# DEFECTS IN CASTINGS

Introduction

The aim of this lecture is to introduce you to the formation and prevention of solidification defects in castings (**Figure 3207.00.01**). These can be sub-divided into three main categories:

. Gas porosity;

. Shrinkage porosity;

. Hot tearing and cracks.

Castings can unfortunately also sometimes contain other types of defects, such as inclusions of slag or moulding sand, but these are not classified as solidification defects and will not be covered in this lecture.
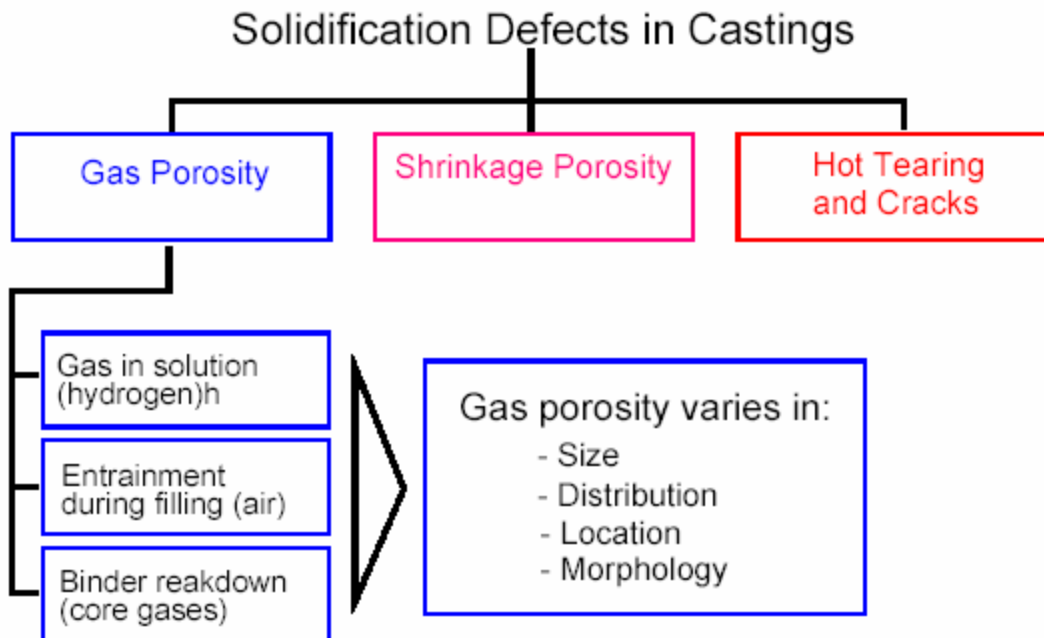
Gas Porosity

We shall start by considering **gas porosity**. This can again be sub-divided into a further three causes:

. Firstly, gas held in solution in the molten metal can be precipitated as the metal solidifies, simply as a result of the reduced solubility on freezing.

. Secondly, if the mould is filled under very poor conditions, air can be entrained in the metal stream and then trapped as the metal solidifies.
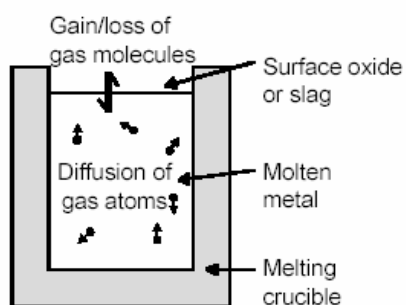
. Finally, the sand binders used to make the moulds and cores often break down when in contact with the molten metal and the gaseous decomposition products can force their way into the solidifying metal, leading to defects which are normally known as 'blows'. These different types of gas porosity defect vary in their size, distribution, istance below the casting surface and morphology. It follows therefore that the cause of such defects in a real casting can be deduced from a careful examination.

a) Gas "precipitation"

Sources of Hydrogen in Castings

- Melting and/ or subsequent handling
  - damp refractories
  - gas/ oil-fired furnaces
- Passage through the running system
- Reaction with mould/ core materials

