceptance, Chvorinov's rule has limitations because of the underlying assumptions used in deriving the equation. As a result of these limitations, the exponent, n, fluctuates between 1 and 2, depending on the shape and size of the casting, and the mould and pouring conditions. One of the reasons for this anomaly is that Chvorinov's rule originally did not take the shape of the casting into consideration. A new geometry-based model (Tiryakioğlu et al., 1997a) proved that the modulus includes the effect of both casting shape and size. These two independent factors were separated from each other by the use of a shape factor k where

$$t = B' k^{1.31} V^{0.67} \tag{10.11}$$

and B' is the mould constant. The shape factor, k, is the ratio (the surface area of a sphere of same volume as the casting)/(the surface area of the casting). In Eqn (10.11), V assesses the amount of heat that needs to be dissipated for complete solidification, and k assesses the relative ability of the casting shape to dissipate the heat under the given mould conditions.

$$k = A_s/A = 4.837V^{2/3}/A \tag{10.12}$$

where A_s is the surface area of the sphere.

Adams and Taylor (1953) were the first to consider *mass transfer* from feeder to casting. They realised that during solidification, a mass of αV_c needs to be transferred from the feeder to the casting (α is fraction shrinkage of the metal). However, as Tiryakioğlu (2002) explains, their development of the concept unfortunately introduced errors so that the final solution was not accurate.

Moreover, the lack of knowledge about the effect of *heat transfer* between a feeder and a casting has led the researchers to the mindset of considering the feeder and casting separately. In other words, almost all feeder models have been based on calculation of solidification times for the feeder and casting independently, and then assuming that the same solidification characteristics will be followed when they are combined. However, it should be remembered that the feeder is also a portion of the mould cavity, and the solidifying metal does not know (or care) which portion is the casting and which is the feeder.

The objective of the foundry engineer when designing a feeding system is to have the thermal centre of the total casting (the feeder-casting combination) in the feeder. In fact, all three requirements for an efficient feeder listed (1–3) previously can be summarised as a single requirement: *the thermal centre of the feeder-casting combination will be in the feeder*. This new approach, which treats the casting-feeder combination as a single, total casting constitutes the foundation of Tiryakioğlu's new approach to characterise heat and mass transfer between feeder and casting.

The new approach

Let us consider a plate casting that is fed effectively by a feeder. Knowing that the solidification contraction of the casting is αV_c, this volume is transferred from the feeder to the casting, resulting in the final volume of the feeder being $(V_f - \alpha V_c)$. Solidification contraction of the feeder is ignored sincebecause it does not change the heat content of the feeder. If the feeder has been designed according to the rules for efficient feeding, the last part to solidify in this combination is the feeder. In other words, the thermal centre of the casting-feeder combination (the total casting) is in the feeder. Therefore the solidification time of the total casting is exactly the same as the feeder, and both have the same thermal centre.

This is not true for the casting, however. The thermal centre of the casting is also in the feeder, but its solidification time may or may not be equal to that of the total casting. Hence

$$k_t^{1.31} V_t^{0.67} = k_f^{1.31} (V_f - \alpha Vc\,)^{0.67} \tag{10.13}$$

where subscript t refers to the total casting. So far, we have ignored the heat transfer between the casting and the feeder. Using optimum feeder data obtained by systematic changes in feeder size for an Al-12wt%Si alloy (Tiryakioğlu, 1964), the solidification times of casting and feeder are compared in Figure 10.34(a) and total casting and feeder in Figure 10.34(b). Figure 10.34(a) shows the $(t_c - t_f)$ relationship when mass transfer is taken into account and heat transfer is ignored. It should be kept in mind that the scatter in Figure 10.34(a) is not due to experimental error because all values were calculated. Figure 10.34(b) shows the relationship between the solidification times of feeder and total casting (feeder + casting combination). Although the agreement in Figure 10.34(b) is encouraging, at low values the error is up to 30%. However, solidification time should be identical for the feeder and the total casting. The error is due to the neglect of the heat exchange between feeder and casting. Mass is transferred from the feeder to the casting throughout the solidification process. Solidification takes place over a temperature range, so that subtracting αV_c from V_f adjusts for mass exchange completely, but for heat exchange this treatment assumes isothermal conditions, and therefore is not sufficient. Hence the feeder solidification time needs to be adjusted for the heat exchange with the casting. We can treat this heat exchange as if it were superheat extracted from/given to the feeder. For the superheat model, we will use the model by E. Tiryakioğlu (1964) for its simplicity and its independence from actual pouring temperature. Equation (10.13) can now be rewritten as

$$k_t^{1.31} V_t^{0.67} = k_f^{1.31} (V_f - \alpha V_c)^{0.67} e^{\xi \Delta T_f} \tag{10.14}$$

where ξ is a constant dependent on the alloy (0.0028°C^{-1} for Al-Si eutectic alloy (Tiryakioğlu, 1964) and 0.0033°C^{-1} for Al-7%Si (Tiryakioğlu et al., 1997b)), ΔT_f is the temperature change (rise or fall) in feeder because of the heat exchange, and can be easily calculated using Eqn (10.14). The sum of change in heat content of the feeder and casting is zero (heat lost by one is gained by the other). Therefore

$$C(V_f - \alpha V_c)\Delta T_f + C(V_f + \alpha V_c)\Delta T_c = 0 \tag{10.15}$$

where C is the specific heat of the metal. Hence

$$\Delta T_c = (V_f - \alpha V_c)\Delta T_f / (V_f + \alpha V_c) \tag{10.16}$$

The total solidification time of the casting can now be written as

$$t_c = k_c^{1.31} (V_f + \alpha V_c)^{0.67} e^{\xi \Delta T_c} \tag{10.17}$$

The solidification times of feeder and casting can now be compared. This comparison is presented in Figure 10.34(c) which shows a practically perfect fit and a relationship that can be expressed as;

$$t_t = t_f = \alpha t_c \tag{10.18}$$

The data for Al-Si alloy shown in Figure 10.34(c) give $\alpha = 1.046$.

In a separate exercise, using the data for steel by Bishop et al. (1955) assuming ξ of 0.0036°C^{-1} for steel (Tiryakioğlu, 1964), a similar excellent relationship is obtained where α is found to be 1.005 (Tiryakioğlu, 2002).

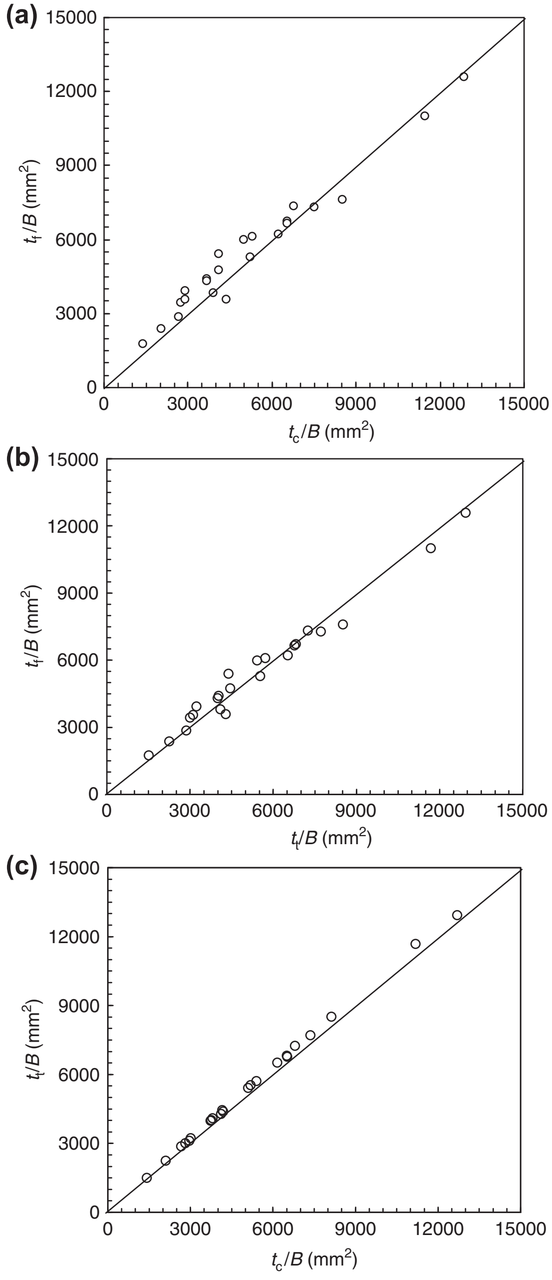

FIGURE 10.34

Comparison of calculated solidification times of (a) casting and feeder; (b) total casting and feeder; (c) total casting (or feeder) versus casting after adjustment to account for heat transfer between feeder and casting (Tiryakiolğu et al., 2002).

Thus the solidification time of optimum-sized feeders in the feeder-casting combination was found to be only a few percent longer than that of castings both for Al-Si alloy and steel castings.

We can conclude that, for an accurate description of the action of a feeder, both mass and heat transfer from feeder to casting during solidification have be taken into account simultaneously. Previous feeder models that account for mass transfer assume that the transfer takes place isothermally and at the pouring temperature. This previous assumption overestimates the additional heat brought into the casting from the feeder. The new model incorporates the effect of superheat and is based on the equality of solidification times of feeder and total casting.

The requirements for efficient feeders: (1) solidification time; (2) feed metal availability and (3) prevention of hot-spot at the junction; can be combined into a single requirement when the casting-feeder combination is treated as a single, total casting. The three criteria reduce to the simple requirement: 'The thermal centre of the total casting should be in the feeder'.

The disarming simplicity of this conclusion conceals its powerful logic. It represents the ideal criterion for judging the success of a computer model of a casting and feeder combination.

10.6.5 FREEZING SYSTEMS DESIGN (CHILLS, FINS AND PINS)

In this section on feeding, we are of course mainly concerned with the action of chills and fins to provide localised cooling of the casting. This is my first choice for avoiding the planting of a feeder on a hot spot. I always try to avoid feeding.

Chills (or fins or pins) can assist directional solidification of the casting towards the feeder, thus assisting in the achievement of soundness. Details of this thermal action are presented in Section 5.1.2 on increased heat transfer.

Chills (or fins or pins) also act to increase the ductility and strength of that locality of the casting. For some alloys, this benefit to mechanical properties results from the well-known Hall–Petch effect associated with shortening slip distances, and the difficulty of re-initiating slip in the adjacent grain. However, it is more likely that most alloys will benefit most from the freezing of bifilms in their compact state. This justification of this conclusion is outlined in Section 9.4 on the mechanical properties of castings.

10.6.6 FEEDING: THE FIVE MECHANISMS

The five feeding mechanisms—liquid, mass, interdendritic, burst and solid feeding—are different ways in which cast material can flow to feed the solidification contraction. The mechanisms are dealt with in detail in Chapter 7. In general, it is fair to state that only liquid, interdendritic and solid are important for our purposes of making sound castings. However, when relying on solid feeding for soundness there are precautions that it is wise to observe.

Dangers of solid feeding

The great benefit of solid feeding is that a casting can be made sound, even though it demands feed metal but cannot be fed by liquid from a conventional feeder. In favourable conditions, the casting can collapse plastically; the shrinkage volume is merely transferred from the inside to the outside of the casting. Here, if the volume is distributed reasonably evenly over a large area of the casting, the shrinkage will cause only a negligible and probably undetectable reduction in the size or shape of the casting.

This feeding benefit is commonly gained in light alloy castings which are relatively weak at their freezing points and so collapse rather easily; no large internal tensile stresses are generated so that pores are not necessarily created internally. Similarly, the easy collapse of weak solid is found in many ferrous and higher temperature alloys in investment casting, where the high mould temperature retains the plasticity of the solidified skin of the casting.

There are, however, dangers in relying on solid feeding for the soundness of a casting, as are illustrated in Figure 9.25.

1. If the outside shrinkage is not distributed so favourably, but remains concentrated in a local region, a surface sink is the result. This may, of course, result in a portion of the casting being out of tolerance, possibly no longer cleaning up on machining, so that the casting is scrapped.

2. When operating without feeders, a second possibility is the formation of shrinkage pores, grown from initiation sites (almost certainly bifilms), so that solid feeding immediately fails. It seems that such events tend to be triggered by rather large bifilms such as arise from random variations in pouring techniques. The achievement of soundness using solid feeding then becomes a hit-and-miss affair.
3. A further possibility will be commonly, but not easily perceived. If the melt has a distribution of small, possibly microscopic, bifilms, these will be unfurled to some extent by the reduced pressure in the unfed region thus being converted from crumpled compact features of negligible size to flat thin extensive cracks. Thus although the casting may continue to appear perfectly sound in the unfed region, and solid feeding declared to be a complete success, the mechanical properties of this part of the casting will be reduced. In particular, although the yield strength of the region will be hardly affected, that part of the casting will exhibit reduced strength, ductility and fatigue resistance.
4. If the localised shrinkage problems are even more severe, the distribution of small bifilms will develop further. After unfurling to become flat cracks, additional reduction of pressure in the liquid will open them further still to become visible microporosity. The pores may even grow to such a size that they become visible on radiographs. This situation is common in steel castings where the high strength of the solidified skin creates large internal stresses, providing irresistible forces to open bifilms and thus bring solid feeding to a halt.

In contrast to these dangers, the action of a feeder to pressurise the melt and so help to resist the unfurling and even the inflating of bifilms, is seen to be a benefit. The bifilms although still present, remain out of sight. Using a domestic analogy from home decoration, there is a very real sense in which adding feeders to castings is almost literally ‘papering over the cracks’.

However, this rather severe censure is not completely valid, in the sense that closed bifilms in a casting do exhibit some strength because they can withstand shear stress (although, of course, not tensile stress at right angles to the bifilm). The pressure provided by the (liquid) feeder has a valuable role in keeping the bifilms closed.

10.6.7 COMPUTER MODELLING OF FEEDING

Good computer models have demonstrated their usefulness in being able to predict shrinkage porosity with accuracy and have therefore brought about a revolution within the industry. A simulation using a reliable modelling software package should now be specified as a prior requirement to be carried out before work is started on making the tooling for a new casting. This minor delay has considerable benefits in shortening the overall development time of a new casting, and greatly increases the chance of being ‘right first time’.

However, the limitations of some computer simulations require to be recognised. For instance, these include

1. no allowance for the effect of thermal conduction in the cast metal (rare nowadays);
2. no allowance for the important effects due to convection in the liquid (common);
3. neglect of, or only crude allowance for, the effect of the heating of the mould by the flow of metal during filling (now commonly addressed); and
4. no capability of any design input. Thus gating and feeding designs will be required as inputs (just beginning to be addressed at this time).

For the future, it is to be expected that software packages will evolve to provide intelligent solutions to all these requirements. Examples of a good start in this direction are shown by Dantzig and coworkers (Morthland et al., 1995 and McDavid and Dantzig, 1998). In the meantime, it remains necessary to use computer models with some discretion. For instance, in the work by the Morthland team, they warn that the results are specific to the feeding criterion used. If a more stringent temperature gradient criterion were used (for instance $2 K cm^{-1}$ instead of $1 K cm^{-1}$) the feeder would have been larger.

The previous approaches to the optimisation of the feeding requirements of castings have involved the use of numerical techniques such as finite element and finite difference methods. Ransing et al. (2004) propose a geometrical method based on an elegant extension of the Heuvers circle technique. This technique was described previously in the section describing the feed path requirements for feeding.

10.6.8 RANDOM PERTURBATIONS TO FEEDING

It is helpful if the computer simulation can provide a robust casting solution. This is because on the foundry floor, several practical issues conspire to change the casting conditions, and thus threaten the quality of the casting.

In aluminium castings, flash of approximately 1 mm thickness and only 10 mm wide has been demonstrated to have a powerful effect on the cooling of local thin sections up to 10 mm thick, speeding up local solidification rates by up to 10 times. The effect is so much reduced in ferrous castings because of their much lower thermal conductivity that any effect of cooling by fins can usually be neglected.

Thus for high-conductivity alloys flash has to be controlled, or used deliberately because, in moderately thick sections, it has to potential to cut off feeding to more distant sections. The erratic appearance of flash in a production run may therefore introduce uncertainty in the reproducibility of feeding, and the consequent variability of the soundness of the casting. Flash on very thick sections is usually less serious because convection in the liquid in thick sections conveys the local cooled metal away, effectively spreading the cooling effect over other parts of the casting, giving an averaging effect over large areas of the casting. In general, however, it is desirable that these uncertainties are reduced by good control over mould and core dimensions.

The other known major variable affecting casting soundness in sand and investment castings is the ability of the mould to resist deformation. This effect is well established in the case of cast irons, where high mould hardness is a condition for soundness. However, there is evidence that such a problem exists in castings of copper-based alloys and steels. A standard system such as statistical process control or other technique should be seen to be in place to monitor and facilitate control of such changes. Permanent moulds such as metal or graphite dies are relatively free from such problems. Similarly many other aggregate moulding materials are available that possess much lower thermal expansion rates, and so produce castings of greater accuracy and reproducibility. Many of these are little, if any, more expensive than silica sand. A move away from silica sand is already under way in the industry, and is strongly recommended.

The solidification pattern of castings produced from permanent moulds such as gravity dies and low-pressure dies may be considerably affected by the thickness and type of the die coat which is applied.

For some permanent moulds, pressure die-casting and some types of squeeze casting the feeding pattern is particularly sensitive to mould cooling. After the development and acceptance of the casting, any further changes to cooling channels in the die, or to the cooling spray during die opening, will have to be checked to ensure that corresponding deleterious changes have not been imposed on the casting. The quality of the water used for cooling also requires to be seen to be under good control if deposits inside the system are not to be allowed to build up and so cause changes in the effectiveness of the cooling system with time.

