CHAPTER

CRACKS AND TEARS

8

8.1 HOT TEARING

8.1.1 GENERAL

A hot tear is one of the most serious defects that a casting can suffer. Although it has been widely researched, and is understood in a general way, it has remained a major problem in the foundries, particularly with certain alloys that are especially prone.

Over recent years, the work by Rappaz (1999–2009) has provided illuminating details of the microscopic and mesoscopic behaviour of metals leading to conditions in which hot tearing occurs. More will be said about this elegant work later in this chapter.

Even so, it has to be admitted that important insights into hot tearing behaviour have emerged not only from scientific experiments in the laboratory, but from experience on the shop floor of foundries. However, these important shop floor findings have not been published in the scientific journals.

Briefly, for those who wish to read no further, the key result from the shop floor is that hot tearing can usually be eliminated in castings by simply improving the filling system. The improved systems are described in Chapter 10 onwards.

In the experience of the author, every hot tear, wherever it appears in a casting, can be eliminated by addressing the problems of the filling system. Once oxide bifilms are eliminated from the melt, the solidifying alloy cannot tear, reinforcing my conclusion that there is no such thing as a solidification defect; there are only casting defects. When subjected to a tensile strain by contraction of the casting, material free from bifilms simply stretches because metals are normally extremely soft and ductile at temperatures close to their melting points. It is essential to keep this basic piece of experience in mind while ploughing through the details of this section. Thus although traditional foundrymen cluster around the hot tear, discussing how the mould or core can be reduced in strength to reduce the stress on the contracting casting, my approach is to ignore the hot tear, simply taking it as the evidence of a poor filling design. Thus the key action is to fix the filling design. The hot tear then disappears like magic. No action to reduce the strength of the mould or core is necessary.

This piece of practical experience appears to be disbelieved, or at least systematically overlooked, by every researcher in this field. This is a pity because there is much useful background that has been clarified by careful, systematic research over the years. This interesting background is reported here. It will become clear that it is consistent with the view that bifilms (as prior cracks suspended in the liquid) introduced by a poor running system are implicated as the fundamental cause, actually *becoming* the hot tear. Nevertheless, the patient reader will find it illuminating to review the experimental data, keeping in mind that bifilms are the major underlying cause.

The reader should also note the large number of terms used for hot tearing, such as hot cracking, hot shortness or hot brittleness. We shall use them interchangeably here.

Characteristics of a hot tear

The defect is easily recognised from one or more of several characteristics:

1. Its form is that of a ragged, branching crack, generally follow intergranular paths. This is particularly clear on a polished section viewed under the microscope (Figure 8.1).
2. The failure surface reveals a dendritic morphology (Figure 8.1).

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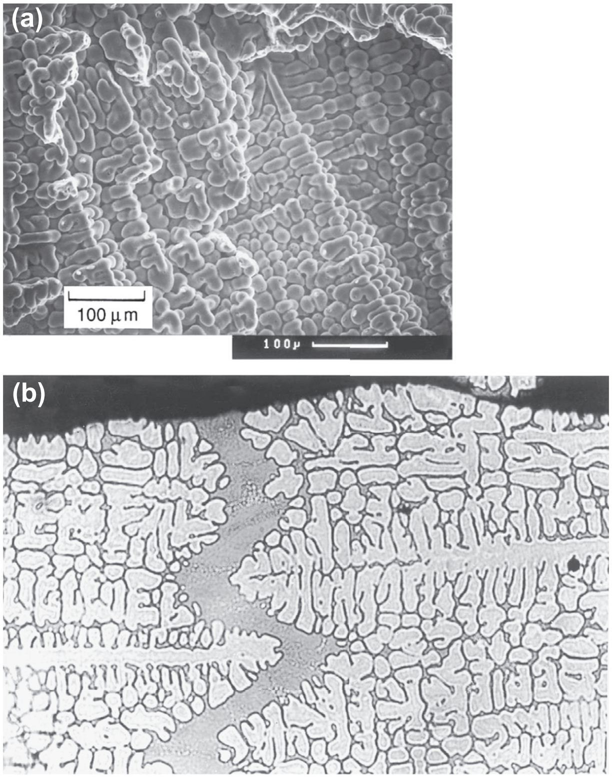

FIGURE 8.1

(a) Scanning electron microscope image of the surface of a hot tear in Al-7Si-0.4Mg alloy sand casting. (b) A filled hot tear in an Al-10Cu alloy (Spittle and Cushway, 1983).

3. The failure surface is often heavily oxidised (prior, of course, to any subsequent heat treatment). This is more particularly true of higher temperature alloys such as steels.
4. Its location is often at a hot spot where contraction strain from adjoining extensive thinner sections may be concentrated.
5. It does not always appear under apparently identical conditions; in fact, it seems subject to a considerable degree of randomness in relation to its appearance or non-appearance and to its extent.
6. The defect is highly specific to certain alloys. Other alloys are virtually free from this problem.

Before we go on to discuss the reasons for all this behaviour, it is worth bearing in mind the most simple and basic observation:

> The defect has the characteristics of a tear.

This disarmingly obvious characteristic immediately alerts us to a powerful clue about its nature and its origin. We can conclude that:

> A hot tear is almost certainly a uniaxial tensile failure in a weak material.

This may appear at first sight to be a trivial conclusion. However, it is fundamental. For instance, it allows us to make some important deductions immediately:

1. Those theories that assume hot tearing is the result of feeding difficulties can almost certainly be dismissed instantly. This is because feeding problems result in hydrostatic (i.e. a triaxial) stress in the residual liquid, causing pores or even layer porosity in the liquid phase. If the triaxial stress does increase to a level at which a defect nucleates, then the liquid separates and expands (triaxially) to create a pore among the dendrites. The dendrites themselves are not affected, and are not pulled apart. They continue to interlace and bridge the newly formed volume defect, as was discussed for layer porosity (Section 7.1.7.2).

This is in contrast to the hot tear, where it is clear from micrographs and radiographs (Figure 8.1(b)) that the dendrites open up a pathway *first*. The opened gap drains free of liquid *later*.

The pulling apart of the dendrites, separating to form a eutectic-rich path through the structure (Figure 8.1(b)) probably brings with it no significant problems of loss of strength or other properties of the casting despite its alarming and unwelcome appearance on radiographs or cut sections. However, if the path had subsequently drained of eutectic liquid, a hot tear would have been formed, which would, of course, have seriously affected strength. This formation of an open tear might have been avoided if feed metal had been locally available to keep the interdendritic regions full of residual liquid.

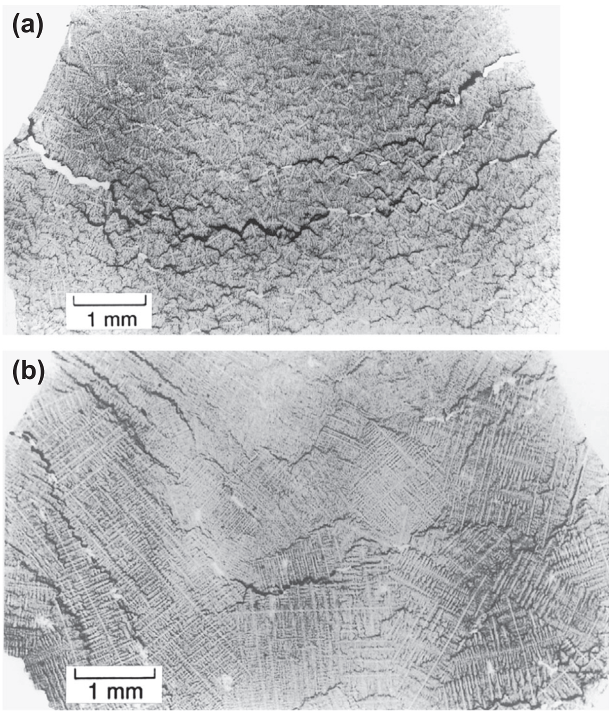

FIGURE 8.2

(a) Radiograph of a hot tear in an Al-6.6Cu grain-refined alloy. Dark regions are Cu-rich eutectic; white areas are open tears. (b) Radiograph of hot tears in Al-10Cu alloy not grain refined.

Rosenberg et al. (1960). Courtesy of Merton C Flemings.

A second confusion arises from the linking of hot tearing in general with the defect formed on the surface of the casting above a grossly underfed region. The severe local collapse of the casting surface above this hot spot will deform the surface considerably, concentrating much strain in this local region. Thus a hot tear may form over this badly fed volume. The experimental arrangement described by Paray et al. (2000) is of this type. This special type of hot tear can also be solved by improving the local feeding, even though the tear itself is the result of the intense local strain; only indirectly the result of feeding.

2. If the defect is a tear, then it has to be understood in terms of its initiation by exceeding a certain critical tensile stress, in common with all types of tensile failure. Rappaz et al. (1999) find a critical rate of tensile strain that will initiate a pore, finding that this gives a remarkably accurate estimate of the susceptibility to hot tearing. (The important work by Rappaz and his team will feature repeatedly in this review of the fundamentals of hot tearing.)

Keeping these general thoughts in mind will help us to keep the important features of the process in perspective while we deal with a host of other aspects. It is not surprising that the research literature on this subject is confusing; the subject is genuinely complicated.

However, before proceeding with the details of the mechanisms associated with this defect, it is probably necessary to dismiss two other contenders for explanations of these hot failures. Dickhaus (1993) has proposed that the effect of surface tension, in the phenomenon of viscous adhesion, might explain hot strengths and the failure mechanisms of solidifying metals. However, simple order-of-magnitude estimates indicate that surface tension is perhaps capable of generating only one hundredth, or even one thousandth, of the stresses involved in hot tearing. In a quite different approach, Fredriksson et al., (2005) has proposed that hot crack formation might occur because of the condensation of vacancies, expected to be present at concentrations of 0.01–0.1% in the metal lattice at these high temperatures. Vacancies are certainly present at maximum equilibrium concentrations at the melting point of metals and certainly condense during cooling. However, in practice, vacancies have never been observed to condense into volume defects such as cracks and pores. In every case examined, and in every metal studied so far, vacancies clusters collapse into either dislocation rings or stacking fault tetrahedra. These studies include direct observation in the electron microscope and computer simulations of the behaviour of large assemblies of atoms (known as molecular dynamics simulations). Basically, the forces between atoms are so great that pores or cracks cannot be opened up without the imposition of huge stresses in the GPa range. Solidifying metals are in general soft and ductile near their melting points, and so cannot generate and sustain such stresses (Campbell, 2010).

This leaves us clear to study the more likely contenders for the generation of these defects.

8.1.2 GRAIN BOUNDARY WETTING BY THE LIQUID

It was C.S. Smith who, during 1949–1952, first formulated the concept of the wettability of grain boundaries by the presence of a liquid phase in the boundary. Figure 8.3 summarises his concept. The shape of the grain boundary particles is largely controlled by the relative surface energies of the grain-to-grain interface itself, γ_{gg} and the grain-to-liquid interface γ_{gL}. The balance of forces is:

$$\gamma_{gg} = \gamma_{gL}\mathrm{Cos}\theta \tag{8.1}$$

It is clear that for most values of the equilibrium dihedral angle 2θ, the grain boundary liquid assumes compact shapes. However, it will, of course, occupy a greater area of the boundary as its volume fraction increases. The relation between (1) the area of the boundary which is occupied by liquid, (2) the dihedral angle and (3) the volume fraction of liquid present is a complicated geometrical calculation which the author was proud to identify and tackle (Campbell, 1971), setting off the subsequent improved treatment by Tucker and Hochgraf (1973), and, finally, the comprehensively solution by Wray (1976). Hochgraf (1976) went on later to develop a fascinating study of the conditions for the spread of the liquid phase under non-equilibrium conditions, where the dihedral angle becomes effectively less than zero.