❑ **Furnace**

Gain/loss of gas molecules

Surface oxide or slag

Diffusion of gas atoms

Molten metal

Melting crucible

❑ **Mould**

- Metal dies:
  - dry, relatively free from H_2
  - metal being cast tends to lose H_2

- Chemically-bonded and greensand moulds:
  - Heat → steam → decomposition to H_2

$$3 H_2O + 2 Al = 3 H_2 + Al_2O_3$$

  - Metal being cast tends to gain H_2

The first type of gas defect that we shall consider is that caused by precipitation of gas from solution in the liquid metal but we will initially need to understand how the gas gets into the metal in the first place (see **Figure 3207.00.02**). In the case of aluminium, we are particularly concerned about hydrogen, which can come from several sources:

1. Melting and/or subsequent handling: a common problem is hydrogen pickup from the use of damp refractories in furnaces or ladles. Another source is from burning hydrocarbon fuels, such as gas or oil.

2. Reaction with the mould during passage through the running system.

3. Reaction with the mould and core materials during and/or after filling.

In practice, however, source (1.) above is usually the only mechanism under the direct control of the casting technologist by the employment of effective degassing of the liquid metal prior to casting. Gas in solution in a liquid metal is in the form of atoms. These can diffuse to the surface, combine to form gas molecules, and evaporate into the environment. A furnace gains or loses gas from contact with its environment, the rate of transfer of gas depending, of course, on the ratio of surface area to volume. There are various ratecontrolling steps in the transfer from the furnace to the atmosphere and vice versa, and any or all of them may be operative in different situations. The environment of the furnace is complex: the top surface of the liquid may be in contact with the air and so able to equilibrate directly with the atmosphere. However, in many cases, a surface oxide film may be present, or a slag or flux layer. These additional layers will present a further barrier to the passage of gas atoms emerging from the metal, slowing equilibration in furnaces even further. Conditions above the liquid may also be changing rapidly as waste combustion products, high in water vapour, are

directed onto the surface, or blow across from time to time. The environment of the liquid metal in the mould is perhaps a little clearer. If the mould is a metal die, then the environment is likely to be dry and thus relatively free from water vapour and its decomposition product, hydrogen. The liquid metal may lose hydrogen to this environment, since the equilibrium pressure of hydrogen in the melt will be less than that of the partial pressure of the environment. In contrast, if the mould is made from sand, either chemically-bonded or especially if bonded with a clay-water mixture as in a greensand mould, then the environment all around the metal will contain nearly pure steam at close to one atmosphere pressure. The water will decompose in contact with the aluminium as follows:

$3 O + 2 Al = 3 H_2 + Al_2O_3$
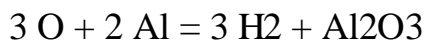
Thus steam will yield equal volumes of hydrogen gas, still at one atmosphere pressure, which will be available for solution in the liquid aluminium. It is likely that the melt willgain hydrogen in this environment.

Gas precipitation from solution in the metal leads to small bubbles, normally in the size range 0.05 - 0.5 mm, as a result of the high internal pressure of gas due to the microsegregation between the dendrite arms. The bubbles are distributed uniformly throughout the casting, with the exception of a bubble-free surface layer about 1 - 2 mm deep. **Figure 3207.00.03** shows a typical example of hydrogen bubbles in a simple aluminium casting and emphasises the size and distribution characteristics.