10.6.9 THE NON-FEEDING ROLES OF FEEDERS

Feeders are sometimes important in other ways than merely providing a reservoir to feed the solidification shrinkage during freezing.

We have already touched on the effect that feeders can have on the metallurgical quality of cast metal by assisting to restrain the unfurling and opening of bifilms by maintaining a pressure on the melt. This action of the feeder to pressurise the casting therefore helps to maintain mechanical properties, particularly ductility and fatigue strength.

A further key role of many feeders, however, is merely as a flow-off or kind of dump. Many filling system designs are so poor that the first metal entering the mould arrives in a highly damaged condition. The presence of a generous feeder allows some of this metal to be floated out of the casting. This role is expected to be hindered, however, in highly cored castings where the bifilms will tend to attach to cores in their journey through the mould.

In general, experience with the elimination of feeders from Al alloy castings has resulted in the casting 'tearing itself apart'. This is a clear sign of the poor quality of metal probably resulting mainly from the action of the poor running system. The inference is that the casting is full of serious bifilm cracks. These remain closed, and so invisible, whilst the feeder acts to pressurise the metal. If the pressurisation from the feeder is removed, the bifilms will be allowed to open, becoming visible as cracks. This phenomenon has been seen repeatedly in X-ray video radiography of freezing castings. It is observed that good filling systems do not lead to the casting tearing itself apart, even though the absence of a feeder has created severe shrinkage conditions. In this situation the casting shrinks a little more (under the action of solid feeding) to accommodate the volume difference.

Summarising and thinking further we have.

1. As we have seen, pressurisation raises mechanical properties, particularly ductility.
2. Pressurisation together with some feeding helps to maintain the dimensions of the casting. Although the changes in dimensions by solid feeding are usually small, and can often be neglected, on occasions the changes may be outside the dimensional tolerance. A feeder to ensure the provision of liquid metal under some modest pressure is then required.
3. Pressurisation can delay or completely prevent blow defects from cores.

In summary, providing the filling system design is good so as to avoid creating large bifilms, and provided the solidification rate is sufficiently fast to retain the inherited population of bifilms compact, castings that do not require feeders for feeding should not be provided with feeders.

10.7 RULE 7: AVOID CONVECTION DAMAGE

10.7.1 THE ACADEMIC BACKGROUND

Convection is the flow phenomenon that arises as a result of density differences in a fluid.

In a solidifying casting, the density differences in the residual liquid can be the result of differences in solute content as a consequence of segregation. This is a significant driving force for the development of channel defects known as the ‘A’ and ‘V’ segregates in steel ingots and as freckle trails in nickel- and cobalt-based investment castings. The name ‘freckles’ comes from the appearance of the etched components that shows channels containing randomly oriented grains that have been partly remelted in the convecting flow and detached from their original dendrites. Although, for many reasons, channel defects are unwelcome, they are usually not life threatening to the product. These defects are discussed earlier in Section 5.3.4 and are not discussed further here.

Convection can also arise as a result of density differences that result from temperature differences in the melt. There have been numerous theoretical studies of this phenomenon taking examples such as the solidification of low melting point materials in simple cubical moulds, of which one side is cooled and the other not. The resulting gentle drift of liquid around the cavity, down the cool face and up the non-cooled face, changes the form of the solidifying front. A schematic example is shown in Figure 10.35. These are interesting exercises, but give relatively little assistance to the understanding of the problems of convective flow in engineering systems.

The results from Mampaey and Xu (1999), who studied the natural convection in an upright cylinder of solidifying cast iron, showed that the thermal centre of the liquid mass was shifted upwards, and graphite nodules in spheroidal graphite irons were transported by the flow. Such studies reflect the gentle action of convection in small, simple shaped, closed systems; the kind of action one would expect to see in a cooling cup of tea. These facts have lulled us into a state of false security, assuming convection to be essentially harmless and irrelevant. We need to think again.

10.7.2 THE ENGINEERING IMPERATIVES

Convection was practically unknown as an important factor in shaped castings until the early 1980s. Even now, it is not widely known nor understood. However, it can be life and death to a casting, and has been the death of several attempts to develop counter-gravity casting systems around the world. Most workers in this endeavour still do not know why they failed. The Cosworth Casting Process nearly foundered on this problem in its early days, only solving the problem by its famous (infamous?) rollover system.

Thus convection is not merely a textbook curiosity. The casting engineer is required to understand and come to terms with convection as a matter of urgency. The problem can be of awesome importance and lead to major difficulties, if not impossibilities, to achieve a sound and saleable casting.

Convection enhances the difficulty of uphill feeding in medium section castings, making them extremely resistant to solution. In fact increasing the amount of (uphill) feeding by increasing the diameter of the feeder neck, for instance,

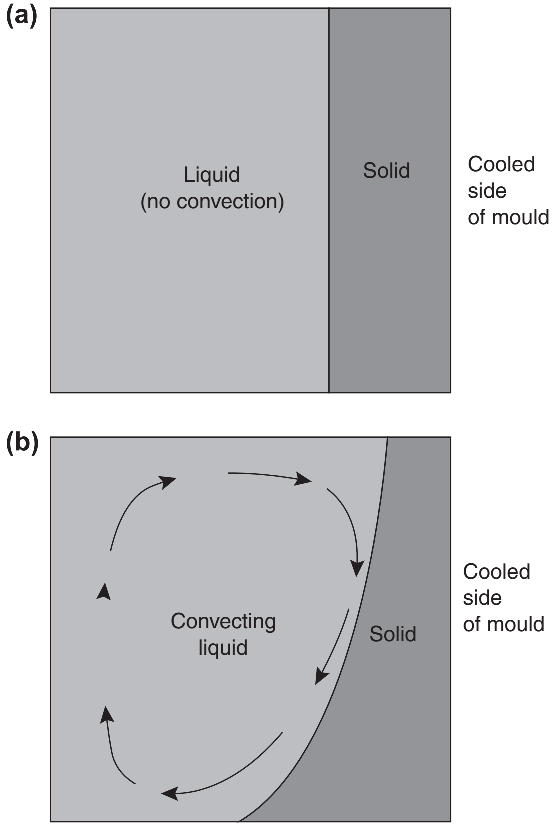

FIGURE 10.35

Solidification in a two-dimensional box of which only the right-hand side is cooled (a) planar front growth in the case of no convection; (b) the distortion caused by convective flow.

makes the feeding problem worse by increasing the opportunity for convection. Many of the current problems of low-pressure casting systems derive from this source.

In contrast, having feeders at the top of the casting, and feeding downwards under gravity is completely stable and predictable, and gives reliable results.

The instability of convective problems is worth emphasising. Because the heavy, cool liquid overlays the hotter less dense liquid, the situation is metastable. If the stratified layers of liquid are not disturbed there is a chance that the heavy liquid will remain wobbling around on the top and may solidify in place without incident. However, a small disturbance may upset the delicate balancing act. Once started, the cold melt will slip sideways, plunging downwards to the bottom, and the hot liquid surge upwards, so that a convective circulation will quickly establish. In practice therefore, several castings may be made successfully if the metastable equilibrium is not disturbed, but, inexplicably, the next may exhibit massive remelted and unfed regions.

Triggers to initiate the unstable flow could arise from many different kinds of uncontrolled events. A significant trigger could be an event such as the rising of a bubble from a core blow, as a result of an occasionally ill-fitting of a core print, leading to the chance sealing of the core vent by liquid metal.

Momchilov (1993) gives one of the very few accounts of the exasperating randomness of convection problems. He found that with the use of two riser tubes from a furnace containing liquid metal into one die cavity, successive castings could be observed to have completely different internal temperature histories. The first casting might be fine. However, the subsequent casting would suffer a die temperature inexplicably overheating by 120°C and the temperature in the furnace simultaneously dropping by 65°C. These are powerful and important exchanges of heat between the die and the crucible below. These changes caused the second casting to be partially remelted.

The use of twin riser tubes by Momchilov raises an important feature of convection. Convective flows are required to be continuous, as in a circulation. Thus in the case of two riser tubes into one die cavity, the conditions for a circular flow, up one tube and down the other, are ideal. It is likely that Momchilov would have solved his problem, or at least greatly reduced it, simply by blocking off one of the tubes.

The elimination of ingates in this way to solve convection problems in counter-gravity *fed* castings should be considered as a standard *first* step. This was found to be a useful measure in the early days of the Cosworth Process when it operated merely as a static low-pressure casting process. (The later development of the rollover concept represented a welcome total solution.)

The only other description of the problems of convection ever discovered by the author comes from a patent by Rogers and Heathcock (1990). They fall foul of convection during the attempt to make an aluminium alloy cylinder block casting in a counter-gravity filled permanent mould. They found that as the mould heated up the problem became worse, and the rate of flow of the convection currents increased. The microstructure of the casting was unacceptable in the area affected by convection. They dealt with the problem by providing strong cooling just above the ingates. This solution clearly threatened the provision of feed metal whilst the casting was solidifying, and so was a risky strategy. There is no record that the patent was ever implemented in production. Perhaps convection secured another victim.

Castings that employ a third mould part to site the running system under the casting are at risk of convective effects causing the melt to circulate up some ingates and down others via the casting above and the runner underneath. This is especially dangerous if the runner is a heavy section. Pressurising the runner with an adequate feeder is a way of maintaining the net upward movement of metal required for the feeding of the casting, thus reducing the deleterious effects of the convection to merely that of delaying freezing. In this case, the worst that happens is the development of a locally coarser structure.

Investment casting often provides numerous convective loops in wax assemblies as a result of attaching the wax patterns at more than one point to increase the strength of the complete wax assembly. A typical wax assembly for the casting of polycrystalline Ni-base turbine blades is illustrated in Figure 10.36(a). The central upright is surrounded by six blades (only two are shown in the section), so that in addition to its heavy section designed to act as a feeder, it is kept even hotter by the presence of the surrounding blade castings that prevents loss of heat by radiation. Conversely, of course, the blades cool quickly because they can radiate heat freely to the cool surroundings. A convective loop is therefore set up with hot metal rising up the central feeder and falling through the cooling castings. The final grain structure seen on the etched component reveals the path of the flow (Figure 10.36(b)). The casting is designed to have fine surface grains nucleated by the cobalt aluminate addition to the primary coat of the mould. However, because hot metal enters the mould cavity from the top, sweeping down through the casting after the chill grains are formed, the original chill grains are remelted in those patches where the flow brushes against the walls. The flow becomes a concentrated channel as it exits the base of the blade. The very narrow section of the trailing edge of the casting is not penetrated, and so escapes remelting, as does the large region in the bottom right that the flow has missed.

Very large blades for the massive land-based turbines for power generation are sometimes cast horizontally. In this case, each end of the casting is subject to convective problems as is seen in Figure 10.37.

The cutting of convective links in wax assemblies is recommended and cries out for wide attention in most current investment casting operations. The strengthening of wax assemblies by wax links inadvertently provides convective links and should be avoided. Ceramic rods can provide strengthening, or, if wax connections are used, they should be plugged with a ceramic disc to avoid metal flow. These simple modifications to the wax assembly will completely change the mode of solidification of the castings, allowing for the first time an accurate understanding of filling and feeding effects.

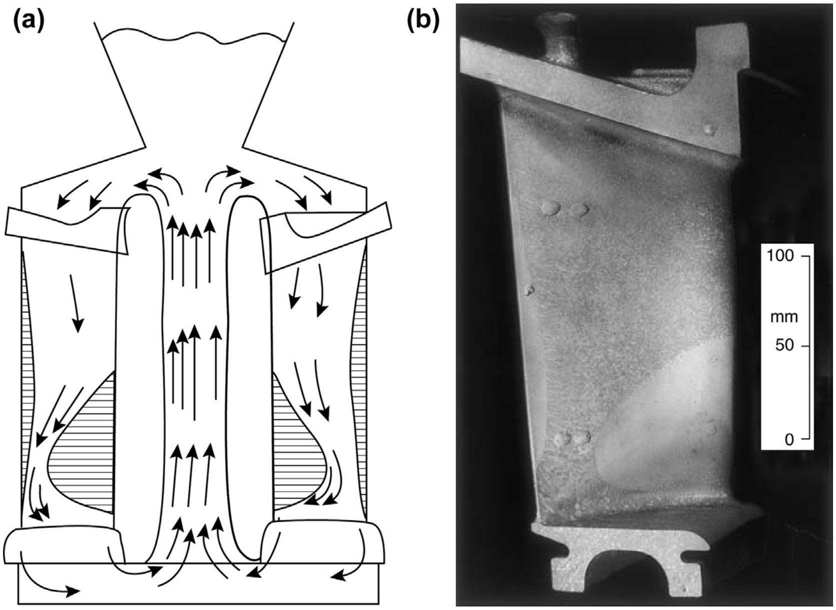

FIGURE 10.36

(a) Lost wax assembly of six Ni-base turbine blades around a central feeder, showing the expected convective loops; (b) an etched blade showing the remelting of the fine surface grains created by the cobalt-aluminate nucleant in the mould surface, and the subsequent growth of coarse grains that define the patches where the flow paths impinged on the casting surface.

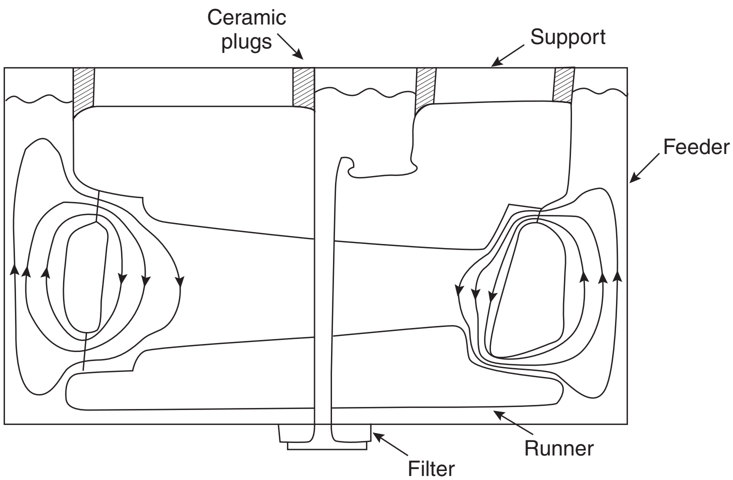

FIGURE 10.37

Horizontal orientation of a large investment-cast turbine blade, illustrating convective loops in the root and shroud. The flows convey heat from the cylindrical feeders, remelting regions of the casting.

Other problems in sand castings are illustrated in Figure 10.38. Gravity die (permanent mould) castings are less prone to these problems because of their more rapid rate of heat extraction by the metal mould. For castings in metal moulds, the sections have to be considerably larger before convection starts to be a threat. Figure 10.39 shows the convection effects from side or bottom feeding compared with the relative stability of top feeding.

It is evident that many computer predictions of heat flow and the feeding of castings will be quite inadequate to deal with convection problems because it is usual to consider the loss of heat from castings simply by conduction. Clearly, thicker sections in a loop will cool more quickly than the computer would predict because convection allows them to export their heat. Conversely, of course, thin sections in the same loop will suffer the arrival of additional heat that will greatly delay their solidification. In fact, if the hot section has an independent source of heating, such as the electrical heating provided in many counter-gravity systems, the sections in the loop can circulate for ever. The computer would have particular difficulty with this.

Even so, the greater speed and sophistication of computing will eventually provide the predictions containing the contribution of convection that are so badly needed. It is hoped that future writers and founders will not need to lament our poor abilities in this area.

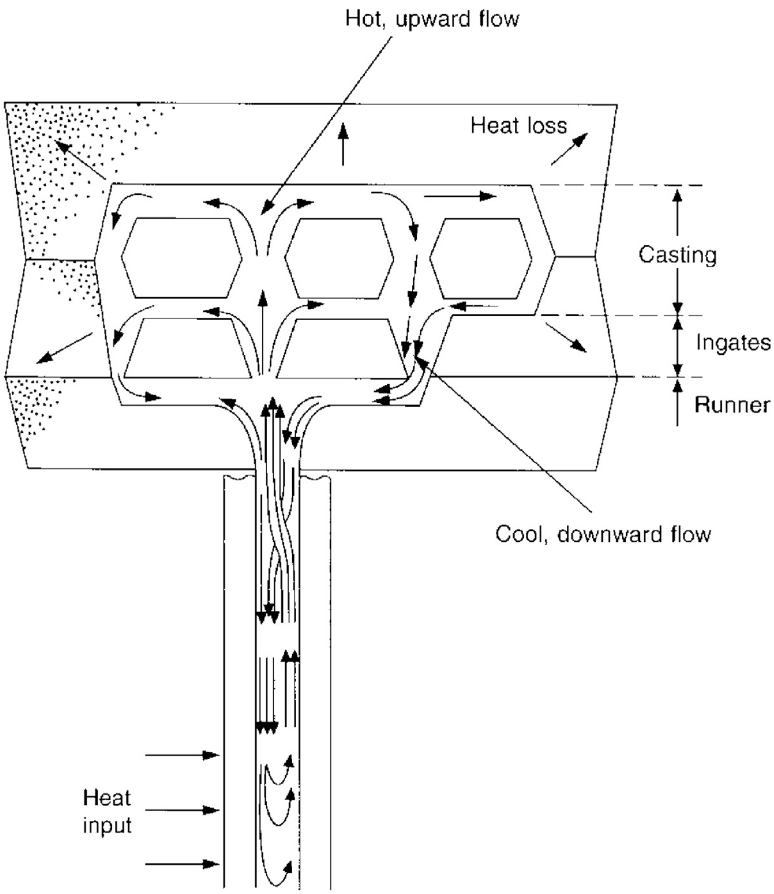

FIGURE 10.38

Convection driven flow within a solidifying low-pressure casting.