The importance of the dihedral angle being zero for complete wetting is illustrated in the work of Fredriksson and Lehtinen (1977). They observed the growth of hot tears in Al-Sn alloys using the scanning electron microscope. The

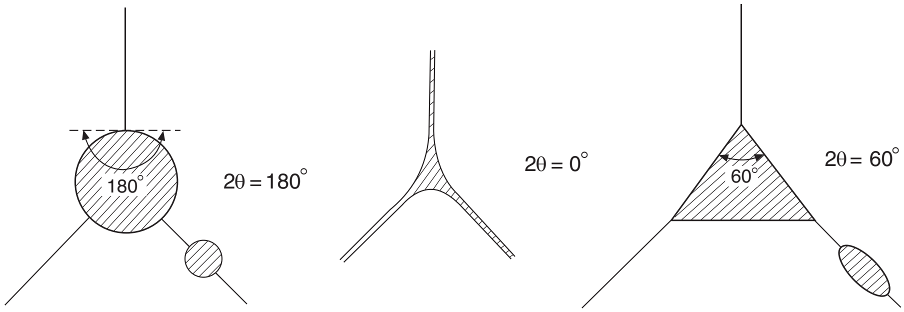

FIGURE 8.3

Shapes of the liquid phase at grain corners as a function of the dihedral angle (Smith, 1948, 1952).

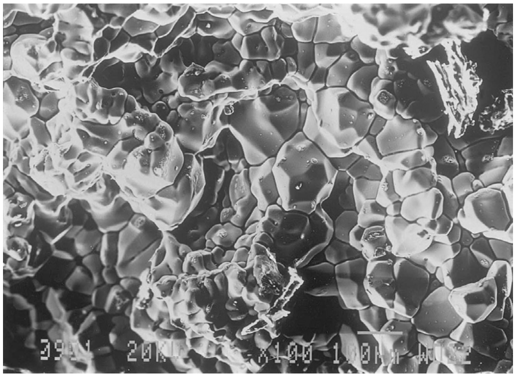

FIGURE 8.4

Hot tear surface of an Al-1%Sn alloy.

Courtesy Chakrabarti (2000).

liquid tin wetted the grain boundaries of the aluminium, leading to intergranular brittle failure when subjected to tension (Figure 8.4). In Al-Cd alloys, the liquid cadmium at the grain boundaries did not wet and therefore did not spread over the boundaries, but remained as compact pools, so that when subjected to tension these alloys failed by ductile fracture.

There have been several observations of failure by hot tearing where, on subsequent observation under the microscope, the fracture surface has been found to exhibit separate, nearly spherical droplets that appear to be non-wetting towards the fracture surface. This has been seen in systems as different as Al-Pb (Roth et al., 1980) and Fe-S (Brimacombe and Sorimachi, 1977; Davies and Shin, 1980). It seems certain that the liquid phase would originally wet a normal grain boundary. It is not clear therefore whether the observation is to be explained by the subsequent de-wetting of the liquid phase after the crack is exposed to the air, or, perhaps more probably, because the boundary consists of a remaining half of a poorly wetted oxide bifilm.

8.1.3 PRE-TEAR EXTENSION

Whilst the casting is cooling under conditions in which liquid and mass feeding continue to operate, if the casting is contracting the solid grains swim about, manoeuvring into new positions. In these conditions clearly no tearing can occur.

The problem starts when grains grow to the point at which they finally collide firmly against each other, but are still largely surrounded by residual liquid.

Patterson et al. (1967) were among the first to consider a simple geometrical model of cubes. We shall develop this concept further as illustrated in Figure 8.5. It is clear that for grains of average diameter a separated at first by a liquid film of thickness b, the pre-tear extension ε can reach approximately:

$$\varepsilon = b/a \tag{8.2}$$

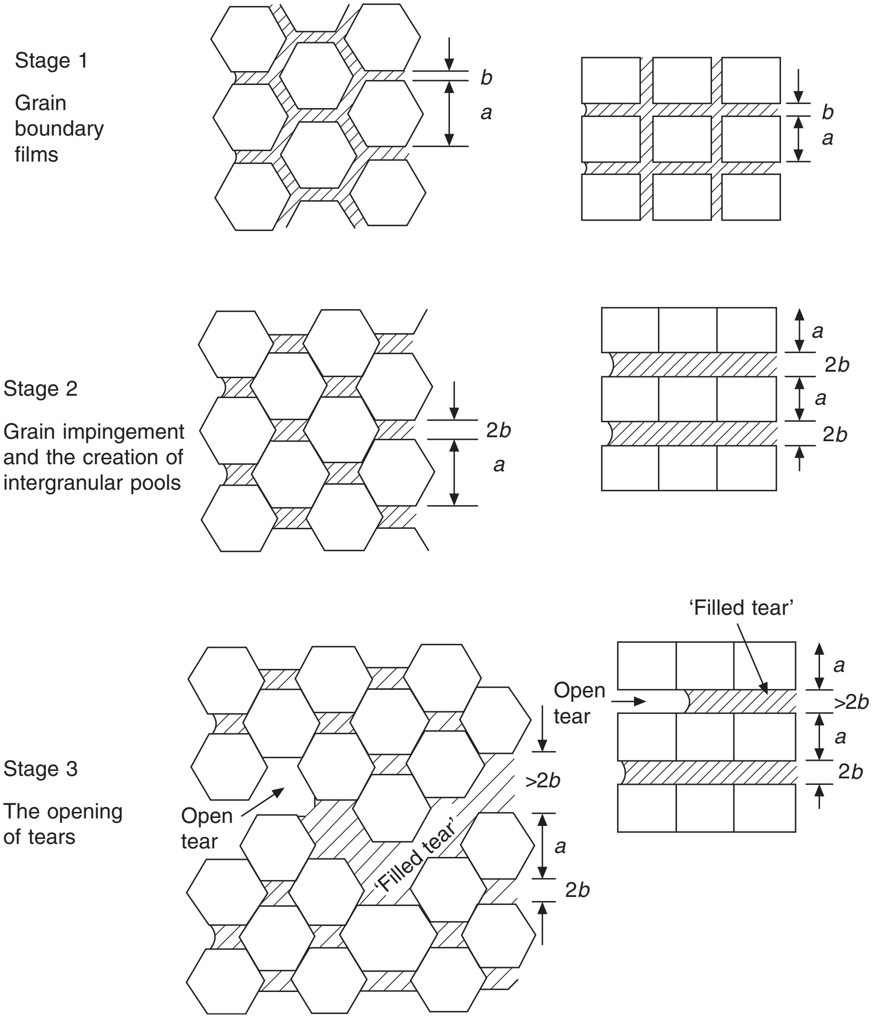

FIGURE 8.5

Stages of hot tearing using two-dimensional square and hexagonal grain models. Grains size 'a' is surrounded by a liquid film of thickness 'b'. Stage 1: unstrained structure; stage 2 shows isolated regions of segregate; stage 3 shows open tears plus so-called 'filled' hot tears. Continued strain eventually will drain the liquid film completely, completing the tear.

For both the cube and the hexagon models in two dimensions, the relation $b/a = f_L/2$ can quickly be seen to be true, where f_L is the volume fraction of liquid. For a three-dimensional cube model, the reader can easily confirm the further relation:

$$b/a = f_L/3 \tag{8.3}$$

Thus for between 3 and 6% residual liquid phase, we have between 1 and 2% extension before the impingement of the grains. The pre-tear extension being proportional to the amount of liquid present is an observation confirmed many times by experiment. Furthermore, those alloys with large amounts of eutectic liquid during freezing, such as the Al-7Si and higher Si alloys, are usually free from hot tear problems probably for this reason; there is plenty of extension that can be accommodated before any danger of the initiation of a tear.

Also, for a given amount of liquid present, the extension is inversely proportional to the grain size. Thus for finer grains, more strain can be accommodated by easy slip along the lubricated boundaries without the danger of cracking.

After the grains have impinged, a certain amount of grain boundary sliding may continue, as we shall discuss later, although this later phase may contribute only a very limited amount of further extension.

Even in the case of the solidification of pure metals, the grain boundaries are known to have a freezing point well below that of the bulk crystalline material (see, for example, Ho et al. (1985) and Stoltze et al. (1988)). The presence of liquid at the grain boundaries even in pure metals, but perhaps only a few atoms thick, may help to explain why some workers have found tearing behaviour at temperatures apparently below the solidus temperature. However, many of the observations are also explainable simply by the presence of minute traces of impurities that have segregated to the grain boundaries. The two effects are clearly additive.

Because of the presence of the grain boundary film of liquid, bulk deformation of the solid will occur preferentially in the grain boundaries, so long as the strain is below a critical value (Burton and Greenwood, 1970). This explains why the extension of the solid during fracture can be accounted for completely by the sum of the effects of (1) grain boundary sliding plus (2) the extension resulting from the opening up of cracks (Williams and Singer, 1968).

Later, during grain boundary sliding where the grains are now in contact over their complete surfaces, there has to be some deformation of the grains themselves. Novikov et al., (1966) found by careful X-ray investigation that the deformation is confined to the surface of the sliding grains. In addition, at a temperature close to the melting point, recovery of the grains is so fast that they do not work harden. Because they remain in a relatively soft condition, the general flow of the bulk material can continue relatively easily. Thus although the flow is now actually controlled by bulk deformation of the grains, the appearance under the optical microscope is simply that of the sliding of the grains along their boundaries.

It is necessary to keep in mind that the total extension resulting from the various kinds of grain boundary sliding (whether 'lubricated' or not) amount to only perhaps 1–2% strain. Further strain of at least this magnitude arises during the extension of the crack itself, as is discussed later.

8.1.4 STRAIN CONCENTRATION

It was Pellini in 1952 that drew attention to the concentration of strain that could occur at a hot spot in a casting. It is instructive to quantify Pellini's theory by the following simple steps.

If the length of the casting is L, and if it has a coefficient of thermal expansion α, during its cooling by ΔT from the liquidus temperature it will contract by an amount $\alpha\Delta TL$. If all this contraction is concentrated in a hot spot of length l, then the strain in the hot spot is given by:

$$\varepsilon = \alpha\Delta TL/l \tag{8.4}$$

Clearly, in the hot spot, the casting contraction strain is increased by the factor L/l.

For a casting 300 mm long and a hot spot of approximately 30 mm length at its end, the strain in the hot spot is concentrated 10 times. This would be expected to be a fairly typical result—although it seems possible that strain concentrations of up to 100 or more may sometimes occur.

It is interesting to note that the problem in the hot spot depends on the amount of strain concentrated in it, and this depends on the size of the adjacent casting and the temperature to which it has cooled while the hot spot remains hot and in a weak state.

We can clarify the size of the problem by evaluating an example of an aluminium casting. Assume that $\alpha = 20 \times 10^{-6}\ \mathrm{C}^{-1}$ and that the casting has cooled 100°C. If its contraction is hindered, the strain that will result is, of course, $20 \times 10^{-6} \times 100 = 0.002 = 0.2\%$. This level of strain puts the material as a whole above the elastic limit even at room temperature. (In materials that do not show clear yield points, the yield stress is often approximated to the so-called proof stress, at which 0.1–0.2% permanent strain remains after unloading.) In the hot spot, therefore, if the strain concentration factor lies between 10 and 100, then the strain will be between 2 and 20%. These are strains giving an amount of permanent plastic extension that is relatively easily withstood by sound material. However, in material that is weakened by the presence of bifilms at the grain boundaries, and which can withstand typically only 1–2% of strain before the start of failure, as we shall see later, it is no wonder that the casting fails.

In addition to the consideration of the amount of strain concentrated into the hot spot, it is also necessary to consider how many grain boundaries the hot spot will contain. If the grain size is coarse, the hot spot may contain only one boundary, with almost certain disastrous consequences, because all the strain will be concentrated in that one liquid film. If the hot spot contains fine grains, and thus many boundaries, then the strain per boundary is reduced. We may quantify this because the number of grains in the length l of the hot spot is l/a for grains of diameter a. Hence if we divide the strain in the hot spot (Eqn (8.4)) by the number of boundaries in it, then we have the strain per boundary ε_b:

$$\varepsilon_b = \alpha \Delta T L a / l^2 \tag{8.5}$$

It is clear that to reduce the strain that is trying to open up the individual grain boundaries, the beneficial factors include (1) reduced temperature differences, (2) smaller overall lengths between hot spots and (3) finer grain size. However, Eqn (8.5) reveals for the first time that the most sensitive parameter is the length l of the hot spot; if this is halved, the grain boundary strain is increased four times.

8.1.5 STRESS CONCENTRATION

The problem of how sufficient stress arises during cooling to initiate and grow the hot tear may not be relevant because the forces available during cooling are massive, greatly exceeding what is necessary to create a failure in the rather weak casting. Thus we may consider the forces available as being irresistible, forcing the casting to deform. Because this deformation will always occur, the question as to whether a hot tear will arise is clearly *not* controlled by stress, but must depend on other factors, as we shall discuss in this section.