Hydrogen Porosity

Characteristics:
- ❑ Small bubbles 0.05 - 0.5 mm diameter
- ❑ Even distribution in casting
- ❑ Bubble-free surface layer 1 - 2 mm deep
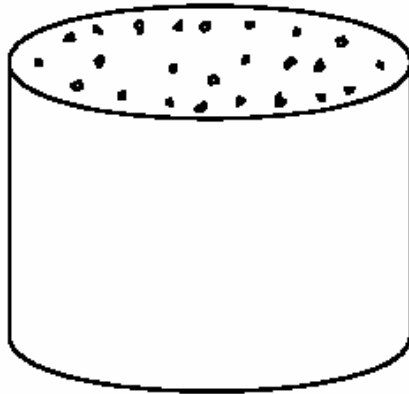
Typical porosity due to precipitation of hydrogen revealed on the cut surface of a cylindrical casting

Figure 3207.00.04 shows a schematic view of a section through a solidifying casting with the mould on the left-hand side, a solidified layer about 1 - 2 mm thick which is free from gas bubbles, and then the dendritic growth front. As further growth of the dendritic front occurs, we find that small hydrogen gas bubbles are precipitated and become trapped in the dendrite forest. We shall now examine the reasons for this behaviour.

Hydrogen Precipitation
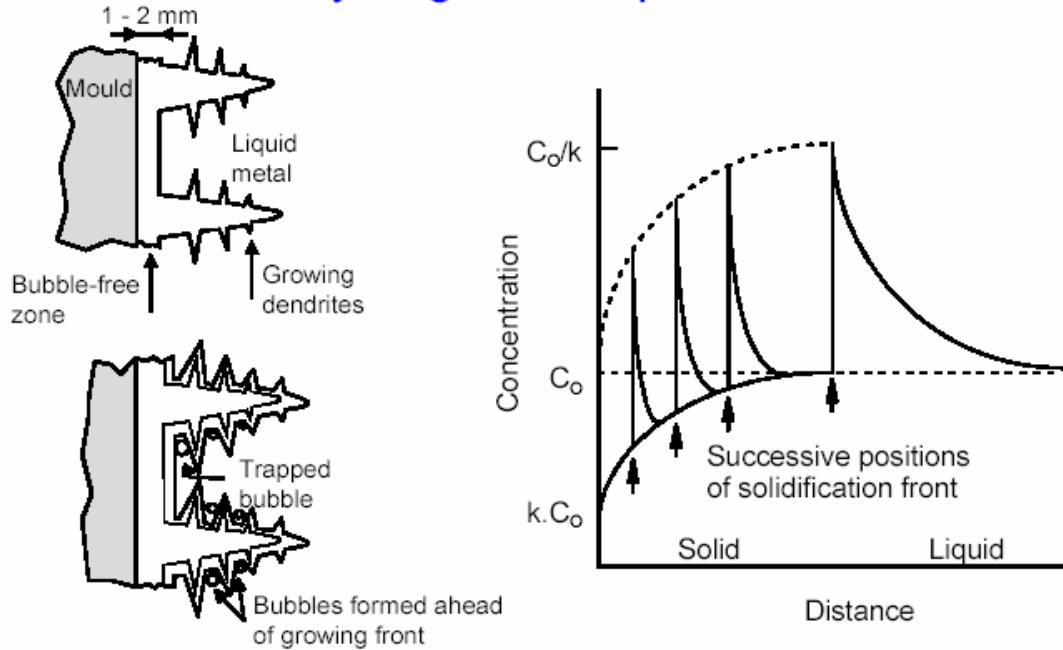
If we consider the concentration changes around the tip of the dendrite as it grows, we have previously seen (**TALAT Lecture 3204**) that when a liquid metal with an initial solute concentration of Co solidifies, the first solid to form has a composition of kCo, where k is the distribution (or partition) coefficient. As solidification proceeds, so more and more solute continuously builds up ahead of the advancing front in a snow-plough effect, as shown in this series of 'snap-shots'. Eventually, steady state is reached to give a concentration at the interface of Co /k. **Figure 3207.00.05**: Hydrogen in solution in molten aluminium behaves in this way and has a value of k of ~ 0.05. Hence solidification leads to an increase in the amount of hydrogen dissolved in solution of 1/0.05 = 20 times. Thus if the initial gas content is 0.1 ml/100g, which is not particularly high, the hydrogen content at the interface would be 2 ml/100g once steady state is reached. This value is

above the solubility limit and thus the liquid aluminium is supersaturated with hydrogen.