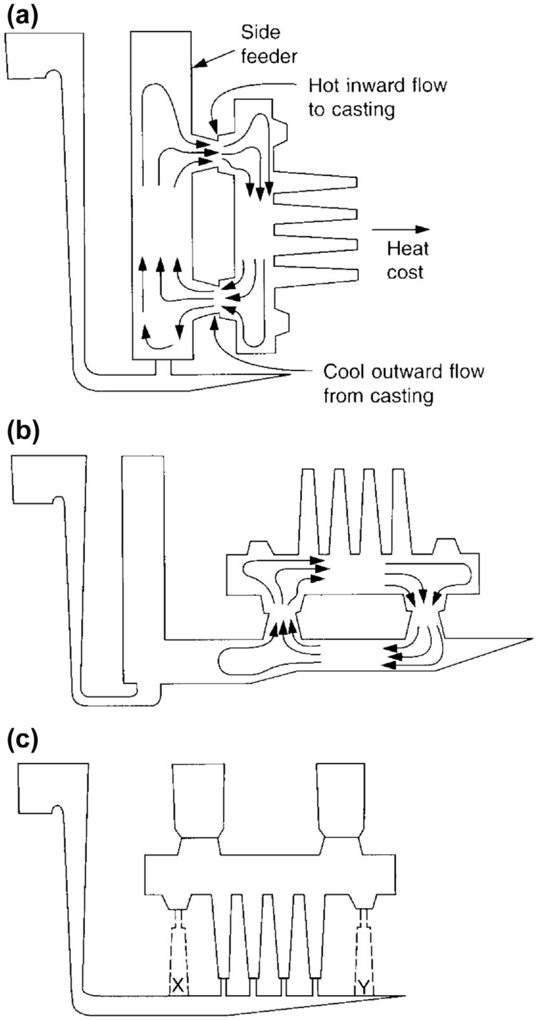

FIGURE 10.39

Encouragement of thermal convection by (a) side feeding; (b) bottom feeding; (c) its near elimination by top feeding.

10.7.3 CONVECTION DAMAGE AND CASTING SECTION THICKNESS

If the solidification time of the casting is similar to the time taken for convection to become established, extensive remelting can be caused by convective flows. Serious damage to the micro- and macro-structure of the casting can then occur. The convective flow takes about one or two minutes to gather pace and organise itself into rapidly flowing plumes. This is occurring at the same time as the casting is attempting to solidify. The flows cut channels through the newly solidified material, remelting volumes of the casting.

Castings that freeze in a time either shorter than 1 min or longer than perhaps 10 min are expected to be largely free from convection problems as indicated below.

Thin section castings are largely free from convection difficulties. They can therefore be fed uphill simply because the thin section gives (1) the viscous restraint of its nearby walls makes any convective tendency more difficult and (2) more rapid freezing allows convection less time to develop and so less time to wreak damage in the casting. Thus instability is (1) suppressed and (2) given insufficient time, respectively, so that satisfactory castings can be made.

Conversely, thick section castings taking perhaps 10 min to 10 h to freeze are also relatively free from convection problems, because the long time available before freezing allows the metal plenty of time to convect, re-organising itself so that the hot metal floats gently into the feeders at the top of the casting, and the cold metal slips to the bottom. All this activity occurs and is complete before any significant amount of solidification has occurred. Thus the system reaches a stable condition before damage can be caused. Once again, castings are predictable.

In what can only be described as a perverse act of fate, convection does its worst in the most common sizes of castings, the problem emerging in a serious way in the wide range of intermediate section castings. These include the important structural castings such as automotive cylinder heads, cylinder blocks and wheels, and the larger investment cast turbine blades in nickel-based alloys amongst many others. Convection can explain many of the current problems with difficult and apparently intractable feeding problems with such common products.

Channels cut through the structure of the casting which is attempting to solidify will contain a coarse microstructure because of their greatly delayed solidification and in addition may contain shrinkage porosity if unconnected to feed metal. This situation is likely if the feeders solidify before the channels as undoubtedly happens on occasions, because the channels derive their energy for flow from some other heat source, such as a very heavy section low down on the casting, or the ingate attached to the riser tube of a counter-gravity furnace for instance.

For conventional gravity castings that require a lot of feed metal, such as cylinder heads and blocks, and that are bottom-gated, but top-fed, this will dictate large top feeders, because of their inefficiency as a result of being furthest from the ingates, and so containing cold metal. This is in contrast to the ingate sections at the base of the casting that will be nicely pre-heated. The unfavourable temperature regime is of course unstable because of the inverted density gradient in the liquid, and thus leads to convective flow, and consequent poor predictability of the final temperature distribution and effectiveness of feeding. It is the standard legacy of bottom filling: the favourable filling conditions leading to the worst feeding conditions. Life was never easy for the casting engineer.

The upwardly convecting liquid within the flow channels usually has a freezing time close to that of the pre-heated section beneath, which is providing the heat to drive the flow. In the case of many low-pressure systems, the metal supply system is artificially heated, leading to a constant heat input, so that the convecting streams rising out of these regions never solidify. This is what happened to the Cosworth system in the early days of its development. When the mould and casting (which should by now have been fully solidified) were hoisted from the casting unit, liquid poured from the base of the mould, emerging from remelted channels to the amazement of onlookers. Everyone had assumed that after the appropriate length of time for solidification, no liquid could possibly still be present, and present in such quantity.

When removing a convecting casting from a counter-gravity filling system in this way, the draining of liquid from the interdendritic regions leaves regions in the casting that appear convincingly like shrinkage defects, and are usually confused as such.

The convection of hot metal up and the simultaneous movement of cold metal down the riser tube of a low-pressure casting unit delays the freezing of the casting in the mould above and can lead to a significant reduction in productivity. A thermocouple in the riser tube reveals the chaotically varying temperature of the hot and cold eddies as they swirl past each other.

The author is aware of a series of castings being made on a low pressure machine whose freezing time kept increasing as the melt was subjected to increasingly thorough rotary degassing treatment. It seems that each rotary degassing treatment reduced the amount of bifilms in suspension. Because the effective viscosity of the melt was progressively reduced in this way the convection increased, extending the time taken for the casting to solidify. Thus clean metal is free to convect, whereas melt with an internal semi-rigid lattice of bifilms will be more resistant to flow. It is perhaps superfluous to add that reliance on the inhibition of convection by filling the liquid with oxides to increase viscosity cannot be recommended.

10.7.4 COUNTERING CONVECTION

Solutions to the problems of convection are summarised as follows.

1. Rollover.

The inversion of the mould after casting effectively converts the pre-heated bottom ingate filling system into a top feeding system, thus gaining a really efficient feeding system.

Furthermore, of course, the massive technical benefit of the inversion of the system to take the hot metal to the top, and the cold at the bottom, confers stability on the thermal regime. Convection is eliminated. For the first time, castings can be made reliably without shrinkage porosity.

Batty (1935) was well ahead of his time, using the roll over technique for steel castings. He showed by careful measurement of the temperature gradient the reasons for his success to convert 'troublesome' steel castings to 'reliable' steel castings.

The massive productivity and economic benefit of this technique when applied to Al aggregate moulded castings follows because the mould now contains its liquid metal all below the entry point in the mould, allowing the mould to be removed from the casting station without waiting for the metal to freeze (which is of course the standard productivity delay suffered by most counter-gravity casting processes). In this way, cycle times can be reduced from about 5 to less than 1 min. This is a powerful and reliable system used by such operations as Cosworth Process which achieved the casting a mould containing two cylinder blocks every 45 s. The technique has made this sand casting process the fastest automotive block production process in the world. In addition, the castings are of a superb integrity.

2. Tilt casting

Those tilt casting processes in which the rollover is used *during* casting actually to effect the filling process can also satisfy the top feeding requirement.

However, in practice care is needed because many geometries are accompanied by waterfall problems, if only by the action of the sliding of the metal in the form of a stable, narrow stream down the sloping side of the mould. Thus meniscus control is, unfortunately, often poor. Where the control of the meniscus can be improved to eliminate entrainment problems, tilt casting techniques are valuable. Ultimately, if the tilt is controlled to perfection, a kind of horizontal transfer of the melt can be achieved. This system does not seem difficult or costly to attain, and is commended strongly.

3. Cut convective loops.

Explore the elimination of ingates when counter-gravity feeding of castings. The widespread use of convective loops in investment castings wax assemblies should be reduced by the wider use of ceramic supports and stops.

10.8 RULE 8: REDUCE SEGREGATION DAMAGE

Section 5.3 describes the many mechanisms whereby concentration of solutes can be enhanced or depleted in some regions of the casting during the short time of solidification.

Microsegregation between dendrite arms is, as we have seen, a necessary consequence of normal freezing of an alloy, but the fine scale of the effect allows a significant amount of redistribution during subsequent homogenising and solutionising heat treatments.

The various forms of macrosegregation, on the other hand, are irreversible and have to be lived with. They can sometimes be sufficiently serious to threaten the serviceability of the casting, and in any case will threaten to scrap the casting if the composition is found to be outside the specification limits in places, whether or not this is important for the function of the casting. (At times, the founder has to live with the injustice of castings rejected for illogical or unreasonable objections, even though destructive testing to simulate service conditions would in principle provide evidence of the complete harmlessness of most deviations from over-tight specifications.)

One of the most common of macrosegregation problems is that arising from dendritic freezing. This perfectly normal mode of segregation occurs in practically all castings to some degree: the positively partitioning solute is concentrated against a cooling surface. In sand castings, the effect is rarely a problem because of the relatively small temperature gradients, but can be serious in metal moulds and when using metallic chills. Thus the use of chills requires caution in those alloys prone to segregation. In addition, the use of a fin or a pin can simulate the action of a chill so as to give a local segregation effect.

For Al alloys, segregation is not usually a problem, especially in alloys such as those based on the Al-Si-Mg system. However, for those Al alloys that contain Cu, particularly if the Cu content is in the region of 4.5–5.5%Cu, the variations in Cu content throughout a casting can be serious.

Figure 10.40(a) gives an example. Dendritic segregation will lead to a concentration of Cu at a chill face, but an opposite pattern adjacent to the feeder. Both regions can easily be well outside specification. A reduced section thickness can act as a cooling fin and so simulate the action of a chill, causing a peak in Cu segregation adjacent to the fin as shown in Figure 10.40(b). However, just inside the fin, the conditions are those of a thin-section casting joined on to a heavy section, thus delaying freezing, so that the opposite pattern of segregation is to be expected. This steep change in composition will, of course, result in a correspondingly large variation in mechanical properties, particularly after heat treatment. It is a concern that these large changes in strength, possibly from brittle to ductile behaviour, are concentrated at the change in section at which externally applied stress will also be concentrated. These effects do not appear to have been investigated so far, but clearly should be explored. Design rules for avoiding or reducing these problems would be welcome.

The other macrosegregation effect that is troublesome in slowly cooled castings is the channel defect. This arises from the effect of gravity on the solute in the dendrite mesh, causing liquid high in solute to migrate, organising its flow into channels that remelt a pathway through the mesh. The effect has been countered in some alloy developments by the design of more neutral buoyancy into the segregated melt. However, this luxury is not normally open to the majority of castings that have to conform to a chemical specification, even though the specification is often rather arbitrary and rarely of any consequence for the satisfactory operation of the product in service.

Channel segregates appear also to concentrate bifilms, as evidenced by the numbers of cracks that can occur in these lengthy pencil-like channels. The bifilms will be pushed into these regions. Later, when most of the casting is solid, the highly segregated liquid in the channel will remain liquid as a result of its low melting point. The local high gas content in solution, and the local reduction in pressure resulting from the problems of feeding down such long channels, partly blocked by tumbling melted-off grains (the ‘freckles’ when viewed by sectioning and etching), will combine to unfurl and expand the bifilms, creating planar cracks across the width of the channel. The cracks are, of course, much more serious than the segregates themselves and are probably the sole reason why channels are considered to be deleterious. If the mould could be filled with good-quality metal, channel defects would probably be harmless.

The vigorous operation of channel segregates in heavy steel castings funnels low-density solutes, mainly high in carbon, into the feeder head of the casting. Unfortunately, this segregated liquid accumulating in the feeder is not harmlessly put away, but awaits its turn to degrade the casting at a later stage, when the feeder comes into operation to

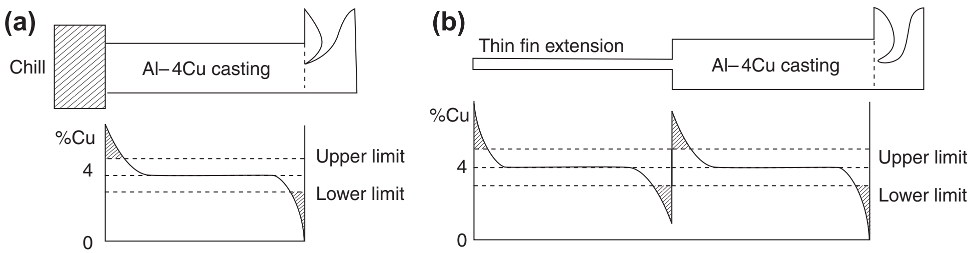

FIGURE 10.40

(a) Dendritic segregation pattern, concentrating solute against a chilled face; (b) analogous pattern produced by a reduction in section thickness acting as a cooling fin.

feed the solidification shrinkage in the body of the casting. At this late stage, the high carbon liquid is drawn down into the casting, giving a carbon-rich region immediately below the feeder (Figure 10.41). If this action is especially severe, as occurs in heavy steel ingots, channels are melted from the feeder into the casting by this sucked-down low melting point liquid. We call these channels 'V segregates'.

The authors Li et al. (2014) describe an excellent solution to the huge segregation problems of very large ingots, especially those intended for power plant forgings. Instead of filling up a single ingot mould with a single pour of metal, the mould is filled in stages, using small pours from a succession of small ladles. Each pour is allowed to solidify to some degree, and when approximately 20% residual liquid remains, the next ladle is poured, and so on until the mould is completely full. Thus a 300 tonne ingot might require 10 pours of 30 tons, each waiting until sufficient solidification had taken place, but with sufficient remaining liquid to ensure a problem-free joint. The final ingot is impressively uniform.

In shaped castings, the carbon segregation under the feeder is controlled to some degree by the diameter of the feeder. Clearly, in the extreme, a feeder of the same diameter of the casting would cause minimal segregation. However, this is clearly not a particularly useful solution but may indicate that a modest increase in feeder diameter might be helpful, and a narrow feeder neck breaker core may do more harm than good. Faster freezing of the casting would help most. The founder needs to be aware of this solute pattern and may ultimately have to target the fine balance between too little carbon in the main body of the casting and too much beneath the feeder. Once again, the foundry engineer would greatly benefit from some quantitative guidance from a program of carefully conducted research.

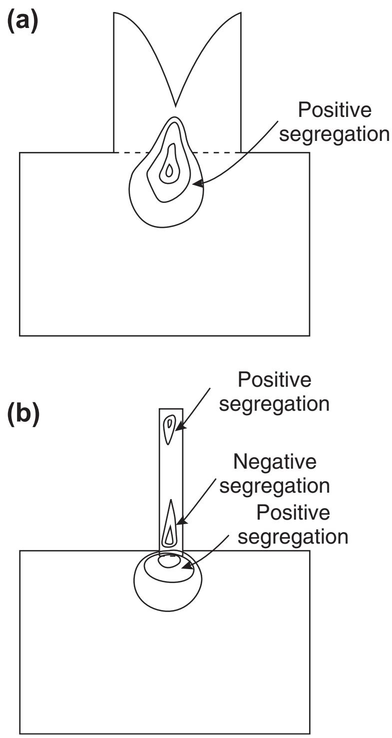

FIGURE 10.41

(a) Positive segregation under a feeder; (b) positive segregation under a cooling fin, but negative closely adjacent inside the fin. These extremes of concentration of solute are both close to a vulnerable change in section, where stress may be concentrated.

The under-feeder segregation is compared and contrasted with the opposite thermal condition: an under-fin condition giving some more complexity overall, but similar results, if on a smaller scale, in the casting (Figure 10.41(b)). Naturally, an under-chill condition is similar. The fin and pin actions are not likely to be noticeable in steel castings as a result of the poor thermal conductivity, but the situation is likely to be non-trivial in Al, Mg and many Cu alloys.

A final observation on segregation is important. All segregations take time to build up. Thus reducing the time available by almost any method is a valuable and powerful general strategy to reduce macrosegregation. The provision of additional chills arranged to reduce the temperature gradients that would have been set up by a single chill, especially using internal chills if possible, is strongly recommended. If the customer would agree, a move to a less segregating alloy would also be an effective strategy.

10.9 RULE 9: REDUCE RESIDUAL STRESS

10.9.1 INTRODUCTION

It seems that, in general, the engineering community has not been made aware of the potentially awesome importance of residual stress in the manufacturing of engineering components. All manufactured components contain internal stress, often high. The problem is that this very real danger is invisible.