Nevertheless, although overwhelmed, the forces of resistance offered by the casting are not quite negligible. Guven and Hunt (1988) have measured the stress in solidifying Al-Cu alloys. Although the stresses are small, they are real, and show a release of stress each time a crack forms. The loads at which failure occur are approximately 50 N in a section 20×20 mm. Thus the stresses are approximately 0.1 MPa (compared with a strength of $>$100 MPa at room temperature). Also, as an interesting detail, a simultaneous change in the rate of heat transfer across the casting-mould interface was detected each time the force holding the casting against the mould was relaxed.

In rough agreement with Guven and Hunts' results, Forest and Berovici (1980) carried out careful tensile tests and found that an Al-4.2Cu alloy has a strength of more than 200 MPa at 20°C, that falls to 12 MPa at 500°C, 2 MPa at the solidus temperature, and finally to zero at a liquid fraction of about 20%.

As we have mentioned before, the other stress that may be present could be a hydrostatic tensile stress in the liquid phase. Although this may contribute to the nucleation of a pore, which in turn might assist the nucleation of a tear, the presence of a *hydrostatic* stress is clearly not a necessary condition for the formation of a tear, as we have discussed earlier. We need a *uniaxial* tensile stress to create a tear.

One final point should be emphasised about stresses at these high temperatures. Because of the creep of the solid at high temperature, any stress will depend on the rate of strain. The faster the solid is strained, the higher will be the stress with which it resists the deformation.

Zhao et al. (2000) have determined the rheological behaviour of Al-4.5Cu alloy, and thereby have determined the stress leading to the critical strain at which hot tearing will cause failure. This novel approach may require the densities of bifilms to be checked for similarity between their rheological sample and their hot tearing test piece, which is clearly poured rather badly. The elegant test piece devised by Rappaz and his team (Mathier, 2009) has no defined filling technique, and one can only assume that the filling is poor, manufacturing entrainment defects that will confuse the results. It would be valuable to redefine this attractive-looking test to ensure that it could be filled without compromising the quality of the metal being tested.

8.1.6 TEAR INITIATION

Probably the most important insight into the problem of tear initiation was provided by Hunt (1980) and Durrans (1981). Until this time, the nucleation of a tear was not widely appreciated as a problem. The tear was assumed simply to form! The experiment by these authors is an education in profound insight provided by a simple technique.

These researchers constructed a transparent cell on a microscope slide that enabled them to study the solidification of a transparent analogue of a metal. The cell was shaped to provide a sharp corner around which the solidifying material could be stretched by the turning of a screw. The idea was to watch the formation of the hot tear at the sharp corner.

The amazing outcome of this study was that no matter how much the solidifying material was stretched against the corner it was not possible to start a hot tear in clean material: the freezing mixture continued to stretch indefinitely, the dendrites continuing simply to move about and rearrange themselves.

However, on the rare occasion of the arrival of a small inclusion or bubble near the corner, then a tear opened up immediately, spreading between the dendrites and away from the corner. In their system, therefore, hot tearing was demonstrated to be a process dependent on nucleation. In the absence of a nucleus, a hot tear did not occur no matter how much strain was applied. This fact immediately explains much of the scattered nature of the results of hot-tearing work in castings: apparently identical experimental conditions do not give identical tears, or at times even any tears at all.

It is necessary to remember that in the Hunt and Durrans' work, the liquid would wet the mould, adhering to the sharp corner, and so would require a volume defect to be nucleated; the defect would not be created easily. In the case of a casting, however, a sharp re-entrant corner may have liquid present at the casting surface, but the liquid will not be expected to wet the mould. In fact there is the complication that the liquid will be retained inside the surface oxide of the casting. The drawing of liquid away from this surface, analogous to the case of surface-initiated porosity, would represent the growth of the crack from the surface, and may involve a nucleation difficulty. Thus in the case of cracks initiated at the surface, the Hunt and Durrans observation may not apply universally. However, the concept is still of value, as we shall discuss next. In the case of internal hot tears, their observation remains crucial. However, in the case of alloys that form strong oxide skins, there may still be a difficulty in drawing inwards a rather rigid surface film, or of nucleating a tear on the underside of the surface oxide film; either way, surface initiation might be difficult in many casting alloys.

However, even at the surface of a casting in an alloy that does not form a surface film, the initiation of a tear may not be straightforward. It is likely that the tear will only be able to start at grain boundaries, not within grains, because the dendrites composing the grain itself will be interconnected, all having grown from a single nucleation site. Dendrites from neighbouring grains will, however, have no such links, and in fact the growing together and touching of dendrite arms has not been observed in studies of the freezing of transparent models. The arms are seen to approach, but final contact seems to be prevented by the flow of residual liquid through the gap. Thus if a grain boundary is not sited conveniently at a hot spot, and where strain is concentrated, then a tear may be difficult to start. This will be more common in large-grained equiaxed castings, as suggested by Warrington and McCartney (1989).

If a grain boundary is favourably sited, it may open along its length. However, on meeting the next grain, which in general will have a different orientation, further progress may be arrested, at least temporarily. Thus a tear may be limited to the depth of a single grain. The effect can be visualised as the first stage of the spread of the tear in the hexagon grain model shown in Figure 8.5. Considerably more strain will be required to assist the tear to overcome the sticking point in its further advance beyond the first grain.

For the case of columnar grains, the boundaries at right angles to the tensile stress direction will provide conditions for easy initiation of a tear along such favourably oriented grain boundaries. The effect is analogous to the rectangular grain model in Figure 8.5.

For fine-grained equiaxed material where the grain diameter can be as small as 0.1–0.2 mm, the dispersion of the problem as a large number of fine tears, all one grain deep, is effectively to say that the problem has been solved. This is because the crack depth would then be only approximately 0.1 mm. This is commensurate with the scale of surface roughness because average foundry sands also have a grain size in the range 0.1–0.2 mm. The fine-scale cracking would have effectively disappeared into the surface roughness of the casting.

Nevertheless, it is fair to emphasise that the problem of the nucleation of tears has been very much overlooked in most previous studies. Nucleation difficulties would help to explain much of the apparent scatter in the experimental observations. A chance positioning of a suitable grain boundary containing, by chance, a suitable nucleus, such as a folded oxide film, would allow a tear to open easily. Its chance absence from the hot spot would allow the casting to freeze without defect; the hot spot would simply deform, elongating to accommodate the imposed strain.

In practice, there is much evidence to support the assertion that most hot tears initiate from entrained bifilms. As has been mentioned previously, the author has personally solved every hot tearing problem he has encountered in foundries by simply improving the design of the casting filling system. The proposal to use such an approach has generated almost universal scorn and disbelief. However, when good filling system designs were implemented systematically in an aerospace foundry, all hot tearing problems in the difficult Al-4.5Cu-0.7Ag (A201) alloy disappeared, to be replaced by surface sink problems. In comparison to the hot tears, the surface sinks were welcomed and easily dealt with by improved feeding techniques (Tiryakioglu, 2001).

The study by Chadwick and Campbell (1997) of A201 alloy poured by hand into a ring mould containing a central steel core showed that failure by hot tearing in such a constrained mould was almost guaranteed. Conversely, when the metal was passed through a filter, and caused to enter the mould uphill at a speed of less than 0.5 ms^{-1} to ensure the avoidance of defects, no rings exhibited hot tears. This was an amazing result, showing no failures in one of the world's most hot tear prone alloys, in a test designed to maximise hot tearing conditions. (Memorably, when Chadwick arrived in my office to report the results of the hot tearing test, I was amazed, shocked into silence to see the cast rings so tightly contracted onto the steel cores that they could not be extracted, but completely without tears. While I was gathering my thoughts to comment appropriately on this immensely important and exciting result, Chadwick leaned back in his chair. "Yes" he said, "The test was a complete failure." I was aghast at this remark and could only manage to get out the word "What!" He went on with resignation "What was the use of a hot tear test that did not hot tear?")

A similar study was repeated for Al-1%Sn alloys by Chakrabarti and Campbell (2000). Surfaces of hot torn alloys illustrate the brittle nature of the failure in this alloy (Figure 8.4). The alloy has such a large freezing range, close to 430°C (extending from close to pure Al at 660°C down to nearly pure Sn at 232°C) that in the hot tear ring test the alloy appears even more susceptible to failure by hot tearing than A201 alloy. When subjected to the uphill filled version of the ring test, most castings continued to fail. However, about 10% of the castings solidified without cracks. Once again, the existence of even one sound casting would be nothing short of amazing.

Further evidence can be cited from the work of Sadayappan et al. (2001), who demonstrated that their as-melted Al alloy gave many and large hot tears, whereas after cleaning the metal by degassing they observed only a few small tears. Dion et al. (1995) found that in their castings of yellow brass, the addition of aluminium to the alloy promoted hot tearing, as would be expected from the presence of the entrained alumina film resulting from their very turbulent filling system.

8.1.7 TEAR GROWTH

We have touched on the problem of tear growth in the previous section. However, it bears some repeating that (1) the birth of the hot tear and (2) its growth, sometimes to awesome maturity, are quite separate phenomena.

The evidence is growing that tears are closely associated with bifilms. It remains to clarify the nature of the link. For instance, (1) do tears initiate on bifilms and subsequently extend into the matrix alloy? Or (2) do bifilms constitute the tears, so that the growth of the tear is merely the opening of the bifilm, so that the defect is revealed, in fact becoming obvious. The evidence is accumulating that the important mechanism is (2).

The easy growth in columnar grains where the direction of tensile stress is at right angles to the grain boundaries has been mentioned. Spittle and Cushway (1983) observed that the linear boundary formed between columnar crystals growing together from two different directions was an especially easy growth route for a spreading crack. This is confirmed by experience in the rolling industry, where the diagonal plane issuing from the corners of rectangular ingots, defining the joint plane of the two sets of columnar grains from the two adjacent sides, is a common failure plane during the early reduction passes. The problem is reduced by rounding the corners, or reducing the levels of critical impurities. In steel ingots the significant impurities are usually sulphur, and the so-called tramp elements such as lead and tin.

The explanation in terms of bifilms is that columnar grains both align the bifilm cracks along their length, but also push other bifilms ahead into intergranular spaces, so that failure along these surfaces is to be expected. The presence of an invisibly thin film organised into position between dendrite arms explains the fracture surface seen in Figure 8.1(a). The fracture follows the bifilm, the fracture surface exhibiting steps at integral numbers of dendrite arms, as explained in Figure 2.43. Naturally, the hot tear morphology is also seen in the room temperature fracture seen in Figures 2.44 and 6.29. Clearly, after being extended and flattened by the growth of dendrites, the bifilm is simply a hot tear waiting to be opened and so be revealed. If is not opened during solidification, then it can wait until opened later in a tensile test, or, more worryingly, lying in wait to open later still, causing a failure in service.

A radiograph of an Al alloy (Figure 2.43(c)) by Fox and Campbell J (2000) in which the bifilms have been opened up by the action of reduced pressure show the bifilms near to the mould surface to be organised at right angles to the surface by the pushing action of the growing columnar grains.

In general, the bifilms and the consequential porosity are sited at grain boundaries, leading to intergranular failures, but as we have noted previously, bifilms can be incorporated into growing grains, leading to transgranular failures. In all cases, however, the individual dendrites that constitute a growing grain never cross a bifilm. They cannot grow through air.

Warrington and McCartney (1989) confirm the findings of Spittle and Cushway (1983) when they find that fine equiaxed grains also promote easy growth conditions for a hot tear. This seems to be because the tear can propagate intergranularly along a path that, because of the fine grain size, can remain almost perpendicular to the applied stress on a macroscopic scale. In terms of bifilms, the result is simply the observation of the opening of that particular bifilm favourably oriented with respect to the stress direction, out of the many bifilms usually present between small grains.

Conversely, coarse equiaxed grains gave increased resistance to the spread of the crack. In this case the bifilms would be segregated to planes sometimes well away from the stress direction, causing greater plastic deformation in the attempted return of the crack to its average growth direction. In addition, of course, the distance travelled by the crack would be significantly increased.

The question of the amount of plastic work that is expended during the propagation of a tear is interesting. The work of deformation is easily shown to be of the order of at least 10^4 times greater than the work required to create the newly formed surfaces of the tear. Thus arguments based on the effect of the surface energy of the crack limiting its growth (as in the case of a classical Griffiths crack in a brittle solid such as glass) are clearly not relevant in the case of the failure of plastic solids such as metals at their melting points.