Hydrogen in Molten Aluminium

Distribution coefficient k = 0.05

If initial hydrogen content = 0.1 ml/100g then steady-state content = 0.1/ 0.05 = 2 ml/100g

$$H_2 \text{ (gas)} \rightleftharpoons 2 \text{ [H] (metal)}$$

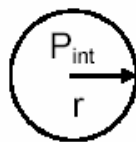Sievert's Law: $K \cdot P_{H_2} = C_H^2$

where
K = equilibrium constant
P_{H_2} = partial pressure of molecular hydrogen
C_H = concentration of atomic hydrogen in solution

Hence solidification leads to: 20x increase in hydrogen content
and   400x increase in gas pressure.

For mechanical equilibrium: $P_{int} - P_{ext} = 2T/r$
where T = surface tension ~ 1 N/m² for aluminium.
So if P_{int} = 0.1 atmophere = 0.1 x 10⁵ N/m² and neglecting P_{ext},
r ~ 2T/P_{int} ~ 0.5 µm.

We now need to think about the concentration of hydrogen in the melt in terms of pressure in equilibrium with the melt. Diatomic gases such as hydrogen dissociate when they dissolve in metals and form monatomic solutions. For example, when hydrogen is in solution in aluminium, the reversible reaction is:

H2 (gas) ™ 2 H (metal)
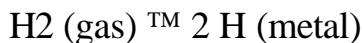
Sievert's Law states that gas will dissolve in the metal in proportion to the square root of the partial pressure in the gas phase, i.e.

$K.P_{H2} = C_H{}^2$

where

K is the equilibrium constant (which is a function of temperature),

PH2 is the partial pressure of molecular hydrogen, and

CH is the concentration of atomic hydrogen in solution.

Thus the gas pressure is proportional to the square of its concentration in the liquid. We have already seen that solidification increases the gas *content* by a factor of 20, and Sievert's Law shows us that this leads to a $20^2 = 400$-fold increase in the equilibrium gas *pressure*. The front has to advance by about 1 - 2 mm to reach these steady state conditions, so the skin of the casting is generally free of pores. This phenomenon is known as sub-surface pinholing. For the mechanical equilibrium of a bubble of radius **r** and surface tension T having an internal pressure Pint and located in molten metal imposing an external pressure of Pext,

$$P_{int} - P_{ext} = \frac{2 \cdot T}{r}$$

If the metal originally contained gas equivalent to an equilibrium pressure of, for instance, 0.1 atmosphere ($= 0.1 \times 10^5 N/m^2$), then at the solidification front, neglecting Pext,

$$r = \frac{2 \cdot T}{P_{int}} = \frac{2 \times 1 N/m}{400 \times 0.1 \times 10^5 \ N/m^2} = 0.5 \ \mu m$$

Smaller bubbles than this will disappear, slowly dissolving away as they are compressed by the effect of surface tension. Larger bubbles will grow. Sizes up to 25 - 100 ⌠m are common. Up to 500 or even 1000 ⌠m is rarer.