Most foundries and machining operations have stories about the casting that flew into pieces with a bang when being machined or even when simply standing on the floor! Benson (1946) describes one such event. The author and several colleagues narrowly missed the shrapnel from an aluminium alloy compressor housing that exploded in their midst while being cut on a band saw. It happened late one evening after the foundry had stopped work for the day. The building was quiet and in darkness. After the initial bang and the return to silence, the sound of fragments bouncing off the roof and clattering echoes of them distantly falling to the floor was awesome.

It is easy to disregard such stories. However, they should be viewed as warnings. They warn that, in certain conditions, castings can have such high stresses locked inside that they are dangerous and unfit for service. In fact, they can act like small bombs. We are unaware that the casting may be on the brink of catastrophic failure because, of course, the problem is invisible; the casting looks perfect.

There are those metallurgists within the industry, some eminent, and whose opinions on other matters I respect, that have taken issue with me. They have argued that the presence of residual stresses, particularly those from quenching, are actually irrelevant because the whole component is in balance with its own stresses. The question of overall balance is certainly true. However, this argument overlooks that the *distribution* of stress is to be expected to be far from uniform, and parts of the component may be near to their failure stress even before the application of any stress in service. Usually, as we shall see, the major tensile stress is in the centre, and it is this part of the component that fails first under tensile load.

Admittedly, not all components are necessarily endangered by internal stress. Indeed, in some cases the stress can be beneficial (some examples are given later). However, in general, the very real risk exists that the stress may not be beneficial. The residual stress may add to the service stress and so promote premature failure at only low service stress, to the bewilderment of the designer who imagines the material of his component to be uniform and who has completely overlooked the possibility that the part contains invisible threats from stresses. He responds by increasing the thickness of the section—a response which is likely to be counter-productive.

Because of the complexity of some castings, and the complexity of the state of stress, it is usually not easy to estimate the magnitude of either the internal residual stress or its precise action. Often however, the stress is at least equal to or exceeds the yield stress. Thus it is not trivial. In fact at this level it will dominate all other designed loads. If the casting is in a fatigue condition, it will certainly lead to early failure. It is ignored at our peril.

This section takes a look at the wide spectrum of stresses in castings, and attempts to clarify those that are important and which should be controlled, in contrast to those that can be safely neglected.

10.9.2 RESIDUAL STRESS FROM CASTING

There have been several test pieces that have been used over the years to help assess the parameters affecting the residual stress in castings. Most of these are based on the form of the three-bar frame casting shown in Figure 10.42.

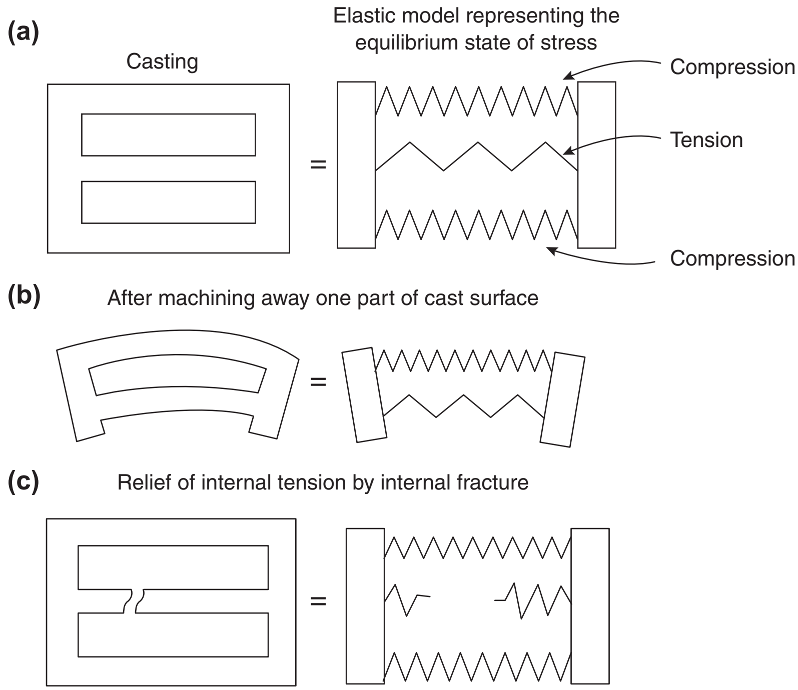

FIGURE 10.42

Heyn's (1914) model of the balance of internal stresses after rapid cooling: (a) quenched casting showing high internal tensile stress and relatively low external compressive stress; (b) the distortion of the casting after one side is machined away; and (c) the common condition of internal tensile failure.

In practice the stress remaining in the casting is usually assessed by scribing two lines on the central bar and accurately measuring their spacing. The bar, of length L, is then cut between the lines, and, usually, the cut ends spring apart as the cut is completed. The distance between the lines is then measured again and the difference ΔL is found. The strain ε is therefore $\Delta L/L$ and the stress σ is simply found, assuming the elastic (Young's) modulus E, by the definition:

$$E = \sigma/\varepsilon \tag{10.19}$$

Such studies have revealed that the residual stress in castings is a function of the cooling rate in the mould, as shown for aluminium alloy castings from the effect of water content of the mould in Figure 10.43. Dodd (1950) cleverly illustrated that this effect is not the result of the change of mould strength by preparing greensand moulds with various water contents, then drying each carefully so that they all had the same water content. This gave a series of moulds with greatly differing strengths. When these were cast and tested, there was found to be no difference in the residual stress in the castings. This result was further confirmed by testing castings made in moulds rammed to various levels of hardness. Again, no significant difference in residual stress was found. Dodd also checked the effect of casting temperature and noted a small increase in residual stress as casting temperature was increased.

As with cases of the constraint of the casting by the mould, removing the casting from the mould at an early stage would be expected to be normally beneficial in reducing residual stress. Figure 10.44 shows the result for iron and a high-strength aluminium alloy. The higher residual stress for cast iron reflects its greater rigidity and strength. The effects of percentage water in the sand binder, and of stripping time and casting temperature, have been confirmed in other work on high strength aluminium alloys and grey iron using a rather different three-bar frame (IBF Technical Subcommittee,

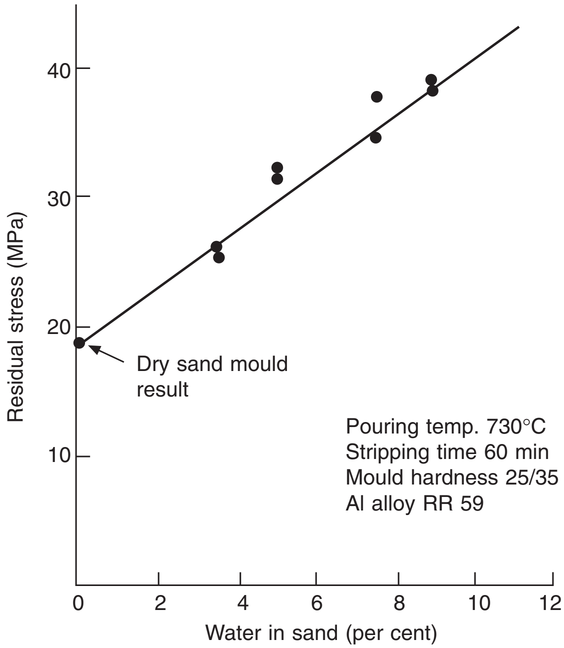

FIGURE 10.43

Internal stress in the centre member of a three-bar frame as a function of water content of the greensand mould (Dodd, 1950).

1949, 1952). It is noteworthy that all these direct measurements show casting stresses to be modest; only approximately 10% of the yield strength.

All these observations appear to be explainable assuming that the main cause of the development of residual stress is the interaction of different members of the casting cooling at different rates. Beeley (1972) presents a neat solution to the problem.

The strain $\Delta L/L$ resulting from differential contraction is determined by the temperature difference ΔT and the coefficient of thermal expansion of the alloy α. Thus, we have the strain

$$\varepsilon = \Delta L/L = \alpha \Delta T \tag{10.20}$$

and so from Eqn (10.19), the stress is

$$\sigma = \alpha E \Delta T \tag{10.21}$$

The stress therefore depends on the temperature difference between members. It is also worth noting that the stress is independent of casting length L.

A rather different result illustrating the influence of the geometry of the three-bar frame casting was found by Steiger (1913). He measured the increase of stress in the centre bar of grey iron castings after increasing the rigidity of the end cross-members. He found that a centre bar of more than twice the diameter of the outer bars would suffer a residual stress of more than 200 MPa, sufficient to fracture the bar during cooling. Working Eqn (10.21) backwards, it is quickly shown that the temperature difference was only about 100°C to produce this failure stress. Clearly, such temperature differences will be common in castings, and often exceeded. Thus, high stresses may be expected for some castings. Even so, this conclusion was reached from very early work (1913), and should really be confirmed with modern techniques and equipment before we place too much faith in this result.

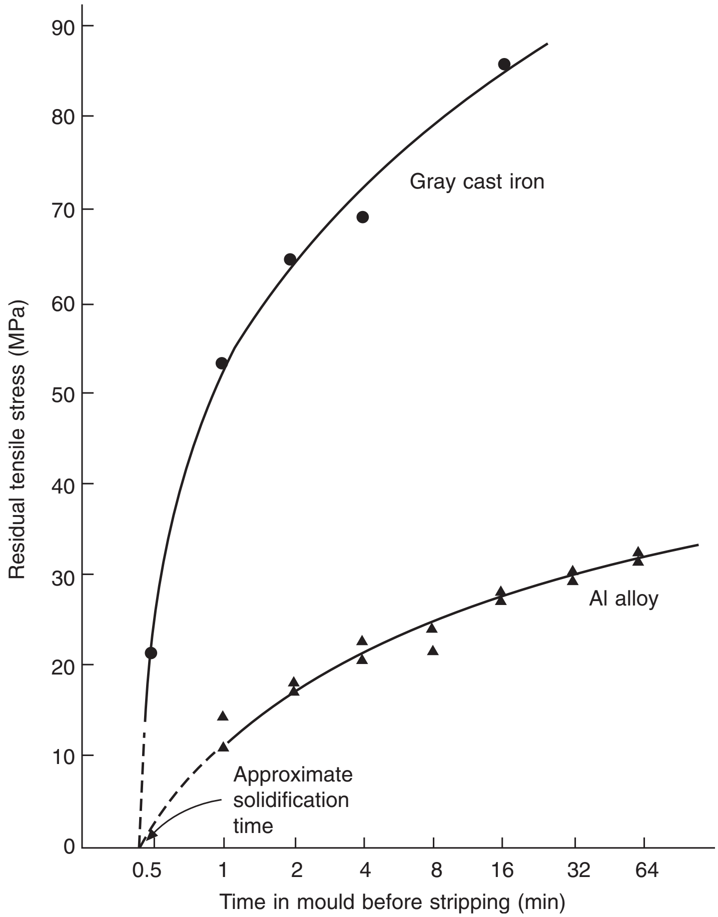

FIGURE 10.44

Residual stress in Al alloy and grey iron castings as a function of stripping time.

Data from Dodd (1950) and IBF Technical Committee (1949).

In ferrous castings that experience a gamma to alpha (austenite to ferrite) phase change during cooling, any stresses that are built up before this event are probably reduced, their memory diluted by the uniform plastic flow that the transformation causes throughout the metal. It seems probable therefore that the temperature differences and cooling rates applying below the gamma-alpha transformation temperature that are the most important for the final remaining levels of stress in these steels.

This fact prompts Kotsyubinskii (1961) to recommend that heavy sections of ferrous castings be cooled by forced air or chills to equalise their cooling rates with those of the thinner sections, up to the point at which the pearlite reaction occurs. Below this temperature, little can be done to avoid the buildup of stress. This is because the metal is largely elastic, and plastic relaxation, occurring only slowly by creep, becomes progressively less effective; thus cooling should at that stage be slow and even, so as to take advantage of as much natural stress relief as possible.

In an aggregate mould, castings are cooled relatively slowly, so that the final internal stress in the product will normally be relatively low and can often be neglected (although the possible exception warned by Steiger previously is noted). It is true that the dimensions of the casting will often be changed by stress during cooling, but on shaking out from the mould the final, residual, stress will not normally be high. In addition, the distortions that have arisen during cooling in the mould are usually extremely reproducible. This is a consequence of the reproducible conditions of production, in which the mould is the same temperature each time, and the metal is the same temperature each time, so that the final shape is closely similar each time. This reproducibility of sand castings is probably greater than for any other casting process.

This repeatable regime is not quite so well enjoyed by the various kinds of die casting, particularly gravity die (permanent mould) casting, as a result of many factors, but in particular the variability of mould size and shape as a result of variation of mould temperature. The somewhat faster cooling, particularly because of the earlier extraction of the casting from the mould, is an additional factor that does not favour low final stress.

In general, internal stress remaining from the casting process is rarely high enough to be troublesome, but we cannot always be complacent about this. The ability to predict stresses using computer simulation will be invaluable to maintain a cautious watch for such dangers.

Ultimately, however, particularly for aluminium alloys, the stresses from casting are usually eliminated by any subsequent high temperature solution heat treatment which acts as an excellent stress-relieving treatment.

In confirmation, Benson (1938) made the important commonsense generalisation that it is the last thermal treatment, and the rate of cooling from this final treatment, that is important so far as residual stresses are concerned. Thus the last treatment might simply be casting, or annealing, or quenching from a solution treatment. Having considered casting, we shall now turn our attention to these subsequent treatments.

10.9.3 RESIDUAL STRESS FROM QUENCHING

When quenching castings from a high temperature heat treatment, the time for cooling the outer sections of castings is shorter than the time required for heat to diffuse out from interior sections. The outer parts of the casting therefore cool first to form a rigid, strong frame that will contract to squash the hot, weak, plastic central features to shorter lengths. Later, the inner sections, now somewhat smaller in length, will cool, gathering strength and contracting further, but this late contraction will suffer the restraint of the outside sections that have by now have become cold and rigid. Thus, the interior sections, now also becoming strong and rigid, go into tension, forcing in turn the outer parts into compression. Thus the major problems of internal tensile stress and distortion of the casting are usually created in quench operations.

Furthermore, the stresses are not significantly reduced by a subsequent ageing treatment. The temperatures and times for ageing treatments are too low to lead to stress relief.

It is unfortunate that many heat treatments require a quenching stage, intended to cool the casting sufficiently quickly to freeze solutes in to a solid solution, thereby preventing them from precipitating too early, and thus still being available for controlled precipitation during the ageing treatment. If the quench is slow some solute may be lost by precipitation from solution, thus making it unavailable for subsequent hardening reactions, so that the final strength of the casting is reduced. This reasoning has driven metallurgists to seek quenching rates as fast as possible.

The problem has been that all such research by metallurgists to optimise heat treatments has been carried out on test bars of a few millimetres in diameter that represent no problem to cool quickly. The outside and inside of the bars are in excellent thermal communication, and the high thermal conductivity of most metals ensures that the cooling throughout the section is essentially uniform. Thus, the world's standards on heat treatment often dictate water quenching to obtain the highest material properties.

Quite clearly, the problem of larger components, or certain components of special geometrical complexity in which uniform cooling is an impossibility, has been overlooked. This is a most serious oversight. The performance of the whole component may therefore be undermined by the application of these techniques that have been optimised by research on small test bars, and which therefore are inappropriate, if not actually dangerous, for many large and complex components.

This is such a common problem that when a troubled casting user telephones me to say words to the effect 'My aluminium alloy casting has broken. What is wrong with it?' This is such a regular question that my standard, and rather tired, reply now is 'Do not bring the casting to me. I will tell you *now* over the telephone why it has failed. It has failed because it has been poured badly and therefore contains bifilms that reduce its strength. However, in addition, you have carried out a solution heat treatment accompanied by a water quench'. The caller is usually stunned, incredulous that I know that he has water quenched his casting, and asks how I know. My experience is this: in all my life investigating the causes of failure of perhaps hundreds of Al alloy castings, only one failed because of embrittlement caused as a result of the alloy being wildly outside chemical specification. All the rest failed for only two reasons: (1) weakening by bifilms, together with (2) massive internal stresses that have loaded the already weakened casting close to its failure stress even before any service stress was applied.

I have to record, with some sadness, that all the standard metallurgical investigations into casting failures that I see appear completely irrelevant; they include such costly and time consuming activities as the determination of the chemical specification, the metallurgical structure, the mechanical properties such as hardness and other standard metallurgical tests. It underlines the importance of understanding the new metallurgy of cast metals in which the residual stresses and bifilms together play the dominating roles in the performance of engineering components, particularly cast engineering components. It is necessary to sympathise with the metallurgist attempting to carry out the traditional failure investigation because both these dominating macroscopic defects are hard to detect: the stress is perfectly invisible, and the bifilms are mostly invisible, but these two invisible factors are in control.

Equation (10.21) explains why not all shapes and sizes of castings necessarily suffer a problem. Compact or small castings, and those for which the quenchant can easily reach all parts, are often not seriously affected, because ΔT is necessarily small. It is essential to remember this: not all castings are seriously affected by quench stresses.

However, for those castings that are affected by uneven temperatures during cooling, it is salutary to estimate the actual magnitude of the strain ε. For aluminium the coefficient of expansion α is about $20 \times 10^{-6}\ \mathrm{K}^{-1}$ and the temperature fall during a quench is approximately 500 K. The strain works out to be therefore approximately as $20 \times 10^{-6} \times 500 = 10^{-2} = 1\%$. For steels, α is approximately 14×10^{-6}, and the temperature change for quench from many heat treatments in the region of 900 K, again giving a strain close to 1%. Because the yield (or proof) strain is only approximately 0.1%, these imposed quench strains are about 10 *times* the yield strain. These dramatic strains can therefore be seen to take the component well into the plastic deformation range.