Novikov and Portnoi (1966) draw attention to the fact that, despite the rather brittle appearance of the fracture, the hot torn surfaces cannot usually be fitted back together again, confirming the expectation that considerable plastic deformation occurs during hot tearing. Furthermore, they found that the gap between the poorly fitting surfaces corresponded almost to the total elongation, indicating that the elongation was associated almost entirely with crack propagation in their work. The further implication was that (1) nucleation of the crack was easy because high stresses would have given high elongations before cracking and (2) the pre-tear extension resulting mainly to grain boundary sliding was limited in their experiments. These observations are again consistent with propagation along a bifilm. The brittle, intergranular appearance of the fracture surface is typical of a crack that has followed the central unbonded interface of a double oxide film.

Although high stresses cannot be envisaged at the stage of the nucleation of a crack (because crack nucleation will find easy start sites such as the opening up of bifilms), when the crack has started to grow, the sharpness of its tip ensures that high stress is available locally at this point. For large-grained material, therefore, the occasional absence of a favourably oriented grain boundary can be expected to result in further propagation by two means:

1. The continuation of the crack across the grain that attempts to block its path. Such occasional transgranular growth has been observed by Davies et al., (1970) in models using low-melting-point Sn-Pb alloys. He saw that in such cases the crack followed the cell boundaries, which were clearly the next best route for the crack. Bifilms would be expected in both grain and cell interfaces of course, and present a similarly cogent explanation.
2. The crack either renucleates a short distance ahead in a favourably oriented grain boundary, or more likely, travels around the grain by a path out of the plane of the section, appearing once again a little ahead. Fredriksson and Lehtinen (1977) have directly observed such behaviour by pulling specimens of Al-Sn alloys in the scanning electron microscope. As the crack continues to open, the intervening region of obstructing grain is seen to deform plastically, like a collapsing barrier. The stepwise propagation of the crack, linked by plastic barriers at the steps, explains the irregular, branching appearance that is a characteristic feature of hot tears. The failure of plastic bridges between randomly sited, disconnected bifilms explains the observations similarly.

Returning to our geometrical model, at the end of the pre-tear extension period, the residual liquid is separated into pools between grains, as is seen in stage 2 of Figure 8.5. The corresponding pools in the real casting are seen in Figure 8.2. These compact segregates must pose a problem for subsequent solution heat treatments, even though their existence does not seem to have been previously recognised. At this point, further strain must cause some concentration of failure at a weak point in the structure of the solidifying casting. If a grain near the surface is separated from its neighbour by the presence of a bifilm, the growth of the crack can then occur with little further hindrance. The necessarily irregular nature of its progress mirrors that of the real crack; once again, compare Figures 8.2 and 8.5. Straighter portions of the crack resemble the cube model in Figure 8.5.

A potentially important feature of the hot tearing literature needs to be raised. From Figures 8.1, 8.2 and 8.5, it is quite clear that the portions of the cast structure in which the dendrites have separated but which still contain residual liquid have *always* contained residual liquid. This may seem self-evident. However, the casting literature is full of references to 'healed' hot tears, meaning tears containing residual liquid, but implying that the tears were once empty, and, fortuitously, were somehow subsequently filled by an inflow of liquid. Whether the term 'healed hot tears' is really intended to imply original emptiness and subsequent refilling is not clear, but the term is misleading and would be better discontinued. The term 'filled tear' is more explicit. 'Un-emptied' tear would be even more accurate, but hardly an attractive name! The *filled tear* is simply the region between grains that have been separated by uniaxial strain, but which still remain full of intergranular liquid. If it solidifies while still full, as in Figure 8.1(b), it will constitute a region of segregate, but will almost certainly still be as strong as the bulk of the casting, and so not constituting a defect that might impair its serviceability. It is *not* a hot tear and has *never* been a hot tear.

Only if it becomes empty does it become a major defect meriting the name of hot tear. There are two quite separate mechanisms that could provide an empty tear.

1. The separation of the grains by continued strain to the point at which the residual liquid is no longer capable of keeping the tear filled. This is the mechanism displayed in our simple model (Figure 8.5), and as seems to be shown in Figure 8.2(a). The available liquid has been insufficient to keep the intergranular regions filled, and so, with increasing demand from surrounding liquid regions as these regions are pulled apart, has simply sucked the main crack region dry.
2. The previous mechanism contrasts with the hot tear whose conditions for formation are identical (i.e. grains are separated by contraction strain) but the region between the grains now contains one or more bifilms that, on separation of the grains, also separates the halves of the bifilms. This creates a crack instantly and easily.

Both of these mechanisms seem possible. Both seem likely to operate. However, the first is expected to be more difficult to grow because the grains deform plastically as the crack attempts to propagate from grain to grain. The presence of the bifilm in the second mechanism is expected to create a defect of a serious size with ease.

8.1.8 PREDICTION OF HOT TEARING SUSCEPTIBILITY

Over the years, there have been many attempts to provide a useful working theory of hot tearing. Recently the attempts have narrowed to a few serious contenders. The exercise that has been found to be most useful to discriminate between them has been the attempt to predict susceptibility to hot tearing as a function of composition for binary alloys.

This is a useful test in alloy systems that display a eutectic. At zero solute content the theory has to contend with a pure metal; at low solute contents, only solid solution dendrites are present; above a critical solute content eutectic liquid appears for the first time, steadily increasing to towards 100% as the solute content increases towards the eutectic composition. The ability to deal with all of these aspects across a single alloy system constitutes a searching test of any theory, and covers the majority of solidification conditions in real castings.

A typical experimental result is shown in Figure 8.6. It reveals a steeply peaked curve that Feurer (1976) has called a lambda curve, after the shape of the Greek capital letter Λ. The problem is to find a theoretical description that will allow the lambda curves to be simulated for different alloy systems.

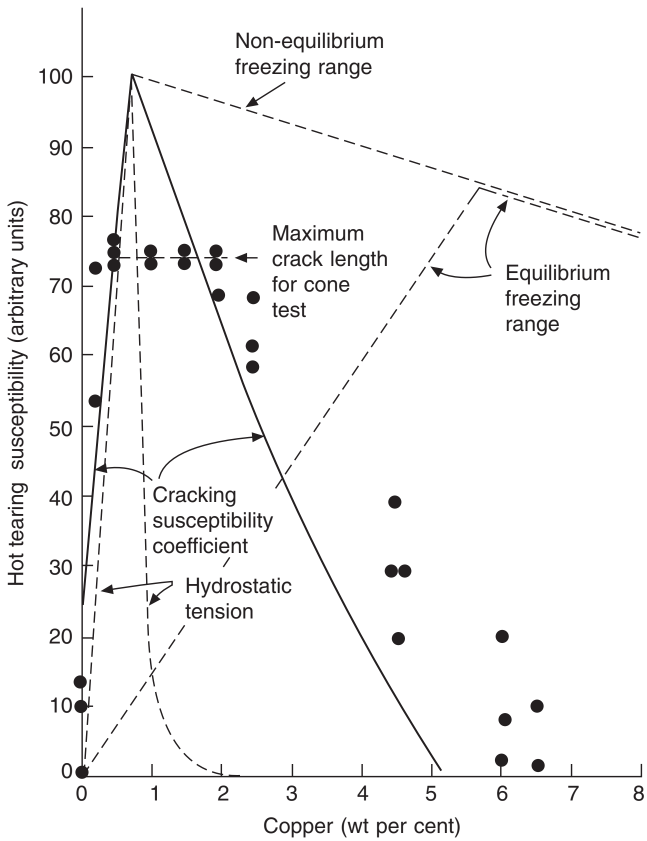

FIGURE 8.6

Hot tearing response of Al-Cu alloys showing a peak (necessarily extrapolated somewhat) at approximately 0.7Cu using the conical ring die test by Warring ton and McCartney (1989) compared with various theoretical models. Freezing ranges and hydrostatic tension by JC (1989); CSC by Clyne and Davies (1977).

It is salutary to note that for the Al-Cu alloy system any prediction based on the equilibrium diagram is completely wrong. Here the maximum freezing range would be predicted to be at 5.7Cu, which might lead the unwary to believe that the maximum problem in porosity and hot tearing should be at this copper content. From Figure 8.6 the problem in hot tearing is clearly centred on a rather dilute alloy of approximately only 0.5Cu. Any problems of hot shortness have disappeared on reaching 5.7Cu!

It is interesting to look at my early effort at prediction (Campbell, 1969a,b) that deals with the analogous problem of porosity in a spectrum of binary alloy compositions. Here the relative hydrostatic tension developed by the flow of feed metal through the dendrite mesh was calculated. The form of the relationship calculated for the Al-Cu system is shown in the figure. This particular result is based on the assumption that the residual liquid is 1% by volume. The peak is almost exactly in the correctly predicted location, confirming the fundamental importance of the arrival of eutectic liquid at that critical concentration of solute. The result closely agrees with the model by Rappaz et al., (1999), as would be expected because both models are based on the development of hydrostatic tension at the root of the dendrites.

The remainder of the predictions based on hydrostatic tension follows this particular experimental data by Warrington and McCartney (1989) poorly, but follows the data by Spittle and Cushway (1983) rather more closely. This intermediate agreement reflects the general capability of the models to achieve fair agreement with experimental data that are in themselves of rather variable quality.

It is surprising that models based on hydrostatic tension agree as well as they do with hot-tear initiation because they are not expected to be necessarily closely related, as we have mentioned before. The hydrostatic tension falls steeply with the arrival of eutectic liquid, dramatically reducing shrinkage porosity as seen in the original work, but not reducing hot tearing so steeply as found experimentally by Warrington and McCartney.

We also need to note that hot tearing is closely related to the peak of the non-equilibrium freezing range, but bears no relation to the subsequent slope of the non-equilibrium freezing range curve, as is clear in Figure 8.6.

The theoretical approach by Feurer (1976) that appears to explain the form of the lambda curves can be similarly discounted because it is also based on the modelling of liquid flow and hence the development of hydrostatic stress, not uniaxial tension (Campbell and Clyne, 1991).

An alternative theoretical approach to hot tearing was proposed by Clyne and Davies (1979). They implicitly assume that the failure is the result of uniaxial tension, but point out that strain applied during the stage of liquid and mass feeding is accommodated without problem by the casting. The problem of accommodating strain only occurs during the last stage of freezing, when the grains are no longer free to move easily. They define a cracking-susceptibility coefficient (CSC):

$$\mathrm{CSC} = t_V/t_R \tag{8.6}$$

Where t_R is the time available for stress-relaxation processes such as liquid and mass flow and t_V is the vulnerable period when cracks can propagate between grains. The concept is clear, but defining the limits of applicability of these various regimes for different alloy systems is not easy. However, as a first attempt the authors assume that the stress-relaxation period spans a fraction liquid f_L of approximately 0.6–0.1, and the vulnerable period spans f_V values 0.1–0.01 (Figure 8.7). The predictions of the scheme for the Al-Cu system are shown in Figure 8.6 to fit the Warrington and McCarthy data better. For the Al-Si system, they correctly predict a lambda curve with the correct form and peak at approximately 0.3Si, as found by their experiments. For the Al-Mg and Al-Zn systems, they found the agreement less good. The agreement for Al-Mg was improved later by Katgerman (1982), who modified the CSC limits. For magnesium-based alloys Clyne and Davies (1981) use their model to predict a peak at 2.0Zn for the Mg-Zn system and 3.0Al for the Mg-Al system. The poorer tearing resistance of the zinc-containing alloy is the result of its considerably greater freezing range. However, from the ring test results shown in Figure 8.8, the peaks are actually observed to be at approximately 1.0Zn and 1.0Al. It is necessary to keep in mind that the experimental data will be significantly affected by the presence of bifilms, which will, in general, reduce the composition of the peak susceptibility, bringing agreement between theory and experiment a little closer in line.

The approach therefore seems basically sound and useful, even though not always especially accurate in its predictions, as we have seen. Thus, in common with most theories, it invites further development and refinement.