The final point about gas porosity is that nucleation of gas bubbles continues as metal continues to solidify (see **Figure 3207.00.06**). This leads to an even distribution
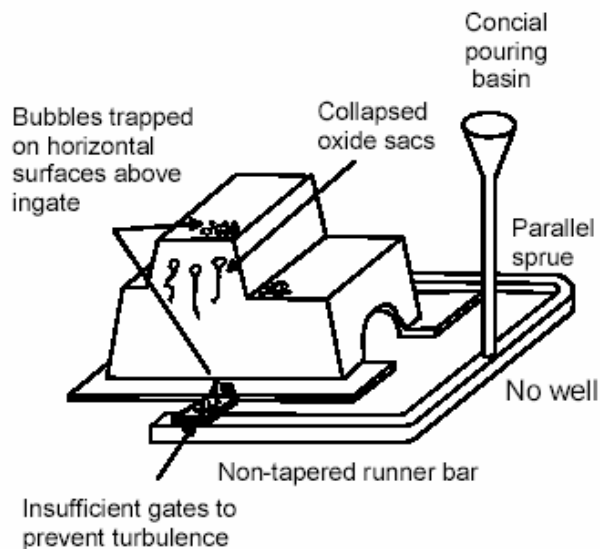
Progress of Gas Precipitation
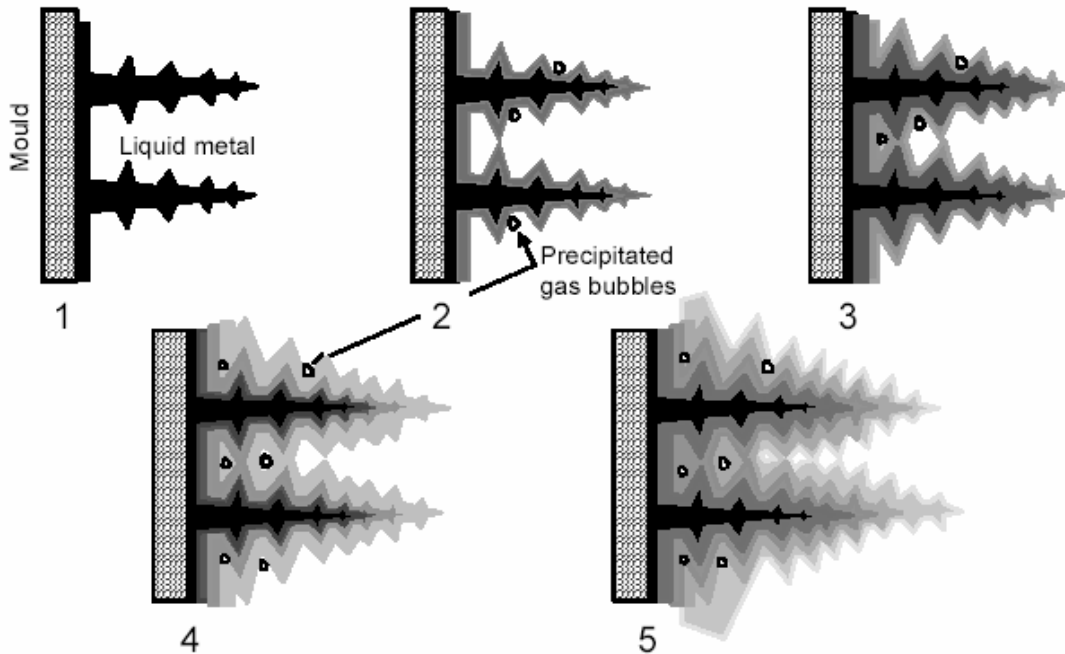
b) Air Entrapment

Moving on to the entrapment of air, we shall take as an example a sump casting that has been deliberately made badly using a conical pouring basin, a parallel downsprue and no well base (**Figure 3207.00.07**). In addition, we have a non-tapered runner bar and insufficient gates. As we now know from **TALAT Lecture 3203**, such a running system generates surface turbulence in the metal stream as it fills the mould, leading to a chaotic, scrambled mess of metal and air. The air cannot escape easily because it is held in place by the oxide film. Furthermore, as the air bubbles move through the molten metal, they leave behind a collapsed sac of oxide, forming a bubble trail which is another form of defect in the casting.

Air Entrainment

☐ Normally caused by incorrect running system design

Concial pouring basin

Bubbles trapped on horizontal surfaces above ingate

Collapsed oxide sacs

Parallel sprue

No well

Non-tapered runner bar

Insufficient gates to prevent turbulence

Characteristics:

☐ Bubbles trapped on horizontal surfaces above ingate and under ledges and apertures in casting.

☐ Irregular in size.