For steels with elongations to failure of perhaps 30–50%, the imposition of 1% strain is usually not a serious threat to the serviceability of the casting, even though the corresponding stress is very high.

However, for many Al alloys, with only 1–2% elongation to failure, the imposition of 1% strain can be highly threatening. The comforting thought for the future is that as elongations for Al alloys approach those of steels because of the steady improvement of cleanness, the danger of residual stress in such castings will decrease.

In the meantime, casting geometries that are particularly susceptible include large, thick section castings, where the heat of the interior takes time to reach the outside of the casting, giving high ΔT. Ingots or other block-type products can be seriously stressed for this reason. Direct chill continuously cast ingots of aluminium alloys are severely cooled by water, but are often larger than 300 mm diameter. Whilst sitting on the shop floor waiting further processing the strong 7000 series alloy ingots have sometimes been known to explode like bombs.

As an aside, the length of time taken before the ingot decides to fail is curious and interesting. It seems likely that the failure under the high internal stresses is initiated from one of the large bifilms that is expected to be entrained during the turbulent start of casting. The gradual precipitation of hydrogen into the bifilm will gradually increase the pressure in the bifilm crack, encouraging it to extend as a stress crack. The hydrogen may be already in solution in the metal or may be gradually accrued by reaction with water vapour in the atmosphere during storage, especially if part of an extensive bifilm is near to the ingot surface where it can receive hydrogen by diffusion from the outside surface, but is therefore able to distribute the pressurised gas over the whole area of the bifilm to create significant stress at the sharp edges of the defect. Other penetrating contaminants may include air to cause additional internal oxidation, or fluxes, or traces of chlorine gas, or sulphides from greases, to act as surface active additions to reduce the surface energy of the metal and so further encourage crack growth. Research to clarify these possibilities would be valuable.

Other varieties of castings that are susceptible to damaging levels of residual stress include those that are hollow, with limited access for the quenchant into the interior parts of the casting, and which also have interior geometrical features such as dividing walls and strengthening ribs (Figure 10.45). This latter series of geometrical requirements might seem to be an unlikely combination of features that would eliminate most castings. Perhaps surprisingly therefore, the list of castings that fulfils these requirements is rather long and includes such excellent examples as automotive cylinder heads

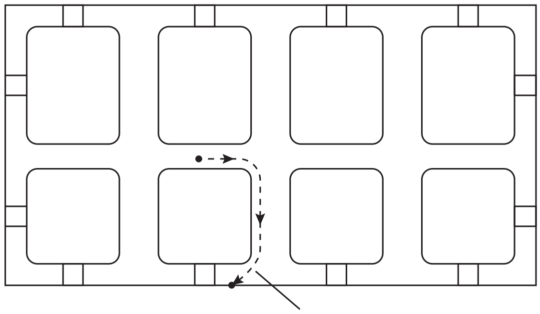

FIGURE 10.45

Schematic representation of a hollow casting with internal walls and small ports to the outside world, such as an automotive cylinder head, illustrating the long diffusion path for heat from its centre during a quench, leading to internal residual tension.

and blocks and housings for components such as compressors and pumps. When immersed in the water quench, the water attempts to penetrate the entrances into the hollow interior of the casting. However, because the casting is originally above 500°C, any water that succeeds in entering will convert almost instantaneously to steam, blowing out any additional water that is attempting to enter. The result is that the outside of the casting in contact with the quenchant cools rapidly, whereas the interior can cool only at the rate that thermal conduction will conduct the heat along the tortuous path via interior walls of the casting to the outer surfaces.

The rate of conduction of heat from the interior to the exterior of the casting can be estimated from the order of magnitude relation

$$x = (Dt)^{1/2} \tag{10.22}$$

where x is the average diffusion distance, D is the thermal diffusivity of the alloy and t is the time taken. The thermal diffusivity is defined as

$$D = K/\rho C \tag{10.23}$$

where K is the thermal conductivity, about 200 $\mathrm{Wm^{-1}K^{-1}}$ for aluminium, the density ρ is about 2700 $\mathrm{kgm^{-3}}$ and the specific heat C is approximately 1000 $\mathrm{Jkg^{-1}K^{-1}}$. These values yield a value for the thermal diffusivity D close to $10^{-4}\ \mathrm{m^2s^{-1}}$. (The corresponding value for steel is approximately $10^{-5}\ \mathrm{m^2s^{-1}}$.) Equation (10.22) is used to generate Figure 10.46 in which the distance for diffusion of heat out of a product indicates the approximate boundaries of safe regimes constituting conditions in which sufficient time is available for the diffusion of heat from the interior during the quench. The time of cooling in different quenchants is provided by results such as that shown in Figure 10.47. These results were obtained by siting a thermocouple in the centre of a 10mm wall of an Al-7Si-0.4Mg alloy casting. Comparative results would be valuable for ferrous materials.

For Al alloys, the temperature range of approximately 500°C down to 250°C appears to be the critically important range during cooling because any beneficial effects of solute retention occur in this range and little can be affected at temperatures below 250°C.

For a solid aluminium bar of 20 mm diameter (approximately equivalent to a solid plate of 10 mm thickness), Figure 10.47 indicates that quenching in water will reduce the temperature in its centre from 500 to 250°C in about 5 s. Substituting 5 s in Eqn (10.22) shows that on average heat will have travelled 20 mm in this time. The 20 mm bar or 10 mm plate will both therefore enjoy a reasonably uniform temperature so that minimal stress will be generated.

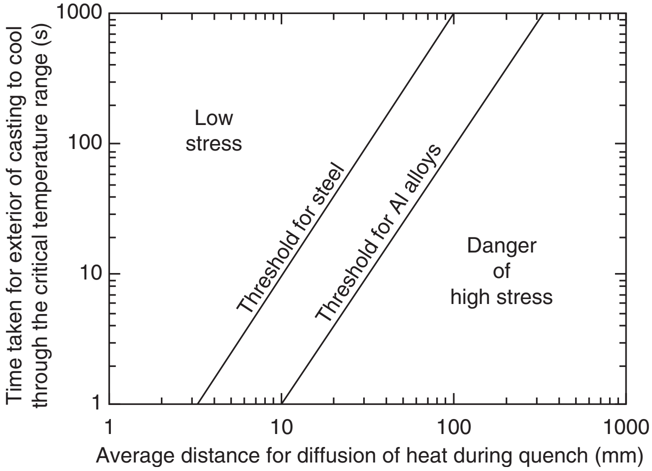

FIGURE 10.46

Regime for low stress in terms of quench rate and distance for heat flow.

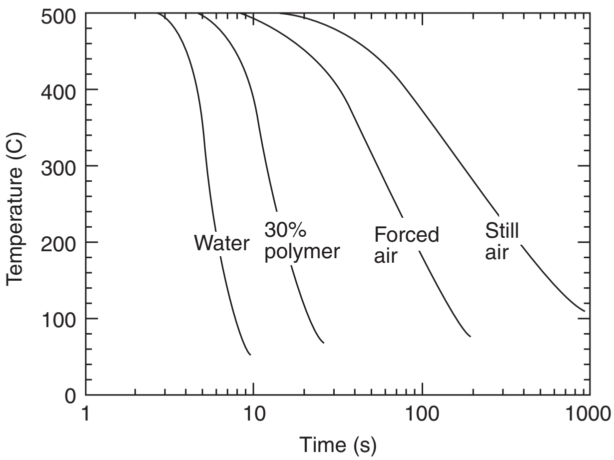

FIGURE 10.47

Quench rates in a 10 mm thick Al plate casting in a variety of quench media.

In such castings as automotive cylinder heads, whose links between the internal sections and the outside world are via small holes in the outer walls and tortuous routes around the water jackets the distance that heat has to diffuse from the centre of the casting is of the order of 100 mm compared with the outside walls which are 10 mm or less. Figure 10.47 indicates that the cooling times are approximately 100 and 5 s, respectively. This large difference in the cooling and contraction results in internal tensile stress over the yield point and well into the plastic flow regime.

10.9.4 CONTROLLED QUENCHING USING POLYMER AND OTHER QUENCHANTS

Figure 10.47 shows that there are other intermediate quenching rate options available. The use of water, whilst being cheap and environmentally pleasant, causes problems for the quenching of most alloys, whether light alloys or ferrous.

The rapidity of the quench is not suitable for larger parts, especially hollow parts with internal structures, as we have seen previously. In addition to this problem, water gives an uneven and non-reproducible quench because of its boiling action. When the parts are immersed in the water, they are at a temperature hundreds of degrees above the boiling point. Thus the water in contact with the hot surface boils, coating it with a layer of vapour that conducts heat poorly, temporarily insulating that area whilst surrounding areas that may happen to remain in contact with the water continue to cool rapidly via a nucleate boiling regime (Figure 10.48; this figure is the result of an investigation by Xiao et al., 2010 with a window-frame shaped casting of 122 mm square with sections of 10–25 mm). The pattern of contact varies rapidly and irregularly as the vapour film forms and collapses in the turbulent water. Agitation of the water is often used to even the cooling effect, so far as possible. However, Figure 10.48 shows that agitation has an effect only on the vapour blanket stage but is not a major effect, and has no effect at all on the nucleate boiling stage which is the process by which heat transfer occurs at maximum rate.

Thus the resulting stress pattern in the casting is complex and different from casting to casting. This is expected to be especially true when the castings are stacked closely in a heat treatment basket because those in the centre of the basket will experience quite different cooling conditions to those on the outside.

To overcome the blanketing action of the vapour, liquids with higher boiling point such as oils have been used. However, the flammability hazard and the smoke and fumes have caused such quenchants to become increasingly unacceptable. Cleaning of the casting after the quench is also an environmental problem. Water-based solutions of polymers have therefore become widely used. They are safer and somewhat less unpleasant in use. Fletcher (1989) reviews their action in detail. We shall simply consider a few general points.

Some polymers are used in solution in water and appear to act simply by the large molecular weight and length of their molecules increasing the viscosity and the boiling point of the water. Such viscous liquids are resistant to boiling and so provide a more even quench, with the quenchant remaining in better contact with the surface of the casting.

Sodium polyacrylate solution in water produces cooling rates similar to those of oils. However, its action is quite different. It seems to stabilise the vapour blanket stage by enclosing the casting in a gel-like casing (Fletcher and Griffiths, 1995).

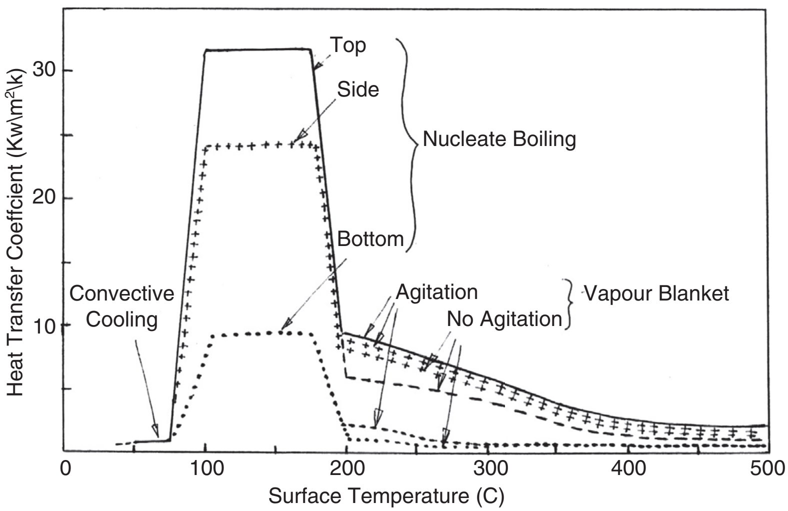

FIGURE 10.48

Heat transfer coefficient (HTC) measured for a small frame casting quenched into water at 50°C. HTC is a strong function of temperature and orientation, but is relatively little influenced by agitation except at high temperature.

Smoothed data from Xiao et al. (2010).

Other polymers have a so-called reverse temperature coefficient of solubility. This lengthy technical description means that the polymer becomes less soluble as the temperature of the solution is raised. Many, but by no means all, of the polymers are based on glycol. One widely used polymer is polyalkylene glycol. This material becomes insoluble in water above about 70°C. The commercial mixtures are usually sold already diluted with water because the product in its pure form would be intractably sticky, like rather solid grease, and would therefore present practical difficulties on getting it into solution. It is usually available containing other chemicals such as antifoaming agents, corrosion inhibitors, and possibly anti-bacteriological components.

Such polymers have an active role during the quench. When the quenchant contacts the hot casting, the pure polymer becomes insoluble. It separates from the solution and precipitates both on the surface of the casting, and in the hot surrounding liquid as clouds of immiscible droplets. The sticky, viscous layer on the casting, and the surrounding viscous mixture, inhibit boiling and aid the uniform cooling condition that is required. When the casting has cooled to below 70°C, the polymer becomes soluble once again in the bulk liquid, and can be taken back into solution. Re-solution is unfortunately rather slow, but the agitation of the quench tank with, for instance, bubbles of air rising from a submerged manifold to scrub the sticky residue off the castings, reduces the time required.

Polymer quenchants have been highly successful in reducing stresses in those castings that are required to be quenched as part of their heat treatment. The properties developed by the heat treatment are also found to be, in general, more reproducible. Capello and Carosso (1989) has shown that the elongation to failure of sand-cast Al-7Si-0.5Mg alloy, using 2.5 times the standard deviation to include 99% of expected results, exhibits greater reliability, as shown in Table 10.1. Thus the average properties that are achieved may be somewhat less than those that would have been achieved by a cold water quench, but the products have the following advantages:

1. The properties of heat treated castings are more uniform and reproducible.
2. The minimum values of the random distribution of strength values are raised.
3. With castings nearly free from stress, the user has the confidence of knowing that all of the strength is available and that an unknown level of internal stress is not detracting from the strength indicated (misleadingly) by a test bar.
4. The castings will have significantly reduced distortion.

Table 10.1 Elongation to Failure Results from Different Quenching Media

	Elongation (%)	
	Minimum	**Mean ± 2.5σ**
Hot-water quench (70°C)	2.01	4.73 ± 2.72
Cold-water quench	4.80	6.47 ± 1.67
Water-glycol quench	4.85	5.81 ± 0.96

These authors carried out quenching tests on an aluminium plate 150 × 100 × 1.5 mm, and found that, taking the distortion in cold water as 100%, a quench with water temperature raised to 80°C reduced the distortion only marginally to 86% of its previous value. Quenching in a mixture of 20% glycol in water gave a distortion of only 3.5%.

Ramesh and Prabhu (2013) introduce a dimensionless cooling performance parameter for the characterisation of quench media. For cylindrical metal bars

$$\text{Cooling parameter} = D^2R/\alpha\Delta T$$

Where D is the diameter of the bar (m), R is cooling rate (K/s), α is its diffusivity (m^2/s) and ΔT is the difference between the initial temperature of the bar and the temperature of the quench medium (K). This quantifying parameter might help to clarify future work in this field. In the meantime, we continue to have to live with some uncertainties as described later.

For many castings, the use of a boiling water quench seems to be of limited help in reducing quench stress, even though some workers claim it to be useful (for instance, Godlewski et al., 2013). Thus, although the rate of quench may be reduced by the use of hot or boiling water, the results are not always reliable. This is almost certainly the consequence of the variability of the vapour blanket that forms around the hot casting. The blanket forms and disappears irregularly, depending on many factors including the precise geometry of the part, its inclination during the quench, and the proximity of other hot castings etc. In addition, from a practical point of view, a hot water quench is costly to install, run and maintain.

For steels, other quenching routes to achieve a low stress casting have been developed involving the use of an intermediate quench into a molten salt at some intermediate temperature of approximately 300°C for approximately 20 s before the final quench into water (Maidment et al., 1984). Despite the advantages claimed by the authors, the expense and complexity of this double quench are likely to keep the technique reserved for aerospace components, if anything.

Quenches into fluid beds at different temperatures almost certainly deserve more attention for the control of quench rates (Ragab et al., 2013). Hard and wear-resistant synthetic aggregates would be expected to make good fluid media to limit the formation of dust. There may be some castings which may not be suitable if their geometry has large horizontal planes which suppress the fluidisation, but for the majority of castings the method achieves excellent heat transfer and good environmental advantages.

However, it is with regret they I conclude that the polymer quench is not currently a long-term practical solution for production for an automotive product. The polymer solution has to be cleaned out of internal cavities where it could lodge; otherwise, it would become concentrated and carbonised during the subsequent ageing treatment, creating copious fumes that pour from the ageing furnace and spread throughout the foundry. In addition, any residual core sand in such locations would be effectively bonded into place by this concentrated polymer binder, and would be practically impossible to dislodge quickly. Perversely, such bonded-in sand is likely to cause damage later in the life of the engine when it finally decides to free itself and starts to clog oil passages and destroy bearings. Finally, the use of polymer in the foundry environment is notoriously messy; splashes from the quench or from the rinse tank cause the surrounding equipment to become black and sticky, and your feet tend to stick to the floor.

In contrast to its problems with automotive castings, polymer is excellent for aerospace castings where the extra trouble to clean each casting individually does not outweigh the benefit of superb heat treatment response and reduced internal stress.

There seems to be little information on another quenchant that holds significant promise: a water/carbon dioxide solution. The mixture is of course highly environmentally friendly. Different concentrations of CO_2 provide different degrees of control over the rate and uniformity of the quench. It seems likely that at least some of the action of the solution is to break up the vapour blanket, making the quench more uniform. More information on this interesting and attractive technique would be welcome.