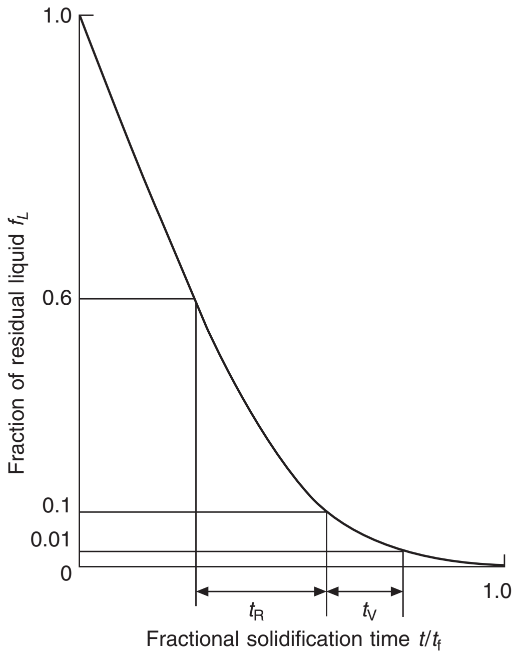

FIGURE 8.7

Model by Clyne and Davies (1977) for the regimes during which either stress relaxation or vulnerability to hot tearing occur.

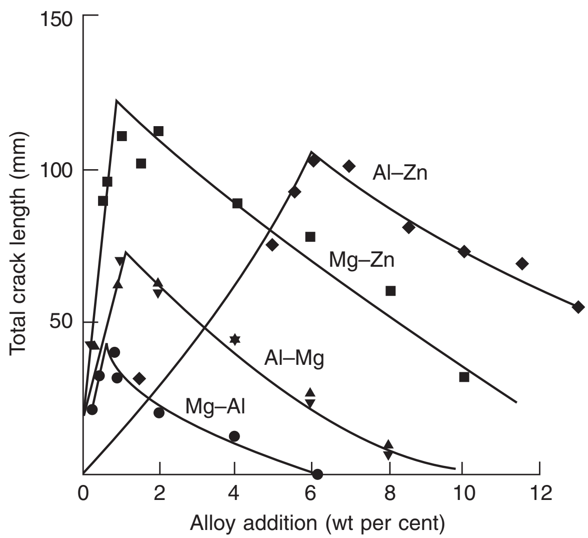

FIGURE 8.8

Hot tearing behaviour of various alloys subjected to the ring die test.

Data from Dodd (1955), Dodd et al. (1957), Pumphrey and Lyons (1948) and Pumphrey and Moore (1949).

As a start it would seem useful to combine the CSC with Eqn (8.5) derived previously for the strain per boundary in the hot spot. This gives a modified CSC as:

$$CSC_b = \frac{\alpha \Delta T L a}{l^2} \cdot \frac{t_V}{t_R} \tag{8.7}$$

There is already considerable evidence to suggest that this more general equation is at least approximately accurate. Figure 8.9 shows how grain size is significant. Also, several researchers with different diameter ring tests confirm that the cracking susceptibility is proportional to the circumference of the ring (Isobe et al., 1978). This proportionality to length is implicit in the design of the various I-beam tests using graded lengths of beam (see later). Pekguleryuz (2010) compared several CSC parameters for Al-Si alloys and found Eqn (8.7) to give the best correlation.

The CSC model has been extended to the cracking of steels (Clyne et al., 1982). In Fe-S alloys, the prediction of a CSC peak at 0.1%S was observed to be accurately fulfilled in experiments by Davies and Shin (1980). Here, as a consequence of the complexity of iron-based alloys, the CSC model is extremely useful in providing a framework to understand the phenomena. For instance, Rogberg (1980) found that for stainless steels that solidified to delta-iron the alloys were insensitive to the impurities As, Bi, Pb Sn, P and Cu, whereas those that solidified to gamma iron suffered serious loss of hot ductility. Kujanpaa and Moisio (1980) confirmed that S and P embrittled gamma iron, but not delta ferrite, but the best resistance to embrittlement was provided by a mixture of gamma and delta irons.

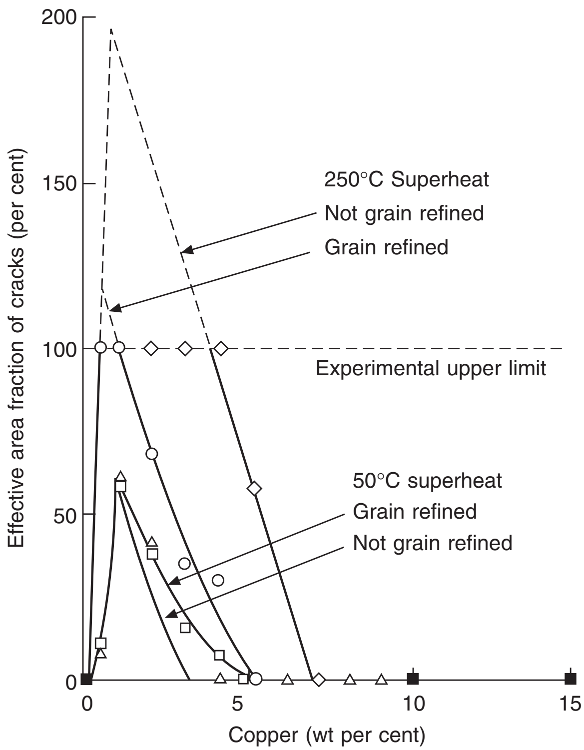

FIGURE 8.9

Hot tearing behaviour of Al-Cu alloys using an I-beam-type test showing the benefits of low casting temperature and grain refinement (Spittle and Cushway, 1983). Peaks are extrapolated to illustrate the close agreement with Figure 8.6.

The most sophisticated development of attempts to predict the susceptibility to hot tearing has been the series of developments since 1999 by Rappaz and colleagues. Initially, their model assessed the linear contraction and volumetric shrinkage contributions to the hydrostatic tension, admitting that the model at that stage only predicted the formation of the initial pore, and did not extend to the development of a crack. A series of publications has followed since 2006 expanding the concept of a granular model, but one in which the grains were randomly distributed. This has been a great advance on the simplistic geometrical models discussed earlier, allowing the details of the nature of the pasty zone to be simulated during the progressive stages of freezing. The network of channels of residual liquid are seen in Figure 8.10. A result for an Al-1Cu alloy pasty zone is illustrated in Figure 8.11, and the highly non-uniform flow through the zone is shown in Figure 8.12. The authors summarise their work in the creation of morphological maps (Figure 8.13). Although at this stage these simulations are only two-dimensional, an extension to three-dimensional is claimed to be relatively easy. It will represent a major step forward in our understanding of the processes within the pasty zone. Even so, the presence of bifilms will complicate these already complex flow patterns still further. It is not surprising that the pasty zone sometimes will allow no flow whatever when jammed with bifilms, which appears to be the only explanation of the results shown in Figure 6.18.

8.1.9 METHODS OF TESTING

A complete survey of the methods used to assess hot tearing is beyond the scope of this book. The interested reader is recommended to various reviews, including historical accounts by Middleton (1953), Dodd (1955) and Hansen (1975), and more recently several more brief accounts by Guven and Hunt (1988) and Warrington and McCartney (1989).

The methods fall into two main groups: (1) the various tests of straight castings solidifying in moulds in which the tensioned test piece is restrained at each end as an I-beam (sometimes known as 'dog bone') tests and (2) the ring or conical mould tests. Both these types of tests use the occurrence of cracks as a measure of hot tearing propensity. Sundry other tests include the pulling of tensile test pieces during their solidification in a tensile testing machine as a direct measure of the resistance of the casting to failure.

The I-beam has many variants. One of the most common is the arrangement of various lengths of rod castings from a single runner. Each of the rods has a T-shaped end to provide a restriction to its contraction. When metal is poured, the contraction of the rods will take place with various degrees of constraint, those with rods greater than a critical length failing by tearing at the hot spot, which is the joint between the runner and the rod. From such a test, therefore, only one result is obtained, and its accuracy is limited by the increments by which the rods increase in length. The limited discrimination provided by the limited number of bars of different lengths in this test is further confounded by the fundamentally limited reproducibility of hot shortness tests.

Other sorts of I-beam test gain a potentially more discriminating result by measuring the lengths of cracks in the hot-spot region. However, even in this case, the actual test volume of material is limited and the stress and strain distribution in the hot spot is far from uniform.

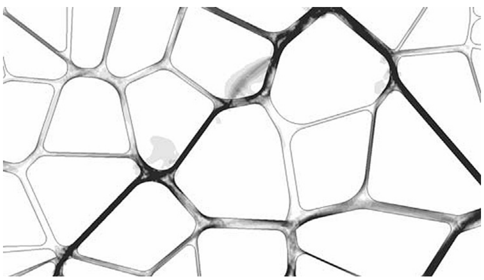

FIGURE 8.10

Preferred flow paths through a simulated intergranular mesh of equiaxed grains (Phillion et al., 2009).

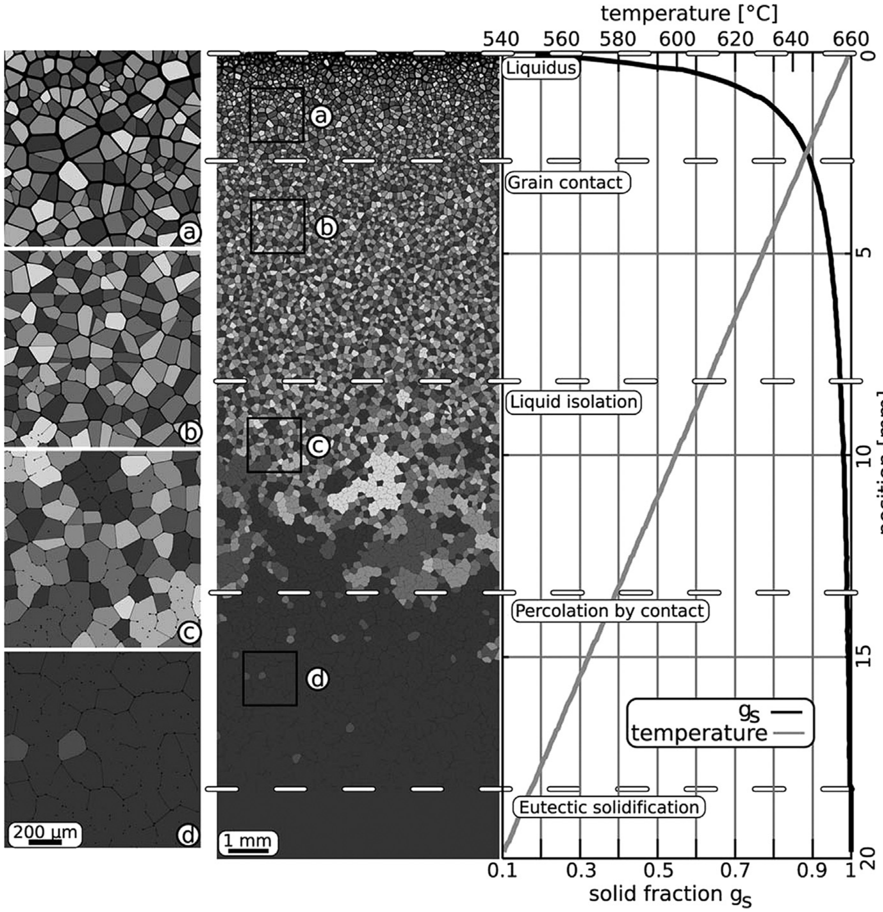

FIGURE 8.11

Calculated pasty zone of equiaxed grains for an Al-1%Cu alloy cooled at –1 K/s in a gradient of 6 K/mm. Grains in mechanical contact are shaded the same grey level (Vernede et al., 2006).

The ring test is a constrained geometry that could hardly be simpler. It is normally carried out in a steel die. Metal is poured into the open annulus between the inner and outer parts of the die. As it cools, it contracts onto the inner steel core. The core also expands slightly as it heats up. The resulting constraint on the casting is severe, opening up transverse tears all around the ring in a susceptible alloy. In this respect, the technique is useful in that it tests a large volume of material, subjecting it all to a uniform constraint condition. The method of assessment is by measuring the total length of cracks. This gives a reasonably large number that can be assessed accurately, making the test more discriminating. Thus, despite the criticisms normally levelled at the ring test, it has many advantages, one of the most important of which is simplicity. In general, the many different researches that have been carried out over the years by different workers using this approach are seen to agree tolerably well. Pumphrey (1955) carried out a series of repeated checks with different workers over a period of months. These showed an impressive consistency. This is more than can be said for most other tests for hot tearing.

Another aspect of the ring test die is worth commenting on: the test is unusual in that its rigid restraint is sufficient to cause hot tearing in the absence of any real hot spot and in the absence of strain concentration.

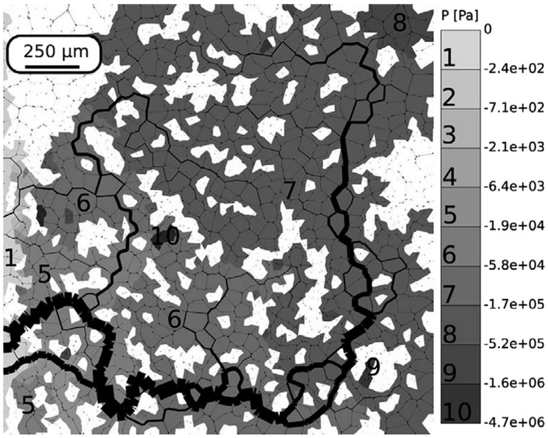

FIGURE 8.12

A model of the fluid flow and pressure distribution in an equiaxed pasty zone as a result of solidification shrinkage in Al-1%Cu alloy at 98.4% solid. The width of each channel is magnified in proportion to the local flow; local pressure is indicated by the grey scale (Vernede et al., 2006).

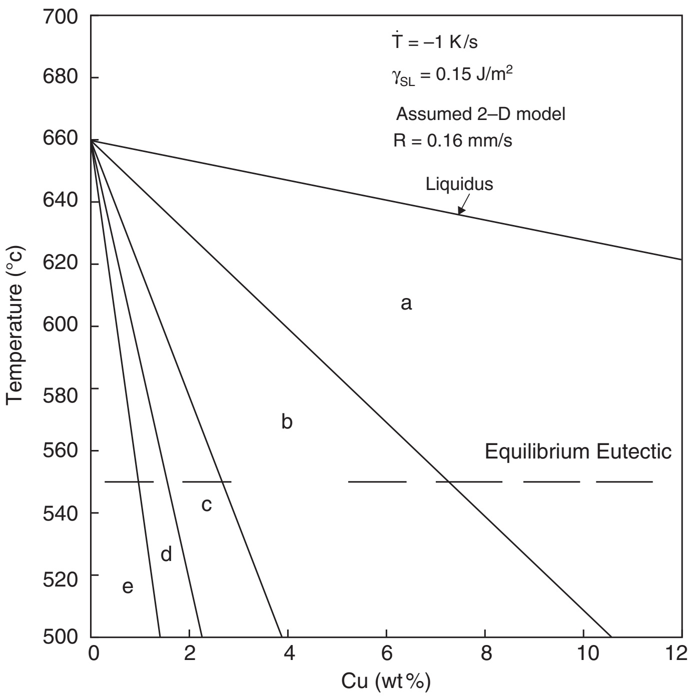

FIGURE 8.13

Map of the pasty zone based on the granular model by Vernede et al. (2006). Region (a) contains mostly isolated grains; (b) some grain contact leading to isolated clusters; (c) larger clusters containing some isolated liquid; (d) a continuous solid network but continuous films no longer exist; (e) solidification complete.

Gruznykh and Nekhendzi (1961) and DiSylvestro and Faist (1977) describe similar ring tests for steels using sand moulds.

The recent variant of the ring test by Warrington and McCartney (1989) provides a tapered centre core that gives a constraint condition increasing linearly along the axis of the cone. This seems a useful test, but is clearly limited in very hot short materials to a maximum tear of the length of the side of the cone. This artificial cutoff is seen in Figure 8.6. More work would be helpful to confirm the limitations of reproducibility of the test. For instance, from Eqn (8.7), it is to be expected that the width of the hot spot that the authors place along the cone by painting a strip of insulating wash down its length will be critical. It is not clear whether such a hot spot is necessary or even desirable in the test, but its presence is certain to have a major effect on results.

Trikha and Bates (1994) employ a nicely designed hot tearing test for steels that consists of opposed conical cores (Figure 8.14). The conical castings are gated at the horizontal joint defining a longitudinal hot spot. Any cracks in the cores can be measured, giving a pair of values to assess the reproducibility of the results. The tearing can be assessed in terms of density of cores and binder materials.

A new approach to measure the strength of Al alloys in the mushy state (but, note, not necessarily a hot tearing test) is described by Mathier et al. (2008). A specially shaped tensile specimen is contained in a hot stainless steel mould and gradually cooled. The precise dimensions of the mould define a precisely shaped specimen which in turn can be closely studied by carrying out, in parallel, a numerical simulation. In this way, the information that can be generated by such a test is maximised. The only disappointment to the reporting of this interesting work is that the filling of the test mould is not specified and has clearly been overlooked as being of critical importance to the integrity of the material under test.

8.1.10 METHODS OF CONTROL

The casting engineer can be reassured that there are several different approaches to tackling hot-tearing problems in castings, or even, preferably, preventing such defects by appropriate precautions.

8.1.10.1 Improved mould filling

In practice, as bears repeating any number of times, the author has never failed to deal with hot tearing by simply up-grading the filling system, ensuring that air is not entrained into the metal at any point. Whole foundries have been revolutionised in this way. Thus this is probably the single most important technique for dealing with hot tearing problems. It is the most convincing evidence that hot tears are strongly linked to the presence of bifilms. To the author's knowledge, sadly, this overwhelming evidence has not yet found its way into the scientific casting literature.

8.1.10.2 Casting design

Much can be achieved at the design stage of the casting. A publication by Kearney and Raffin (1987) concerns itself almost exclusively with the prevention of hot tears by adjustments to the casting design. In general it can be summarised by the following.

1. Do not design sharp re-entrant corners;
2. Do not provide a straight join between two potential hot spots; curve such members;
3. Provide curves in gates so that deformation can be accommodated easily;
4. Angle and off-set stiffeners and ribs to allow easier accommodation of strain; and
5. For gravity die (permanent mould) casting the rapid removal of any internal steel core is recommended to reduce constraint.

The reader will appreciate the general philosophy, although this severe condensation hardly does justice to the original work of these authors.

It is also worth bearing in mind that flash on castings as the result of the poor fitting of moulds and cores can be a major source of casting constraint. To identify the real sources of constraint is necessary, therefore, to check the casting straight out of the mould (not after it has been nicely dressed to appear on the manager's desk!).

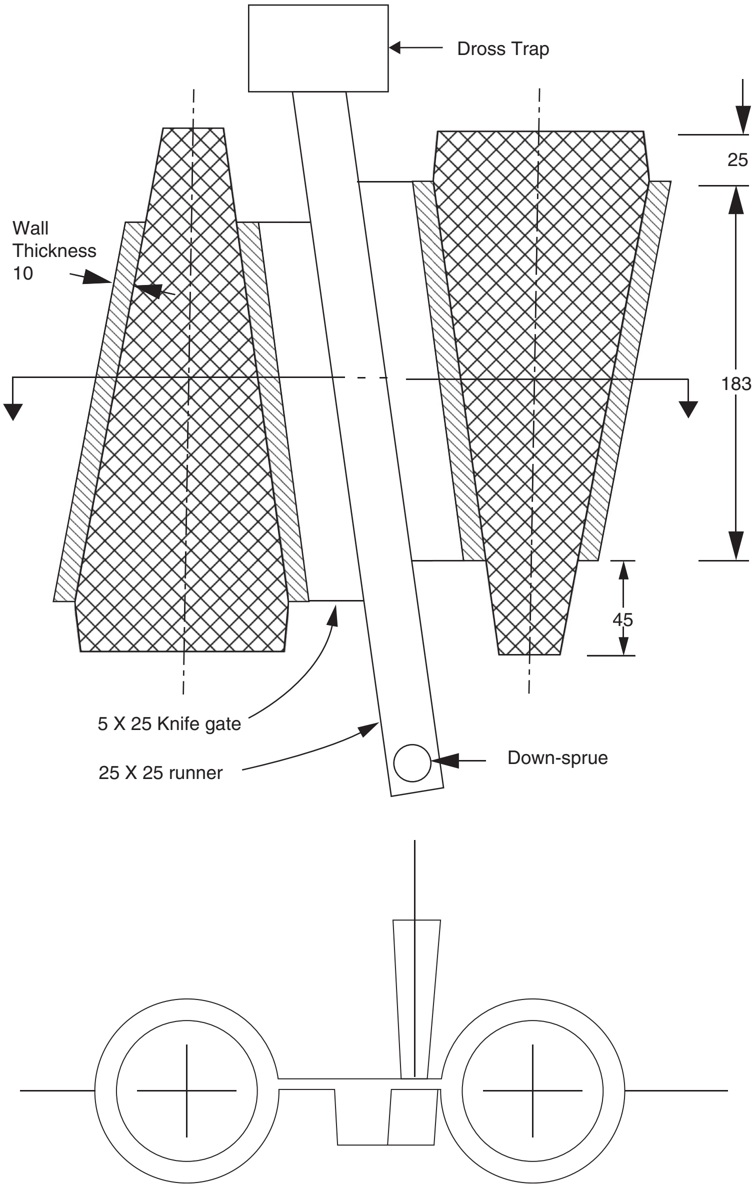

FIGURE 8.14

The hot tearing test of Trikha and Bates (1994) consisting of opposed cones, giving a pair of results that can be compared for reproducibility.

This list of dire warnings needs to be set into context. In the experience of the author, the provision of sharp corners and other so-called geometrical dangers have never given a problem providing the liquid metal is of good quality, and provided the metal has not been impaired by a poor filling system design.

In any case, it is usual to find that the design is fixed, and any changes will involve significant negotiation with the customer and/or designer, and may eventually not be agreed. In such situations, the casting engineer has to fall back on other options. These include the following.

8.1.10.3 Chilling

The chilling of the hot spot is a useful technique. This reduces the temperature locally, thus strengthening the metal by taking it outside of its susceptible temperature range before any significant strain and stress is applied. By reducing the temperature locally nearer to that of the casting as a whole, the temperature differential that drives the process is reduced, and any strain concentration is redistributed over a larger region of the casting. Local chilling is therefore usually extremely effective.

In addition to this conventional explanation, recent research has revealed that the action of chills is more complex. Chills cause the solidification front to move away from the chill in a rapid, unidirectional movement. Bifilms, with their enclosed layer of air, are the ultimate barrier to dendrites because the solidification cannot progress through the air. The result is that the bifilms are pushed ahead of the front, and so pushed away from the vulnerable hot spot.

This is the reason that many entrainment defects, as opposed to genuine shrinkage problems, appear to be cured by the placing of a chill. The usual (and usually quite wrong) interpretation being that the chill has cured the 'shrinkage' porosity.

Some oxide bifilms appear, however, to be attached to the casting surface (possibly at the point at which they were originally folded in) so that they are not completely free to be pushed ahead. Also, some are oriented so that some of the tips of the dendrites in the advancing dendrite raft pass either side of the bifilm, as seen in Figure 2.43. Either way, the bifilm is organised into a planar sheet parallel to the dendrite growth direction and then pinned in this position until solidified in place. Cracks at right angles to the chill surface can be seen in the radiograph shown in Figure 2.43. Such planes easily separate along the air layer as the result of a tensile strain. This may occur during freezing, in the form of a hot tear, as shown in Figure 8.1. Alternatively, the planes and the characteristic steps can be seen in room temperature tensile fracture surfaces as in Figures 2.44 and 2.46. Clearly, once again, the message is clear: it is far more effective to remove the bifilms than to impose chills.

8.1.10.4 Reduced constraint

A reduction in the stress on the contracting casting can be achieved by reducing the mould strength. But as we have noted before, this is more easily said than done. The options are: (1) reducing the level of binder in the sand core, although there is normally little scope for this, because those foundries not already operating at minimum strengths to reduce costs and ease shake-out are using strength as an insurance against core breakages and defects as a result of mishandling; and (2) weakening the core by using it in a less dense form (such as produced by blowing rather than by hand ramming), or modifying its design by making it hollow. Earlier shake-out from the mould, and more rapid decoring, may also be useful. However, Twitty (1960) found the opposite effect; his white iron castings suffered more hot tears when shaken out earlier as a result of the extra stresses put on the casting during the removal of the mould!