In a further development, the gas supply companies have been investigating such CO_2 quenchants, combining them with inert nanoparticles such as alumina to form liquid muds. These are claimed to give sufficiently slow quenches to avoid the well-known problems of the quench-cracking of highly alloyed steels (Stratton, 2010).

10.9.5 CONTROLLED QUENCHING USING AIR

The author recalls with pain memories of quenching complex cylinder heads into water, requiring the consequential banana-shaped products to be straightened with a huge 50,000 kg press specially bought-in to rectify the damage. The castings subjected to straightening were those that appeared to have survived failure from cracking in the quench itself (although internal cracks inside the water jackets were often difficult to detect, being usually only found on subsequent return of service failures). In addition, it was found, the castings failed by fatigue in service after only short lives. (These were highly stressed racing cylinder heads so that the disappointment of failure was to some extent compensated by the rare, if not unique, benefit of rapid feed-back of service life data.)

It was this experience that drove the author to the air quenching of cylinder head castings. When the cylinder head casting, considered previously, was subjected to quenching in a blast of air, Figure 10.47 indicates that cooling was now a leisurely 100 s or so. Thus sufficient time was available for the internal sections to lose their heat to the outside so that the casting maintained a reasonably uniform temperature during the quench. The generation of high internal stress and resulting distortion was avoided.

Air quenching proved to be a complete solution for automotive castings. It was clean, quick, environmentally friendly, low cost and quickly and easily implemented in a series production environment. The castings retained their accuracy, and quench failures and fatigue failures disappeared. We were able to restore productivity and profitability (and sell the 50,000 kg straightening press to get our money back).

10.9.6 STRENGTH REDUCTION BY HEAT TREATMENT

Figure 10.49 illustrates how the overall effective strength of a casting can be reduced by a heat treatment designed to strengthen the alloy.

Figure 10.49(a) shows the stress-strain curve for the alloy, and the imposition of 1% tensile strain on the inner parts of the casting as a result of a water quench. This quench strain results in a quench stress close to the failure stress of the material. If no ageing treatment is carried out, this stress is locked into the component for the rest of its life. Naturally, it has little residual strength left, and is likely to fail on the first application of a stress in service.

However, after an ageing treatment designed to double the yield strength of the alloy, the situation is shown in Figure 10.49(b). Assuming the benefit of a small amount of stress relief (the amount indicated in the figure may be rather generous), the residual quench stress is only slightly lower; substantially unchanged. If additional service stress in tension is applied to the central parts of the casting, the residual tensile stress in these parts is effectively a starting point for the additional loading. Thus, effectively, the new stress-strain curve for the component is shown in Figure 10.49(c). It is clear that the new overall stress-strain capability of the casting has been reduced compared to the original unheat-treated material; thus, tragically, as a result of our lengthy, complex and costly heat treatment the component is effectively weaker.

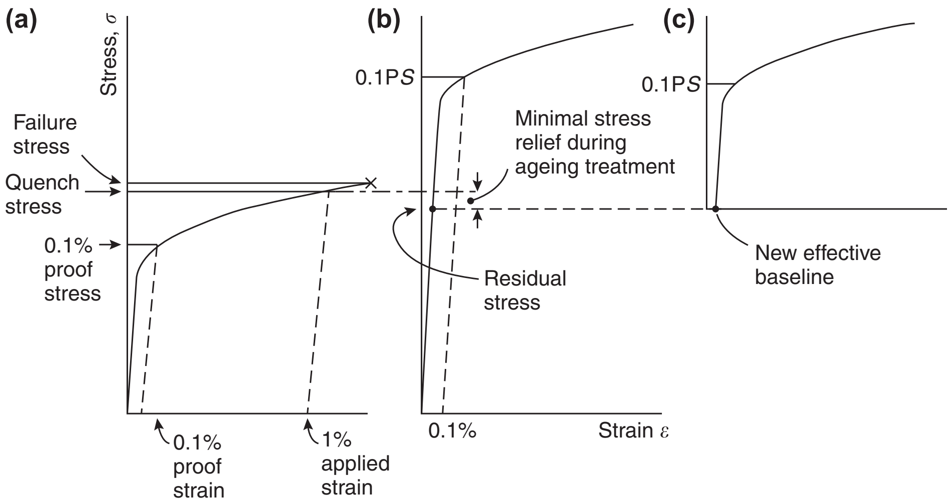

FIGURE 10.49

Evolution of the stress/strain properties of a precipitation-hardened Al alloy as heat treatment progresses: (a) after quench; (b) after ageing to double the yield strength; (c) the final effective stress/strain relation after allowing for the presence of residual stress. Although clearly the alloy is stronger, the load bearing ability of the casting has been reduced.

In summary, for certain sensitive castings such as automotive cylinder heads, the residual stress in aluminium alloy castings quenched into water in this way are well above the yield point of the alloy. Even after the strengthening during the ageing treatment, the stress remains at between 30 and 70% of the yield stress, with a useful working approximation being 50%. Thus the useful strength of the alloy is reduced from its unstressed state of 100%, down to around 50%. This massive loss of effective strength makes it inevitable that residual tensile stresses are a significant cause of casting failure in service, particularly fatigue failure because the residual stress is always generously above the fatigue limit of the alloy.

Turning now to steels: in contrast to the behaviour of Al alloys, the thermal diffusivity D is approximately 10 times lower, of the order of only $10^{-5}\ m^2s^{-1}$. The reader can quickly show that the corresponding distances to which heat can flow are 7 mm in 5 s but only 30 mm in 100 s. For a given rate of quench, therefore, steels will suffer a higher residual stress (Figure 10.46). Nevertheless, they are much more able to withstand such disadvantages, having not only higher strength, but more particularly, higher elongation to failure (Figure 10.50). Thus, although the final internal stress is high, the steel product is nowhere near the failure condition experienced by the aluminium alloy casting. The aluminium alloy casting also experiences about 1% imposed elongation but has only a few percent, perhaps on occasions even less than 1% elongation before failure. Thus it can fail actually in the quench, or early in service. In contrast, the steel casting has 10–20 times greater elongation (as a result primarily of its reduced bifilm content). Thus although the 1% or so of imposed quench strain resulting from unequal cooling may result in 1% or so of distortion of the steel casting, its condition is far from any dangerous condition that might result in complete failure because enormously greater strain has to be imposed to reach a failure condition (Figure 10.50).

The previous statements are so important they are worth repeating in different words for additional clarity. The rapid quenching of steels for metallurgical purposes (such as the stabilisation of austenite for Hadfield Manganese steel) is not usually a problem. The reason is that most steels are particularly clean because of the rapidity with which entrainment defects are deactivated and/or detrained after pouring with the result that they typically have

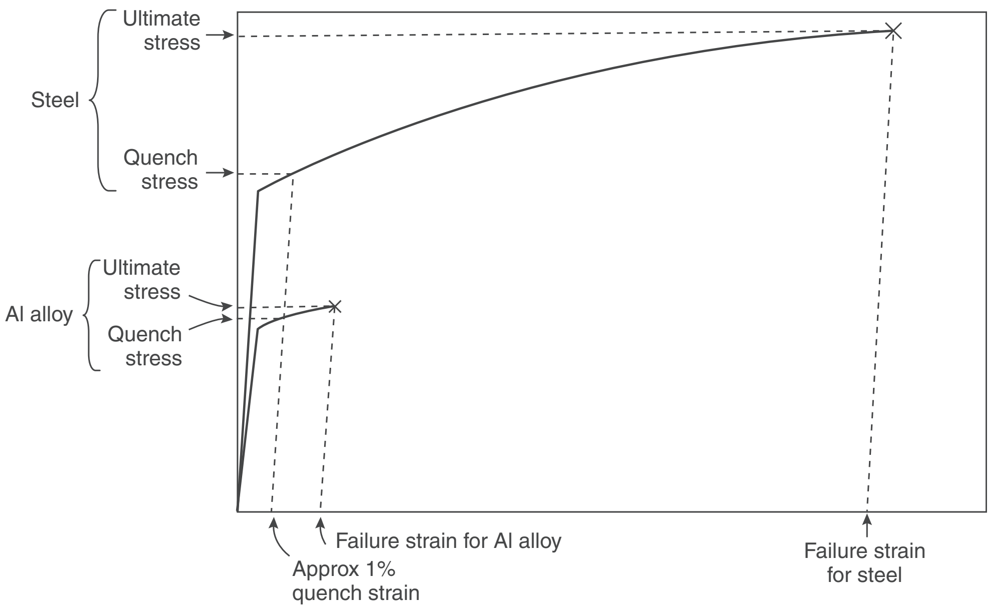

FIGURE 10.50

Comparison of the stress/strain relations for an Al alloy and a steel illustrating the relatively dangerous condition of the Al alloy after a quench.

elongations to failure of 40–50%. In contrast, most Al alloys (and probably most Mg alloys) do not enjoy this benefit; suffering from a high density of neutrally buoyant bifilms they typically achieve less than a tenth of this ductility. Thus, the application of 1% strain takes the aluminium alloy close to, or even sometimes in excess of its failure strain. For steels, even though the 1% strain applied by the quench will take the part into the plastic region, causing huge stresses, the steel remains safe; its greater freedom from bifilms permits it to endure enormously greater extension before it will fail (Figure 10.50).

For the future, the production of Al alloys with low bifilm concentration promises to offer ductilities in the range of that of steels. Already, good foundries know that high strengths together with elongations of 10–20% are achievable if good care is taken.

Slower quenching techniques are safer, although, of course, the strength attained by the heat treatment is somewhat reduced. Even so, the reduced mechanical strength when using slower and more controlled quenches such as a polymer or a forced air quench is more than compensated by the benefit of increased reliability from putting unstressed (or more accurately, low-stressed) castings into service. Thus, the casting designer and/or customer needs to accept somewhat reduced mechanical strength and hardness requirements in order to gain a superior performance from the product. The reductions of strength and hardness are expected to be in the 5–10% range, but the improvement in casting performance can be expected to be approximately 100% or more (as a result of avoiding the loss on the order of 50%). These are huge benefits to be gained at no extra cost.

The proper development of quenching techniques to give maximum properties with minimum residual stress is a technique known as quench factor analysis. It is also much used to optimise the corrosion behaviour of aluminium alloys. The method is based on the integration of the effects of precipitation of solute during the time of the quench. In this way, any loss of properties caused by slow quenching or stepped quenching can be predicted accurately. The interested reader is recommended to the nice introduction by Staley (1981), and his later more advanced treatment (Staley, 1986).

10.9.7 DISTORTION

Residual stresses in castings are not only serious for parts that require the ability to withstand stress in service: they are also of considerable inconvenience for parts that are required to retain a high degree of dimensional stability. This problem was understood many years ago, being first described as early as 1914 in a model capable of quantitative development by Heyn. His model was the three-bar casting shown in Figure 10.42. The internal stresses were represented by two outer springs in compression, each carrying half of the total load of internal compressive stress and an inner spring in tension carrying all of the internal tensile stress. If one of the surfaces of the casting was machined away, one of the external stresses would be eliminated. It is predictable therefore that the casting would deform to give a concave curvature on the machined side as illustrated in the figure.

The distortion of castings both before and after machining is a common fault and typical of castings that have suffered a water quench. Once again, it is a problem so frequently encountered that I have, I regret, wearied of answering the telephone to these enquires too. After all, it is difficult to understand how a casting could avoid distortion if parts of it are stressed up to or above its yield point.

For light alloy castings in particular, a more gentle quench, avoiding water (either hot or cold) and choosing polymer or air will usually solve the problem instantly. As mentioned briefly previously, polymer performs well for aerospace castings but is expensive and messy, whereas air is recommended as being clean, economical and practical for high volume automotive work. Otherwise, stress relieving of castings by heat treatment before machining is strongly recommended. In all cases, of course, some fraction of the *apparent* strength of the product has to be sacrificed.

10.9.8 HEAT TREATMENT DEVELOPMENTS

Although not strictly relevant to the question of reducing residual stresses, it is worth emphasising the newer developments in heat treatments that give approximately 90% or more of total attainable strength, but with much reduced stress and greatly reduced cost. The reduced cost is always an attention-grabbing topic and materially helps the introduction of technology that can deliver an improved product.

Figure 10.51 illustrates the progression of recent developments in heat treatment of Al alloys where the problem of stress is central.

The traditional full heat treatment of a precipitation-hardened alloy, that constitute the bulk of cast structural components at this time, consists of a solution treatment, water quench and age as illustrated in Figure 10.51(a). The treatment results in excellent apparent strength for the material, but is energy intensive in view of the long total times, and the water quench may be severely deleterious (depending on the geometry of the casting) as we have seen.

Illustration Figure 10.51(b) shows how the traditional treatment can be reduced significantly in modern furnaces that enjoy accurate control over temperature, thereby reducing the risk of overheating of the charge because of random thermal excursions. An increase in temperature by 10°C will allow, to a close approximation, an increase in the rate of treatment by a factor of 2. Thus times at temperature can be halved. These benefits are cumulative, such that a rise of 20°C will allow a reduction in time by a factor of $2 \times 2 = 4$, or a rise of 30°C a reduction in time of a factor $2 \times 2 \times 2 = 8$, although, of course, there are clearly limits to how far this simple relation can be pushed. Even so, it seems to hold within acceptable accuracy over many decades of temperature change. The reader will appreciate that the tiny additional energy required by the higher temperature is of course generously offset by the savings in overall time at temperature.

In Figure 10.51(a) and (b) require separate furnaces for solution and ageing treatments (if long delays waiting for the solution furnace to cool to the ageing temperature are to be avoided). Thus floor space requirement is high. Floor space requirement is increased further by the quench station, and, if a polymer quench is used, by a rinsing tank station.

If an air quench is used to gain the benefits of reduced residual stress, the additional benefits to the overall cycle time are seen in Figure 10.51(c); the quench is now easily interrupted and the product transferred to the ageing furnace already

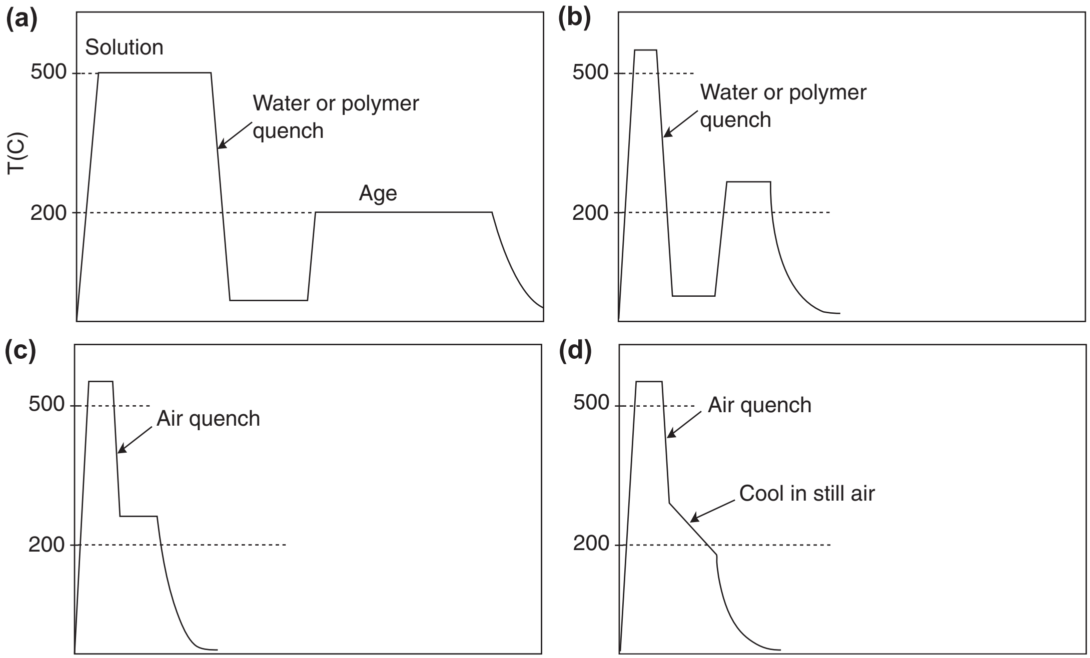

FIGURE 10.51

A progression of developments for the precipitation heat treatment of Al alloys. (a) Traditional full T6 heat treatment giving excellent properties but taking 12–24 h; (b) shortened treatment giving nearly equivalent results; (c) use of air quench to reduce time, energy and residual stress; (d) a possible ultimate short and simple cycle, ageing during the high temperature cool in still air.

at the correct temperature for ageing, saving time and reheat energy. Additional benefits include the fact that the air quench is environmentally friendly; the castings are not stained by the often less-than-clean water; the conveyor is straightforward to build and maintain; there is no mechanism required for lowering into water that normally results in complex and rusting plant. As we have repeatedly emphasised, the products from this type of furnace have somewhat lower apparent strengths and hardness, but greatly improved performance in service.

Figure 10.51(d) shows an ultimate system that might be acceptable for some products. The ageing treatment is simply carried out by interrupting the air quench slightly above the normal ageing temperature, and allowing the part to cool in air (before final rapid cooling by fans if necessary). This represents a kind of natural ageing process in which no ageing furnace is required. Strengths will suffer somewhat, but the lower costs and simplicity of the process may be attractive, making the process suitable for some applications.

10.9.9 BENEFICIAL RESIDUAL STRESS

Not all residual stress need be bad.