EI-Mahallawi and Beeley (1965) show how appropriate tests can be carried out on sands containing different binders. Their deformation/time test for sand under a gradient heating condition would be expected to provide a good assessment of the constraint imposed by different types of sand/binder systems. DiSylvestro and Faist (1977) use their sand moulded ring test to check the effect of different sand binders on the hot tearing in carbon steels. In later work on steel, SCRATA (1981) list the sand binders in increasing hot-tearing tendency for steel casting sections of less than 30 mm. These are:

Greensand (*least hot-tearing*)
Dry sand (clay bonded)
Sodium silicate bonded (CO_2 and ester hardened types)
Resin bonded shell sand
Alkyd resin/oil (perborate or isocyanate types)

Oil sand
Phenol formaldehyde resin/isocyanate
Furan resin (*worst tearing*)

In the case of the worst sand binder, furan resin, it is difficult to believe that the thermal and mechanical behaviour of the binder is responsible for the increased incidence of hot tearing. It seems likely that the sulphur (or phosphorus) contained in the mixed binder will contaminate the surface of the steel casting, promoting grain boundary films of sulphides or phosphides, and thus rendering the metal more susceptible to tearing.

In thicker sections, the greater amount of heat available causes more burning out of the binder, with better collapse of the sand mould. Organic binders benefit from this effect. Paradoxically, the inorganic system based on sodium silicate in this list is also seen to benefit, probably as a result of the extra heat leading to greater general softening and/or melting of the binder at high temperatures. As the binder cools, however, it is well known to become a fused, glassy mass which can be difficult to remove from the finished casting. Even so, modern silicate binders contain breakdown agents that appear to have overcome this problem.

For ferrous castings, the inclusion in the sand of material that will burn away rapidly, leaving spaces into which the sand can move, allows faster and greater accommodation of the movement of the casting. Common additions to greensand are wood flour, cellulose and polystyrene granules. Slabs of polystyrene foam of 25–50 mm thickness have been inserted in the mould or core at approximately 6–12 mm away from the casting/mould interface, depending on the thickness of the casting. Conversely, reinforcing rods or bars in sand moulds or cores greatly reduce the collapsibility of that part of the mould in which they are placed, and may cause local tearing.

In the case of gravity die casting, in which the core may be made from cast iron or steel, it is common to construct the core so that it can be withdrawn or can be collapsed inwards as soon as possible after casting. Almost all aluminium alloy pistons are made in this way, with complex collapsible five-piece internal cores.

Ultimately, however, it has to be emphasised that removal of constraints after casting is not always a reliable technique because the timing is difficult to control precisely; too early an action may result in a breakout of liquid metal, and too late will cause cracking. It is really better to rely on passive systems.

8.1.10.5 Brackets

The planting of brackets across a corner or hot spot can sometimes be useful. The brackets probably serve not only for strengthening but also as cooling fins. Perhaps predictably (because of the poorer conductivity of steel castings compared with those in aluminium, for instance), for steel castings they are generally less good at preventing tearing than are chills (SCRATA, 1981). Even so, some large steel castings are sometimes seen to be covered in ‘tear brackets’, the casting bristling like a porcupine. Such features confirm the existence of their poor filling system designs.

8.1.10.6 Grain refinement

In general, fine grains reduce the susceptibility to hot tears, as predicted by Eqn (8.7). Figure 8.9 shows the improved resistance to tearing by grain refinement of Al-Cu alloys. Davies (1970) finds a similar result for other Al alloys. Novikov and Grushko report the beneficial effects of refinement by Sc and Zr in Al-Cu-Li alloys. Twitty (1960) confirmed that his white cast iron alloyed with 30% chromium was severely hot short when not grain-refined, whereas 0.10–0.25Ti addition reduced the grain size and eliminated hot tearing.

However, in a really illuminating experiment to reduce hot cracking of a direct chill (‘continuous’) Al alloy, Nadella et al. (2007) introduce grain refiner in the ladle before casting, successfully grain refining and eliminating cracking. However, when they added the same grain refiner into the metal stream during pouring into the mould, the casting was successfully grain refined, but cracked. This is clearly an instance of the Ti-rich grain refiner precipitating on bifilms in the ladle and sedimenting to the bottom of the ladle, so that the cast metal was clean. When the grain refiner was added to the launder bifilms were clearly unable to separate, and thus were carried over into the casting, creating excellent grain refinement, but also creating excellent conditions for cracking in the presence of cooling strains. This experiment seems an excellent vindication of the critical action of bifilms in hot tearing and cracking.

8.1.10.7 Reduced casting temperature

Reduction of the casting temperature can sometimes help, as is seen clearly for Al-Cu alloys (Figure 8.9). This effect is likely to be the result of the achievement of a finer grain size. If, however, the effect also relies on the reorganisation of bifilms, then the effect may be sensitive to geometry: some directions may become more prone to failure if bifilms are caused to lie across directions of tensile strain.

8.1.10.8 Alloying

A variation of the alloy constituents, often within the limits of the chemical specification of the alloy, can sometimes help.

The addition of elements to increase the volume fraction of eutectic liquid can be seen to help by (1) increasing the pre-tear extension by lubricated grain boundary sliding and (2) decreasing the CSC. Couture and Edwards (1966) confirm that copper-based alloys benefit from increased amounts of liquid during the final stages of freezing.

Manganese in steels is well known for reacting with sulphur to form MnS, so that the formation of the deleterious FeS liquid at the grain boundaries is reduced.

For other more complicated alloy systems the answers are not so straightforward. For instance, in early work on the Al-Cu-Mg system by Pumphrey and Lyons (1948), the relation between hot tearing and composition is complex, as was confirmed by Novikov (1962) for various Al-Cu-X systems. Ramseyer et al. (1982) investigated the Al-Cu-Fe system and found that for certain ranges of composition increasing levels of iron were desirable to control tearing. This is a most surprising conclusion which most metallurgists would not have predicted; iron is usually an embrittling impurity in most high-strength aluminium alloys when assessed by room temperature tests of ductility. Later work by Chadwick (1991) revealed that the effect of the iron is to provide a network of iron-rich crystals around the primary aluminium dendrites, as a scaffold framework that appears to support and reinforce the weaker dendrite array. When attempting to reinterpret this fascinating result in terms of bifilm theory, it seems that the scaffold framework would consist of beta-Fe particles which had grown on and straightened bifilms. However, the beta-Fe framework appeared to have impressive rigidity and strength, despite, we assume, its associated central cracks.

8.1.10.9 Reduced contracting length

Shortening the length over which the strain is accumulated is sometimes conveniently achieved by placing a feeder in the centre of the length. Such a large concentration of heat in the centre of the cooling section will allow the strain to be accommodated in the plastic region close to the feeder. Any opening up of intergranular pathways is likely to be easily fed from the nearby feeder. The careful siting of feeders in this way effectively splits up a casting into a series of short lengths. If each length is sufficiently short, then strains that can cause a tear will be avoided. The siting of insulating coatings can also be used in the same way. In terms of Eqn (8.7), the technique is equivalent to a method of reducing the strain concentration by multiplying the number of hot spots and thus increasing the total length l of the hot spot whilst reducing the contracting length L.

8.1.11 SUMMARY OF THE CONDITIONS FOR HOT TEARING AND POROSITY

The findings of research such as that by Spittle and Cushway (1983) are summarised schematically in Figure 8.15(a). The benefits of normal metallurgical controls during casting such as fine grain size and reduced casting temperatures are clear. However, a liberty is taken to add to the figure the newly discovered but as yet relatively little known overriding benefits from clean metal.

The differences between the conditions for the occurrence of porosity and the occurrence of hot tearing have been mentioned several times. Figure 8.15(b) represents a summary of these conditions.

Clearly, porosity forms in the hydrostatic tension as a result of poor feeding, especially in the kind of conditions found in a pasty freezing alloy. The interdendritic feeding leads to a reduction in pressure described by the so-called Darcy relation, in which laminar flow through interdendritic channels suffers viscous drag, causing the pressure to fall. Thus these conditions exist in those alloys that solidify as a solid solution, growing as dendrites whose interdendritic

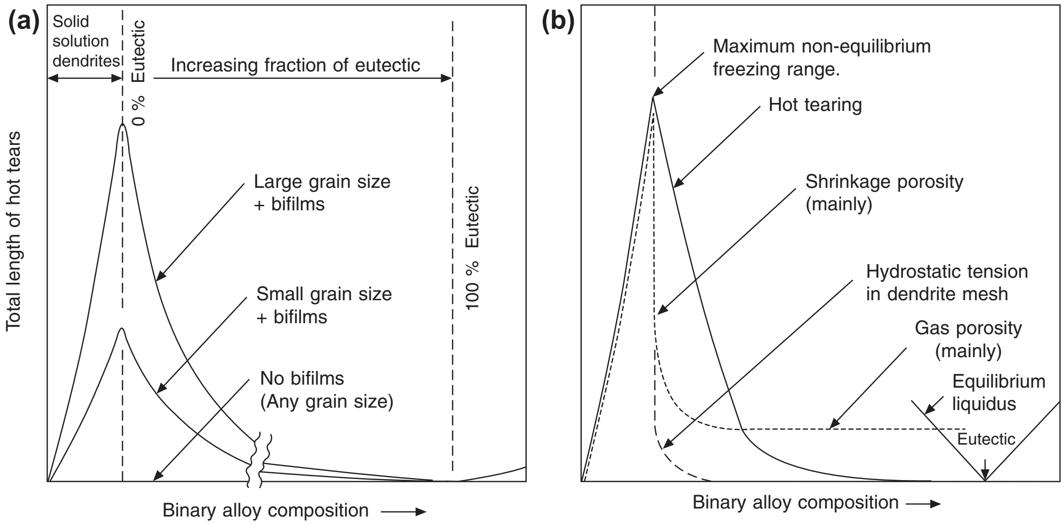

FIGURE 8.15

Summary of (a) the modest effects of grain size and the powerful effects of bifilms on hot tearing and (b) the close and potentially confusing relation between the conditions for porosity (dependent on hydrostatic pressure) and hot tearing (mainly dependent on uniaxial strain).

channels taper down to nothing, therefore maximising viscous drag, and maximising the potential for the creation of porosity.

Increasing the solute content of the alloy sufficiently to form some eutectic, particularly in non-equilibrium freezing conditions, the hydrostatic (triaxial) tension is immediately greatly reduced because the dendrite channels now no longer taper to a point, but are now truncated with a planar front of eutectic. The potential for porosity now falls precipitously as is clear in Figure 8.15(b).

If any uniaxial tension exists, the potential for hot tearing at first increases similarly to the potential for pore formation, but the steep drop at the first appearance of the non-equilibrium eutectic liquid may or may not correspond to the hot tearing susceptibility depending on how this is measured experimentally.

The different regimes of triaxial and uniaxial tension correspond to the regimes for the incidence of porosity and hot tearing in alloy systems as illustrated in Figure 8.15. The continuing low level of porosity at higher solute contents approaching the eutectic is the result of a residual amount of gas porosity. This too, of course, usually only occurs because of the presence of a population of (usually) relatively small bifilms.

The maximum in the freezing range corresponds to the maximum potential for porosity and hot tearing (Campbell, 1969a,b; Rappaz 1999). Interestingly, this peak also corresponds closely to the compositions of many wrought alloys. This is because the wrought alloys have been optimised for maximum alloy content in solution, avoiding the presence of hard eutectic phases which contribute little additional strength and reducing other properties such as resistance to corrosion. Thus these alloys should suffer maximum hot tearing and cracking problems during casting. The only reason such compositions are successful in practice is because of the great precautions taken by the wrought industry to cast metal as clean as possible, and under the highest temperature gradient possible provided by intense water cooling. Even so, the continuous casting industry is troubled by cracking of ingots, particularly of some of the stronger alloys, and particularly at the start of casting. This is because after all the cleaning of the melt, the cast is started by the melt being dropped several hundred millimetres into the mould where, of course, it re-creates huge bifilms. These cause the cracking of the cast material around the dovetail key at the top of the starter bar. Although, once the mould is filled, the remainder of the pour creates no further bifilms; the cracking troubles sometimes extend a long way up the length of the casting because of the mixing and progressive dilution of the original melt pool as the cast proceeds, spreading the original bifilms that are free to float far up the length of the product.