Bean and Marsh (1969) describe a rare example in which the stress remaining after quenching was used to enhance the service capability of a component. They were developing the air intake casing for the front of a turbojet engine. The casting has the general form of a wheel, with a centre hub, spokes, and an outer shroud. In service the spokes reached 150°C and the shroud cooled to −40°C. The expansion of the spokes and contraction of the shroud gave problems in service. With additional high loads from accelerations up to 7*g* and other forces, some casings were deformed out of round, and some even cracked. To counter this problem, the casting was produced with tensile stress in the spokes and compressive loading in the shroud. This was achieved by wrapping the spokes in glass fibre insulation, whilst allowing the shroud to cool quickly at the full quenching rate, but the spokes to cool and contract later. By this means, approximately 40 MPa tensile stress was introduced into the spokes. This was tested by cutting a spoke on each fifteenth casting, and measuring the gap opening of approximately 2 mm.

Another method of equalising quenching rates in castings is by the clamping of shielding plates around thinner sections to effectively increase their section. The method is described by Avey et al. (1989) for a large circular clutch housing in a high-strength aluminium alloy. The technique improved the fatigue life of the part by more than 400%.

It may be significant that both of these descriptions of the positive use of residual stress relate to rather simple rotationally symmetric castings.

10.9.10 STRESS RELIEF

For traditional grey iron foundries of the past century, the original method of providing some stress relief in iron castings was simply to leave the castings in the foundry yard. Here, the long passage of time, of weeks or months, and the changeable weather, including rain, snow, frost and sun, would gradually do its work. It was well known that the natural ageing outdoors was more rapid and complete than ageing done indoors, presumably because of the more rapid and larger temperature changes.

Nowadays, the more usual method of reducing internal stress is both faster and more reliable (although somewhat more energy intensive!). The casting is reheated to a temperature at which sufficient plastic flow can occur by creep to reduce the strain and hence reduce the stress. This is designed to take place within a reasonable time, of the order of an hour or so. As pointed out earlier, it is then most important that stress is not re-introduced by cooling too quickly from the stress-relieving treatment.

An apparently perverse and quite exasperating feature of internal tensile stress in castings is that the casting will often crack while it is being reheated as part of the stress-relieving process to avoid the danger of cracks! This happens if the reheating furnace is already at a high temperature when the castings are loaded. The reason for this is that if the casting already has a high internal tensile stress, on placing the casting in a hot reheating furnace the outside will then be heated first and expand, before the centre becomes warm. Thus the centre, already suffering a high stress, will be placed under additional tensile load, the total being sometimes sufficient to exceed the tensile strength.

Table 10.2 Approximate 1-h Stress-Relieving Temperatures

Alloy	Stress-Relief Temperature (°C)
Al-2.2Cu-1Ni-1Mg-1Fe-1Si-0.1Ti	300
Brass Cu-35Zn-1.5Fe-3.7Mn	400
Bronze Cu-10Sn-2Zn	500
Grey iron 3.4C-2Si-0.38Mn-0.1S-0.64P	600
Steel (C-Mn types)	700

The problem is avoided by reheating sufficiently slowly that the temperature in the centre is able, within tolerable limits, to keep pace with that at the outside. Consideration of the thermal diffusivity using Eqn (10.22) will give some guidance of the times required.

Figure 10.52 shows the temperatures required for stress relief of various alloys. (Strictly, the figure shows results for 3 h, but the results are fairly insensitive to time, a factor of two reduction in time corresponding to an increase in temperature of 10°C, hardly moving the curves on the scale used in the figure.) It indicates that nearly 100% of the stress can be eliminated by an hour at the temperatures shown in Table 10.2.

Bhaumik et al. (2010) describes experiments to prove an elegant and simple technique to determine the times and temperatures needed to achieve specific degrees of stress reduction. The technique is known as a stress relaxation test. A tensile test bar of the material is mounted in a tensile testing machine, and the test bar is enclosed by a heater. The test bar is loaded to near its yield stress while the heater is raised to the test temperature. The tensile test machine will then automatically record the fall in stress with time. At low temperatures, the times to achieve a 90% drop in stress might take hours or days, whereas at high temperatures this degree of stress relief might take only minutes. This simple, accurate, quantitative test is strongly recommended.

There are numerous examples of the use of such heat treatments to effect a valuable degree of stress relief. One example is the work by Pope (1965) on cast iron diesel cylinder heads that were found to crack between the exhaust valve seats in service, despite a stress-relief treatment for 2 h at 580°C. A modification of the treatment to 4 h at 600°C cured the problem. From Figure 10.52 and Table 10.2, we can see that even 1 h at 600°C would probably have been sufficient.

The work by Kotsyubinskii et al. (1968) highlights the fact that during the thermal stress relief the casting will distort. They carried out measurements on box-section castings in grey iron, intended as the beds of large machine tools, for which stress-relieving treatment is carried out after some machining of the top and base of the box section. He suggests that the degree of movement of the castings is approximately assessed by the factor $(w_1 - w_2)/w_c$ where w_1 and w_2 are the weights of metal machined from the top and base of the casting, respectively, and w_c is the weight of the machined casting. This interesting observation has, to the author's knowledge, never been confirmed.

Moving on now from heat treatment, there are other methods of stress relief that are sometimes useful. In simple castings and welds, it is sometimes possible to effect relief by mechanical overstrain as described in the excellent review by Spraragen and Claussen (1937).

Kotsyubinskii (1962) describes a further related method for grey iron in which the castings are subjected to rapid heating and cooling between 300°C and room temperature at least three times. The differential rates of heating within the thick and thin sections produce the overstrain required for stress relief by plastic flow.

More drastic heating rates are required to effect stress relief by differential heating in aluminium alloys because of the thermal smoothing provided by the high thermal conductivity. Hill et al. (1960) describe an 'up-quenching' technique in which the casting is taken from cryogenic temperatures, having been cooled in liquid nitrogen, and is reheated in jets of steam. This thermo-mechanical treatment introduces a pattern of stresses into the casting that are opposite to those introduced by normal quenching. One of the benefits of this method is that it is all carried out at temperatures below normal ageing temperatures, so that the effects of the final heat treatment and the resulting mechanical properties are not affected. One of the

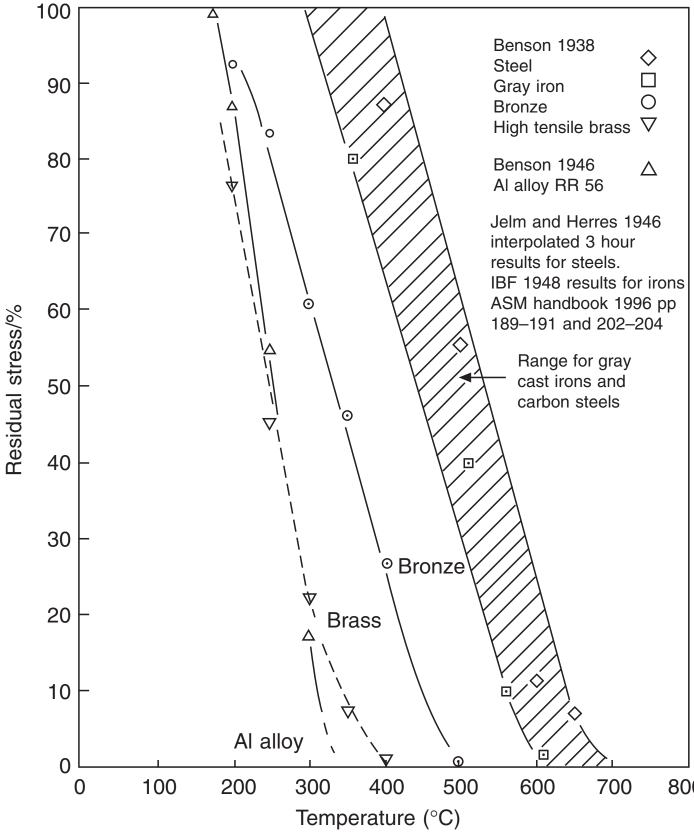

FIGURE 10.52

Stress relief of a selection of alloys treated for 3 h at temperature.

Data from Benson (1938), Jelm and Herres (1946) and Institute of British Foundrymen (1948).

possible disadvantages that the authors do not mention is the enhanced tensile stress in the centre of the casting during the early stages of the up-quench. Some castings would not be expected to survive this dangerous moment.

A variant of these approaches is stress relief by vibration. This is probably effective in some shapes, but it is difficult to see how the technique can apply to all parts of all shapes. This is particularly true if the component is treated at a resonant frequency. In this condition, some parts of the casting will be at nodes (will not move) and some parts at antinodes (will vibrate with maximum amplitude). Thus the distribution of energy in the casting will be expected to be highly heterogeneous. Some investigators have reported the danger of fatigue cracks if vibrational stresses over the fatigue limit are employed (Kotsyubinskii, 1961). The technique may require some skill in its application because null results are easily achieved (IBF Technical Subcommittee, 1960).

There may be greater certainty of a valid result with sub-resonant treatment. This technique emerged later as a possible method of stress relief (Hebel, 1989). In this technique, the casting is vibrated not on the peak of the frequency–amplitude curve, but low on the flank of the curve. At this off-peak condition the casting is said to absorb energy more efficiently. Furthermore, it is claimed that the progress towards complete stress relief can be monitored by the gradual change in the resonant frequency of the casting. When the resonant frequency ceases to change, the casting is said to be fully stress relieved. If this technique could be verified, then it would deserve to be widely used.

10.9.11 EPILOGUE

Although the *strength of the material* will therefore be lowered by a slower quench, the *strength of the component* (i.e. the failure resistance of the complete casting acting as a load bearing part) in service will be increased.

If water quench is avoided with a view to avoiding the dangers of internal residual stress, it is common for the customer to complain about the 5% or so loss of apparent properties. In answer to such understandable questions, an appropriate reply to focus attention on the real issue might be 'Mr. Customer, with respect, do you wish to lose 5% or 50% of your properties?'

In the experience of the author, several examples of castings that have been slowly quenched, losing 5–10% of their strength, are demonstrated to double their performance in service. This was the case for a roadside compressor casting subjected to a full T6 heat treatment. The compressor housing exploded in service, fortunately not resulting in any injury to passersby. After the second explosion the manufacturer requested help. I recommended a stress relief treatment. This was declared to be impossible because it was claimed the compressor housing needed maximum strength. Despite the reservations of the manufacturer, stress relieved castings (that had only half the strength of the fully heat treated product) were tested to destruction in comparison with fully heat-treated castings. The stress-relieved products proved twice as resistant to failure under pressure compared with the fully heat-treated parts with the metallurgically 'strong' alloy.

Finally, therefore, it remains deeply regrettable, actually a scandal, that many national standards for heat treatment continue to specify water quenching without any warning of the dangers for certain geometries of casting. This disgraceful situation requires to be remedied. In the meantime, the author deeply regrets to have to recommend that such national standards be set aside. It is easy for the casting supplier to take refuge in the fact that our international and national standards on heat treatment often demand quenching into water and thereby avoid the issue that such a production practice is risky for many components—and in any case provides the user with a casting of inferior performance. However, the ethics of the situation are clear. We are not doing our duty as responsible engineers and as members of society if we continue to ignore these crucial questions. We threaten the performance of the whole component merely to fulfil a piece of metallurgical technology that from the first has been woefully misguided.

Our inappropriate heat treatments have been costly to carry out and have resulted in costly failures. It has to be admitted that this has been nothing short of a catastrophe for the engineering world for the past half-century, and particularly for the reputation of light-alloy castings, not to mention the misfortune of users. As a result of the unsuspected presence of bifilms, they have suffered poor reliability so far, but as a result of the unsuspected presence of residual stress, this has been made considerably worse by an unthinking quest for material strength regardless of component performance.

10.10 RULE 10: PROVIDE LOCATION POINTS

This rule is added simply because the foundry can accomplish all the other nine rules successfully, and so produce beautiful castings, only to have them scrapped by the machinist. This can create real-life drama if the castings have been promised in a just-in-time delivery system. This rule is added to help to avoid such misfortunes and allow all parties to sleep more soundly in their beds.

Before describing location points, their logical precursors are datum planes. We need to decide on our datums first (for language purists the plural of datum is data, but this would introduce unwelcome confusion with the conventional meaning of the word, thus we all need to grit our teeth with the inelegant plural 'datums').

10.10.1 DATUMS

A datum is simply a plane defining the zero from which all dimensions are measured.

For a casting design, it is normal to choose three datum planes at right angles to each other. In this way, all dimensions in all three orthogonal directions can be uniquely defined without ambiguity.

In practice, it is not uncommon to find a casting design devoid of any datum, there being simply a sprinkling of dimensions over the drawing, none of the dimensions being necessarily related to each other. On other designs, the dimensions relate with great rigour to each other and to all machined features such as drilled holes etc., but not to the casting. In yet other instances that the author has suffered, datums on one face have not been related to datums on other views of the same casting. Thus the raft of features on one face of the casting shifts and rotates independently of the raft of features on the opposite face.

It is fortunate today that such poor dimensioning of castings is not easy, and is perhaps impossible, with computer-aided design. Even so, there are good reasons to be cautious.

Figure 10.53(a) shows a sump (oil pan) for a diesel engine designed for gravity die casting in an aluminium alloy. The variations in die temperature and ejection time result in variability of the length of the casting that are well-known with this process, and not easily controlled. The figure shows how the dimensioning of this part was very logically laid out, but has made the part nearly unmanufacturable. Three fundamental criticisms can be made:

1. The datum is at one end of the product. If the datum had been defined somewhere near the centre of the part, then the variability produced by the length changes of the casting would have been approximately halved.
2. There is only one feature on the component whose location is critical; this is the dipstick boss. If the boss is slightly misplaced then it fouls other components on and around the engine. It will be noticed that the dipstick boss is at the far end of the casting from the datum. Thus variability in length of the casting will ensure that a large proportion of castings will be deemed to have a misplaced boss. If the datum had been located at the other end of the casting, near to the boss, the problem would have been reduced to negligible proportions. If the datum had been chosen as the boss itself, the problem would have disappeared altogether, as in Figure 10.53(b).
3. The datum is not defined with respect to the casting. It is centred on a row of machined holes which clearly do not exist at the time the casting is first made and when it is first required to be checked and before any machining of the casting has taken place. Depending on whether the machinist decides to fix the holes in the centre of the flange, or relate them by measurement to the more distant dipstick boss, or to the centre of the casting averaged from its two ends, or any number of alternative strategies, the drilled holes could be almost anywhere, including partly off the flange or even completely off the flange!

Figure 10.53(b) shows how these difficulties are easily resolved. The datum is located against the side of the dipstick boss and hence is fixed in its relation to the casting and goes some way to halving errors in the two directions from this

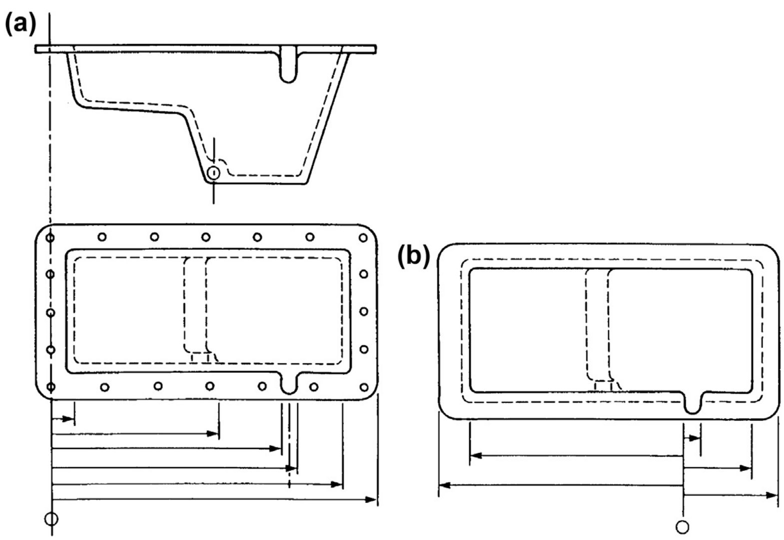

FIGURE 10.53

(a) A badly dimensioned sump casting, resulting in a casting manufactured only with difficulty and high scrap losses.
(b) A change of datum to the most critical feature of the casting results in easy and efficient manufacture.

plane. It also means that the dipstick boss itself is now impossible to misplace, no matter how the casting size varies; all the other dimensions are allowed to float somewhat because movement of other points on the casting is not a problem in service. This part can then be produced easily and efficiently, without trauma to either producer or customer!

In summary, the rules for the use of datums (partly from Swing (1962)) are:

1. Choose three orthogonal datum planes.
2. Ensure the planes are parallel to the axes of motion of the machine tools that will be used to machine the part (otherwise unnecessary computation and opportunity for error is introduced).
3. Fix the planes on real casting features, such as the edges of a boss, or the face of a wall (i.e. not on a centreline or other abstract constructional feature). Choose casting features that are
   a. critical in terms of their location and
   b. as near the centre of the part as possible.

10.10.2 LOCATION POINTS

Location points are those tiny patches on the casting that are used to locate the casting precisely and unambiguously in three dimensions. They are required by all parties involved in the manufacture of the product including.

1. the toolmaker because he can construct the tooling with reference to them;
2. the founder, to check the core and mould assembly if necessary and the casting once it has been made, and
3. the machinist, who uses them to locate the casting before the first machining operation.

These features therefore integrate the manufacture of the product, ensuring its smooth transmission as it progresses from toolmaker to founder to machinist.