8.1.12 HOT TEARING IN STAINLESS STEELS

The failure of stainless steel castings during solidification and cooling can be complicated by the phase changes that can occur during freezing. The critical parameter is the Cr_{eq}/Ni_{eq} ratio as discussed in some detail in Section 6.6. Nayal (1986) does not hesitate to make the useful, if obvious point, that it is the structure of the steel during freezing that is important, rather than the structure at room temperature. He categorises the steels according to their structural reactions during freezing:

A.	$L \rightarrow L + \gamma \rightarrow \gamma$
	$L \rightarrow L + \gamma \rightarrow L + \gamma + \delta \rightarrow \gamma + \delta$
B.	$L \rightarrow L + \delta \rightarrow L + \gamma + \delta \rightarrow \gamma + \delta$
C.	$L \rightarrow L + \delta \rightarrow \delta \rightarrow \gamma + \delta$

The range of solidification mode B occurs over values of Cr_{eq}/Ni_{eq} from 1.49 to 2.0. In later work, the researchers find in the case of freezing entirely in mode B that no cracks were observed, whereas the presence of some component of A and C always led to some cracking.

The researchers Kujanpaa and Moisio (1980) find that P and S are particularly harmful in encouraging tearing in steels that solidify purely to austenite, but do not affect those that solidify with some delta ferrite. The division between the two steels is sharply defined at $Cr_{eq}/Ni_{eq} = 1.5$. Another Scandinavian worker, Rogberg (1980), confirms this general rule for the particular case of two steels pulled at high temperature in a tensile testing machine. The steel solidifying to austenite was sensitive to a wide range of impurities, but those containing ferrite seemed insensitive to these problems.

8.1.13 PREDICTIVE TECHNIQUES

The ability to predict the occurrence of tearing in a casting would be valuable. This quest has driven several different approaches. One of the early attempts assuming the rheological behaviour of solidifying alloys has been made by Qingchun et al. (1991), with some success. A not dissimilar viscoplastic model has benefited from the power of computers by Pokorny, Monroe and Beckermann (2009), who have shown that the sites and severity of hot tears appears to be predictable in simulations of relatively complex light alloy castings. These forerunners of examples of early computing approaches are likely to become routine provisions of commercial computer simulations in due course. But for more scientific insights into the mechanisms of hot tearing, the highly descriptive granular model under development by Rappaz et al. (1999–2009) is a likely contender as a future useful model to understand and predict tensile failure in the partly solidified state. This model describes the intergranular and interdendritic flow through the pasty zone, defining a condition in which viscous drag grows to reach a nucleation threshold for a pore, which is then assume to grow into a crack. The model may one day be tested for bifilm-free melts, but in the meantime appears not to apply as a result of bifilms blocking all intergranular and interdendritic flow of liquid as demonstrated by the experiments of Fuoco (Figure 6.18).

8.2 COLD CRACKING

8.2.1 GENERAL

Cold cracking is a general term used to emphasise the different nature of the failure from that of hot tearing. Whereas the word ‘hot’ in hot tearing implies a failure occurring at temperatures above the solidus, ‘cold’ simply means lower than the solidus temperature; in some cases, therefore, it can be rather warm! Solidus here means, of course, the real non-equilibrium solidus (not necessarily the value picked off an equilibrium phase diagram!).

The term ‘cracking’ is also to be contrasted with ‘tearing’. Whereas a tear is a ragged failure in a weak material, a crack is more straight and smooth and occurs in strong materials. Because it represents the failure of a strong material, the stress required to nucleate and propagate the failure is high (in hot tearing stress was generally insignificant, whereas strain was important).

Occasionally, a failure appears to fall somewhere between the tear and crack categories. Such borderline problems include the cracks that form in steels because of the presence of low melting point residual elements, such as copper-rich phases, at the grain boundaries of steels. The copper-bearing phase is liquid between 1000 and 1100°C, with its dihedral angle falling to zero in this temperature range, thus wetting the grains. Over this range of temperature, therefore, steel is particularly susceptible to tensile failure (Wieser, 1983). The temperature is well below the solidus (in terms of the Fe-C system), and cracking will not occur easily at higher temperatures because the liquid phase does not wet so well and therefore does not cover such a large proportion of the grain boundary area.

Returning to the ‘cold’ crack, the driving force for the nucleation and spread of a cold crack is stress. The various ways in which stress can arise and be concentrated in castings have been dealt with previously and are not repeated here. However, one mechanism not previously discussed is phase change in the solid state. In steels transforming from a δ- to a γ structure, the large volume reduction has been suggested as a potential source of stress because of the large strains involved (Gelperin, 1946; Grill and Brimacombe, 1976). In fact, the volume change occurring during the collapse of the body centred cubic ferrite phase to the close packed austenite phase is close to 1.14%, corresponding to a linear contraction 0.38% (being one third of the volume contraction). This is a massive linear strain, well over the yield point (in the region of a 0.05–0.1% strain), and so would be expected to open up any fragments of bifilms to form the nucleation sites of larger cracks that would propagate as the steel cooled, generating high stresses to propagate the cracks to observable size. This mechanism seems to be the cause of the widely experienced cracking of low carbon and low alloy steel continuously cast billets in the carbon range 0.05–0.20 with a peak at 0.10°C. Figure 8.16 is given as an example.

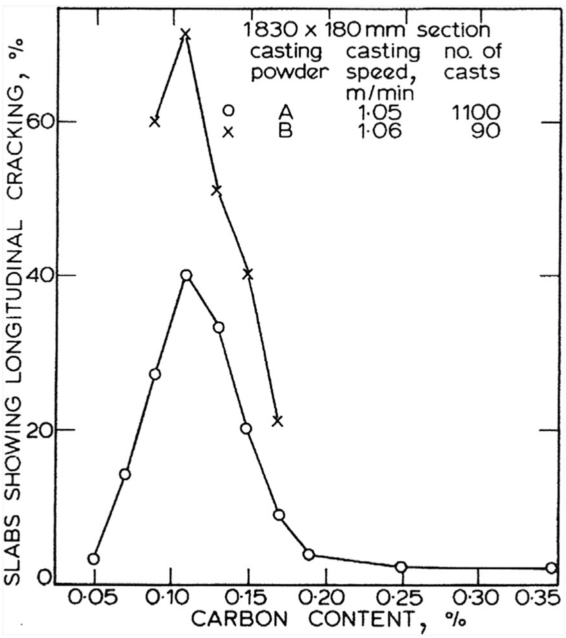

FIGURE 8.16

Effect of carbon content on the longitudinal cracking of continuously cast steels.

The peak corresponds nearly exactly with the peak extent of the ferrite plus austenite transformation range. The reduced cracking at less than 0.05°C and higher than 0.20°C reflects the solidification of the steel directly to delta ferrite or to austenite respectively, avoiding the strains associated with the solid state transformation between the two. These considerations are in agreement with Nayal's list.

8.2.2 CRACK INITIATION

Cracks start from stress raisers. A stress raiser can be an abrupt change of section in a casting. However, this problem is well known to designers, and in any case is not likely to cause increases of stress by much more than a factor of two.

More severe stresses are raised by sharper natural defects which are assumed to be present in all metals, known as Griffith cracks. I have had problems acknowledging the existence of Griffith cracks because such features cannot be generated by solidification, and recent computer studies (molecular dynamics) appear to show that cracks cannot exist in solid metal until the theoretical strength of the metal is exceeded. Such stresses are in the realm of many tens of GPa. The molecular dynamic studies indicated that volume defects such as pores and cracks cannot exist in the metal lattice because the bonds between atoms are so strong; any such volume defect collapses to form dislocation rings and stacking fault tetrahedra and other similar atomically dense, coherent structures. The logical outcome is the conclusion that bifilms, whose presence *is* to be expected, actually *are* the Griffith cracks (JC, 2011).

These cracks which are cast into place at the time of the filling of the casting are all the more dangerous because they can occupy a large portion of the section of a casting, but at the same time are difficult to detect. Oxide bifilms are probably the most important initiators of cracks in castings of all types. This is clear because from daily foundry experience, castings made with good running and gating systems are usually not sensitive to problems of cracking; the heads of large castings can be flame cut off without the generation of any cracks.

It is important to keep in mind that if it is true that all cracks initiate from bifilms, which I think is not only possible but probable, then cracking of metals can be controlled, and can be eliminated. The prospects of eliminating failure by cracking would revolutionise metallurgy and engineering. Such thoughts are so exciting to be beyond belief. Time will tell whether this will come to pass.

Turning back once again to more pedestrian issues, it is interesting to consider briefly the welding of steels. Stresses involved during and after the solidification of a weld can be extremely high because of the constraint of the surrounding solid metal. This surrounding material is near to room temperature and thus very strong. In such a case, Dixon (1987) has observed that cracks start from such innocuous sites as micropores. However, such small and relatively rounded defects are unlikely initiators despite the very high driving force. It is also most likely that the micropore is sited on a bifilm, and it is the bifilm that causes the initiation of the crack. This is just one illustration of the possible universality of cracks only forming from bifilms. Only very careful observation using the scanning electron microscope would be capable of resolving such issues.

To clarify further the issues relating to welding: it seems probable that all welding cracks originate from bifilms introduced into the parent plate material when originally cast and which have survived processing into plate and sheet. Thus the huge amount of research on ways to reduce weld cracking almost certainly require to be redirected from the weld conditions and turned to address the casting conditions of their parent plate materials.

In the case of the strong 7000 series Al alloys, Lalpoor et al. (2009) used finite element analysis to estimate the stresses in semi-continuously direct chill cast billets and slabs. Knowing the levels of stress, they estimated the critical defect sizes to initiate catastrophic failure. The diameters of such critical penny-shaped cracks need to be between 3 and 10 mm to initiate failure. Oxide bifilms are to be expected easily exceeding this size as a result of the turbulent initial fill of the mould. In fact, bifilms up to 100 mm across would not be surprising. Thus sudden brittle failures are to be expected.

8.2.3 CRACK GROWTH

As the casting cools, stress-relaxation by creep becomes progressively slower and eventually stops altogether. Thus from this stage onwards, further contraction only builds up elastic strain and consequent stress. The accumulating stress is available to grow the crack with great speed to large size. Lees (1946) reports the casting of a high-strength aluminium alloy into a sand mould designed to produce hot tearing by constraining the ends of the cast bars. If the restraint was not released before the castings cooled to 200°C, then a loud crack was heard, corresponding to the complete fracture of the bar. Similar failures during the cooling of steel castings are also well known, as was, for instance, reported by Steiger as long ago as 1913.

In steels, if the crack is open to the atmosphere, the colour of its surface is a useful guide to when it formed: an uncoloured metallic surface will indicate that the crack occurred at a temperature near to room temperature; the normal 'temper colours' (the light interference colours reflecting the thickness of the oxide) range from light straw, formed at somewhere near 300°C, through yellow, blue and finally to brown indicate greater exposure to time at temperature, with temperatures probably approaching 600–700°C for the darker colours.

After the strains of delta to austenite transformation between about 1400 and 1500°C, depending on the carbon content, during the further cooling of steel castings there are a succession of particularly vulnerable temperature ranges. The following list is not expected to be exhaustive.

1. Carbon steel castings will embrittle if they dwell for an excessive period between approximately 950 and 600°C (Harsem et al., 1968). Butakov et al. (1968) cites hydrogen, sulphur and phosphorus as increasing embrittlement in the 900–650°C range, and the intergranular fracture surface as exhibiting various forms of sulphides, particularly MnS and FeS. Harsem et al., (1968) adds carbides and AlN to this list. From our privileged gift of the wisdom of hindsight, it seems probable that many of these researches were influenced by the presence of bifilms. At least some of these problems are likely to be associated with the strains of the austenite to ferrite transition. New, clarifying researches are now needed.
2. Impure low-alloy steels suffer similarly in the range 550–350°C (Low, 1969).
3. Low-carbon steels are susceptible to brittle failure if deformation occurs in the range 300–150°C, the so-called blue brittleness, or temper embrittlement range (Sherby, 1962). In his review, Wieser (1983) lists the principal contributors to temper embrittlement as antimony, arsenic, tin and phosphorus. These elements have been found to segregate to the prior austenite grain boundaries. Because these boundaries will in general coincide with the ferrite boundaries, then the crack will usually be intergranular once again. However, the effect of these and other elements in the various kinds of steels is complicated, not being the same in every case; for instance, tin at concentrations up to 0.4% embrittles Ni-Cr-Mo-V steel but not 2.25Cr-1Mo steel. Copper, manganese and silicon are also deleterious in some steels, whereas the effect of molybdenum is generally beneficial.
4. Most body-centred cubic iron alloys become brittle at subzero temperatures, −150 to −250°C. This is often known as the cold brittleness range. The ductile to brittle transition is one of the most important parameters which many steels are required to meet before being accepted for service. It is generally determined from a series of impact tests at successively lower temperature.