Whereas the casting datums are invisible planes that define the concept of a zero in the dimensional space in and around a casting, the tooling points are real bits of the casting. The datums are the software, whereas the tooling points are the hardware, of the dimensioning system.

It is useful, although not essential, for the datums to be defined coincident with the tooling points.

The location points are required to be *actual cast-on features* of the casting. This point cannot be over-emphasised. It is not helpful, for instance, to define a location feature as a centre-line of a bore. This invisible feature only exists in space (perhaps we should say 'free air'). Virtual features such as centrelines have to be found by locating several (at least three) points on the internal as-cast solid surface of the bore, and its centre thereby calculated. Clearly, these 'virtual' or 'free-air' so-called location points necessarily rely on their definition from other nearby as-cast surfaces. These ambiguities are avoided by the direct choice of as-cast location features.

The location features need to be cast nicely, without obscuring flash, or burned-on sand, and definitely should not be attacked by an enthusiastic finisher wielding an abrasive wheel.

It is essential that the location points are *not* machined. If they are machined, the circumstance poses the infinitely circular question 'What prior datums were used to locate and position the casting accurately to ensure that the machining of the machining locations (from which the casting would be picked up for machining) were correctly machined?' Unfortunately, such indefensible nonsense has its committed devotees.

Location points are known by several different names, such as machining locations (which is a rather limiting name) or pickup points. On drawings, tooling point, called TP, is used as the common abbreviation for the drawing symbol. Although in practice I tend to use all the names interchangeably it is proposed that 'location points' describes their function most accurately and will be used here.

Before the day of the introduction of location points in the Cosworth engine-building operation, I was accustomed to a complex cylinder head casting taking a skilled man at least 2 h to measure to assess how to pick up the casting for machining. The casting was repeatedly re-checked, re-orienting it slightly with shims to test whether wall thicknesses were adequate and whether all the surfaces required to be machined would in fact clean up on machining. After the 2 h, it was common to witness our most expensive casting dumped unceremoniously on the scrap heap; no orientation could be found to ensure that it was dimensionally satisfactory and could be completely cleaned up on machining. All this changed on the day when the new foundry came on stream. A formal system of six location points to define the position of the casting was introduced. After this date, no casting was subjected to dimensional checking. All castings were received from the foundry and on entering the machine shop, and were immediately thrown on to a machine tool, pushed up against their location points, clamped, and machined. No tedious measurement time was subsequently lost, and no casting was ever again scrapped for machining pickup problems.

It is essential that every casting has defined locations that will be agreed with the machinist and all other parties who require to pick up the casting accurately.

For instance, it is common for an accurate casting to be picked up by the machinist using what appear to be useful features, but which may be formed by a difficult-to-place core, or a part of the casting that requires some dressing by hand. Thus although the whole casting has excellent accuracy, this particular local feature is somewhat variable in location. The result is a casting that is picked up inaccurately, and does not therefore clean up on machining. As a result it is, perhaps rather unjustly, declared to be dimensionally inaccurate.

The author suffered precisely this fate after the production of a complex pump body casting for an aerospace application that achieved excellent accuracy in all respects, except for a small region of the body that was the site where three cores met. The small amount of flash at this junction required dressing with a hand grinder, and so, naturally, was locally ground to a flat, but at various slightly different depths beneath the curved surface of the pump body. This hand-ground location was the very site that the machinist chose to locate the casting. The result was disaster. Furthermore, it was not easily solved because of the loss of face to the machinist who then claimed that the location options suggested by the foundry were inconveniently awkward. The fault was not his of course. The fundamental error lay in not obtaining agreement between all parties before the part was made. If the location point used by the machinist really was the only sensible option for him, the casting engineer and toolmaker needed to ensure that the design of the core package would allow this.

Ultimately, this rule is designed to ensure that all castings are picked up accurately and conveniently if possible, so that unnecessary scrap is avoided.

Different arrangements of location points are required for different geometries of casting. Some of the most important systems are listed later.

Rectilinear systems

1. Six points are required to define the position of a component with orthogonal datum planes designed for essentially rectilinear machining, as for an automotive cylinder head or block. (Any fewer points than six are insufficient to define the position of the casting, and any more than six will ensure that one or more points are potentially in conflict.)

On questioning a student on how to use a six-point system to locate a brick-shaped casting, the reply was ‘Oh easy! Use four points around the outside faces and one top and one bottom.’ This shows how easy it is to get such concepts wildly wrong!

In fact, the six points are used in a 3, 2, 1 arrangement as shown in Figure 10.54. The system works as follows: three points define plane A, two define the orthogonal plane B and one defines the remaining mutually orthogonal plane C. The casting is then picked up on a jig or machine tool that locates against these six points. Example (a) shows the basic use of the system: points 1, 2 and 3 locate plane A; points 4 and 5 define plane B; and point 6 defines plane C. Planes A, B and C may be the datum planes. Alternatively, it is often just as convenient for them to be parallel to the datum planes, but at accurately specified distances away.

Clearly, to maximise accuracy, points 1, 2 and 3 need to define a widely based triangle, and points 4 and 5 similarly need to be as widely spaced as possible. A close grouping of the locations will result in poor reproducibility of the pickup of the casting; tiny errors in the position or surface roughness of the tooling points will be magnified if they are not widely spaced.

Example (b) shows an improved arrangement whereby the use of a tooling lug on the longitudinal centreline of the casting allows the dimensions along the length of the casting to be halved. The largest dimension of the casting is usually subject to the largest variability, so halving its effect is a useful action.

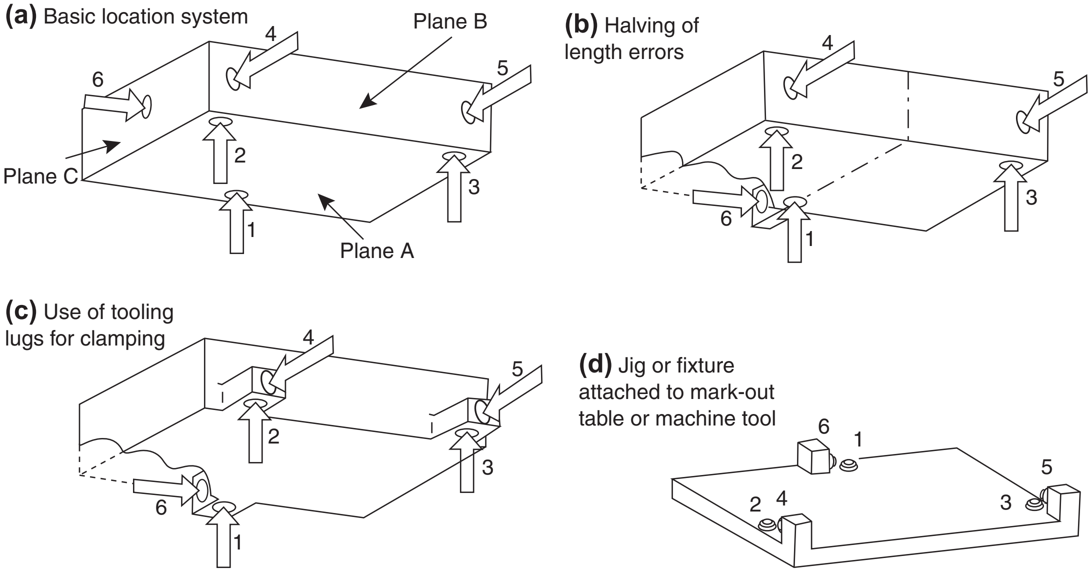

FIGURE 10.54

(a) The six-point location system; (b) halving of length errors; (c) use of lugs to combine tool location with clamping points; (d) a jig to cradle the casting during dimensional checking or machining (clamps are omitted for clarity).

Example (c) is a further development of this idea, creating lugs that serve the additional useful purpose of allowing the part to be clamped immediately over the points of support and off the faces that require to be machined. In practice the lugs may be existing features of the casting, or they may be additions for the purpose of allowing the casting to be picked up for inspection or machining.

A final development of the concept (not shown in our illustration) uses all the lugs arranged on all the centrelines of the casting so as to halve the errors in all directions.

For maximum internal consistency between the tooling points, all points should be arranged to be in one half of the mould, usually the fixed or lower half, although sometimes all in the cope. The separation of points between mould halves, or having some defined from the mould and some from cores, will compromise accuracy. However, it is sometimes convenient and correct to have all tooling points in one internal core, or even one half of an internal core (defined from one half of the core box) if the machining of the part requires to be defined in terms of its internal features.

It is noteworthy that this introduces a small error of principle. The faces of tooling points 1, 2 and 3 are formed in the drag, whereas 4, 5 and 6 are formed in the cope. This is a small error in this case, because the cope-drag joint is among the most reliable of all the mould joints, introducing only a small error in the definition of the location of plane A. It is a good rule to place all six points in one piece of mould, often in the drag, so that they all relate to each other accurately, but because of the small potential for cope/drag error, an exception may be allowable in this case.

In general, it is useful if the tooling locations on the casting can remain in position for the life of the part. It is reassuring to have the tooling points always in place, if only to resolve disputes between the foundry and machine shop concerning failure of the part to clean up on machining. It is therefore good practice to try to avoid placing them where they will be eventually machined off. Using existing casting features wherever possible avoids the cost of additional lugs, and avoids the cost of subsequent removal if necessary.

The definition of the six-point locations, preferably on the drawing (nowadays the computer model), before the manufacture of the casting, is the only method of guaranteeing the manufacturability of the part.

Cylindrical systems

Most cylindrical parts do not fall nicely into the classical six-point rectilinear system as described previously. The errors of eccentricity and diameter both contribute to a rather poor location of the centre using this approach. The unsuitability of the orthogonal pick-up system is analysed nicely by Swing (1962).

In fact, the obvious way to pick up a cylinder is in a three-jaw chuck. The self-centring action of the chuck gives a useful averaging effect on any out-of-roundness and surface roughness, and is of course insensitive to any error of diameter.

In classical terms, the three-jaw chuck is equivalent to a two-point pickup because it defines an axis. We therefore need four more points to define the location of the part absolutely. Three points longitudinally abutting the jaws will define the plane at right angles to the central axis, and one final point will provide a 'clock' location. Figure 10.55 shows the general scheme.

Another location method that is occasionally useful is the use of a V block. This is a way of ensuring that a cylindrical part, or the round edge of a boss, is picked up centrally, averaging errors in the size and, to some extent, the shape of the part. The method has the disadvantage that errors in diameter of the part will cause the whole part to be shifted either nearer to or away from the block, depending on whether the diameter is smaller or larger. (The reader can quickly confirm these shifts in position with rough sketches.)

A widely used but poor location technique is the use of conical plugs to find the centre of a cored hole. Even if the hole is formed by the mould, and so relatively accurately located, any imperfection in its internal surface is difficult to dress out, and will therefore result in mis-location. If a separate core forms the hole then the core-positioning error will add to the overall inaccuracy of location. Location from holes is not recommended.

It is far better to use external features such as the sides of bosses or walls, as previously discussed. These can be more easily maintained cleanly cast.

Triangular systems

For some suitable parts of triangular form, such as a steering gear housing, a useful and fundamentally accurate system is the cone, groove and plane method (Figure 10.56).

Thin-walled boxes

For prismatic shapes, comprising hollow, box-like parts such as sumps (oil pans), the pickup may be made by averaging locations defined on opposite internal or external walls. This is a more lengthy and expensive system of location often tackled by a sensitive probe on the machine tool that then calculates the averaged datum planes of the component and

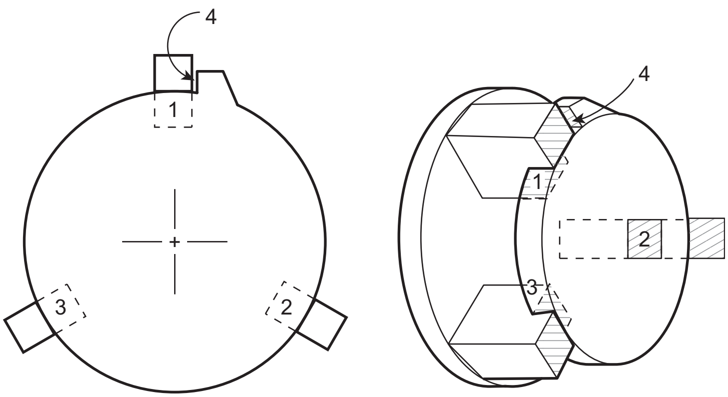

FIGURE 10.55

Use of a three-jaw self-centring chuck for casting location and clamping (note that location 4 is occasionally required as a 'clock' location).

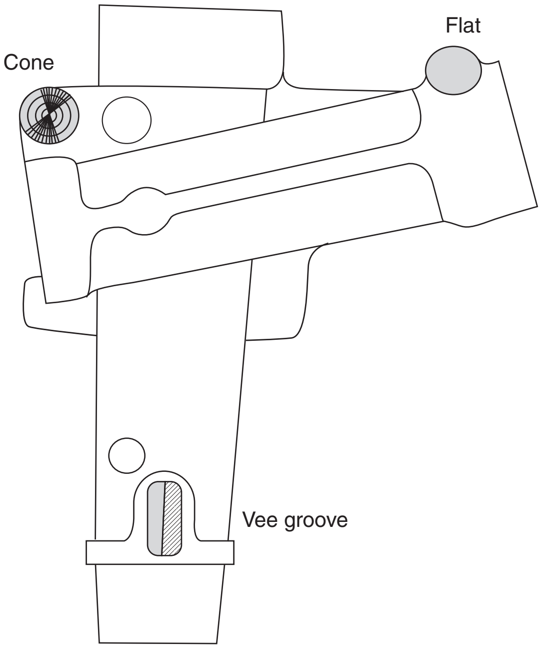

FIGURE 10.56

Plan view of a steering housing for an automobile, showing a flat, a groove and a cone location system.

orients the cutter paths accordingly. This technique is especially useful where an average location is definitely desirable, as a result of the casting suffering different degrees of distortion of its relatively thin walls.

10.10.3 LOCATION JIGS

Figure 10.54(d) illustrates a basic jig that is designed to accept a casting with a six-point location system. The jig is simply a steel plate with a series of small pegs and blocks. It contrasts with many casting jigs, which are a nightmare of constructional and operational complexity.

Our simple jig is also simple to operate. When placing the casting on the jig, the casting can be slid about on locations 1, 2 and 3 to define plane A, then pushed up against locators 4 and 5 to define plane B, and finally slid along locators 4 and 5 until locator 6 is contacted. The casting is then fixed uniquely in space in relation to the steel jig plate. It can then be clamped, and the casting measured or machined. The six locations can, of course, be set up and fixed directly onto, effectively integrated into the machine tool that will carry out the first machining operation.

After the first machining operations, it is normal to remove the casting from the as-cast locations and proceed with subsequent machining using the freshly machined surfaces as the new location surfaces. McKim and Livingstone (1977) go on to define the use of functional datums which may become useful at this stage. They are machined surfaces that normally relate to features locating the part in its intended final application.

Other jigs can easily be envisaged for cylindrical and other shaped parts.

10.10.4 CLAMPING POINTS

During machining, the forces on the casting can be high, requiring large clamping loads to reduce the risk of movement of the casting. Clamping points require to be planned and designed into the casting at the same time as the location points.

This is because the application of high clamping loads to the casting involves the risk of the distortion of the casting, and of spring-back after the release of the clamps at the end of machining. Surfaces machined flat are apt to become curved after unclamping because of this effect.

The great benefit of using tooling lugs as shown in Figure 10.54(c) can therefore be appreciated. The location point and the clamping point are exactly opposed on either side of the lug. In this way, the clamping loads can be high, without introducing the risk of the overall distortion of the casting.

Further essential details of the design of the clamping action include the requirement for the action to move the part on to, and hold it against, the location point.

For softer alloys that are easily indented, the clamp face needs to be 5–10 mm in diameter, similar to the working area of the tooling point. Even so, a high clamping load will typically produce an indentation of 0.5 mm in a soft Al alloy, decreasing to 0.2 mm in an Al alloy hardened by heat treatment, and correspondingly less still in irons and steels.

10.10.5 POTENTIAL FOR INTEGRATED MANUFACTURE

As we have already emphasised, the tooling points should be defined on the drawing of the part (i.e. more usually on the computer model), and should be agreed by (1) the manufacturer of the tooling, (2) the caster and (3) the machinist. It is essential that all parties work from *only* these points when checking dimensions and when picking the part up for machining.

The method allows an integrated approach right from the start of the creation of the tooling, because the pattern or tool maker can use the tooling points as the critical features of the tooling in relation to which all measurements will be defined. The foundry engineer can check his core and mould assembly, and will know how to pick up the casting to check dimensions after the production of the first sample castings. The machinist will use the same points to pick up the casting for machining. They all work from the same reference points. It is a common language and understanding between design, manufacture, and inspection of products. Disputes about dimensions then rarely occur, or if they do occur, are easily settled. Casting scrap apparently resulting from dimensioning faults, or faulty pick up for machining, usually disappears.

This integrated manufacturing approach is relatively easily managed within a single integrated manufacturing operation. However, where the pattern shop, foundry and machinist are all separate businesses, all appointed separately by the customer, then integration can be difficult to achieve. It is sad to see a well-designed six-point pickup system ignored because of apparent cussedness by one member of the production chain.

The industry and its customers very much need purchasing and manufacturing policies based on teamwork and cooperation, together with the adoption of integrated and fundamentally correct systems. We all hope to arrive in this utopia one day.