☐ Normally 0.5 - 5 mm

Solution:

☐ Improve running system

We find that the bubbles tend to get trapped on horizontal surfaces, such as above ingates, on the cope surfaces or under any window-type features in the vertical sides of a casting. These bubbles are intermediate in size between those precipitated from solution and those blown from cores. They are also irregular in size, reflecting the andomness, or chaos, inherent with turbulence. They normally fall into the size range 0.5 - 5 mm and are often only found when ingates are cut off or the casting is shot blasted or machined. Since they arrive with the incoming metal, they are always close to the casting surface, and usually only the thickness of the oxide skin separates them from the casting surface. This partly explains the size range of the bubbles: they are only the remnants which were too small to generate sufficient buoyancy force to break through the oxide on the surface of the liquid, whereas their bigger neighbours escaped. When viewed on a polished cross section under the optical microscope, the bubbles are always seen to be

associated with considerable quantities of oxide films - the remnants of bubble trails. It is important to diagnose this type of defect correctly. It is all too often hought - incorrectly - to be 'gas' but the problem will certainly *not* be solved by degassing the

metal. The solution will almost certainly lie in the design of the running system (i.e. the methoding) of the casting.

c) Gas coming from cores

The final type of gas defect is blown from cores (**Figure 3207.00.08**). When a metal is poured into a sand mould containing cores, the gas present in the core expands and attempts to escape. Furthermore, the resin binders used in core manufacture start to break down and generate additional gas. The gas can escape from the core via the core prints, but if the core prints are too small or if the mould and core have a low permeability, the gas pressure will build up inside the core. If the pressure reaches the level where it exceeds the opposing pressure of the molten metal, a bubble can be formed in the metal and float up towards the top of the casting.
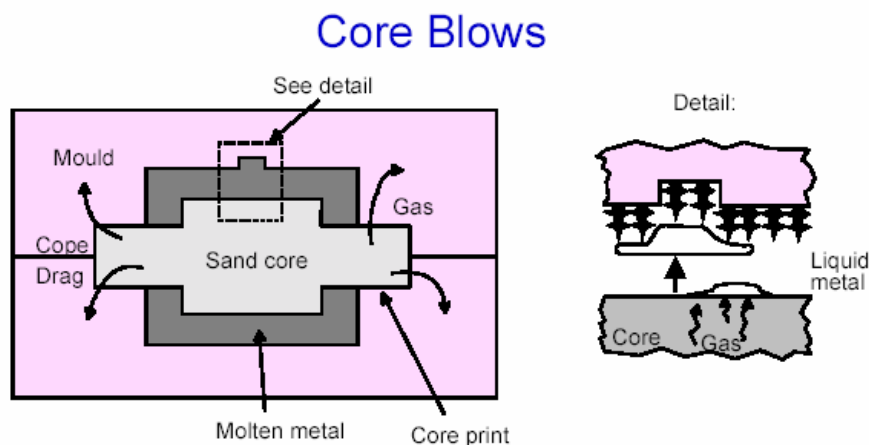
Such 'core blows' are large - typically 10 - 100 mm. The gas pressure in an enclosed core takes some time to build up, so any bubble is released after some freezing has already occurred. Thus core blows are usually trapped under a substantial thickness of solidified skin. If such bubbles are sufficiently large, their top surface will follow the casting contour and their lower surface will be horizontal. They may be located above the core which has caused the blow, but often they are sufficiently large and mobile to migrate to the highest portion of the casting, and can make this region completely hollow. A succession of bubbles from core outgassing will leave bubble trails. This combination of core blow and associated bubble trails constitutes a serious defect which not only mechanically weakens the casting, but also creates a leak path, thus harming a casting destined for an application requiring leak-tightness. The solutions to this problem include:

1. Ensuring that the cores are properly vented, i.e. that there is a means for the gas to escape to the atmosphere.

2. Using sand binders which are low in volatile content and/or which break down slowly.

3. Filling rapidly to a hydrostatic pressure in the liquid metal above that of the pressure of gas in the core, thus suppressing the expansion of gas out into the liquid.

Figure 3207.00.09: The minimum thickness of the bubble in a liquid aluminium alloy when lodged under a horizontal flat surface is usually approximately 12 mm (this corresponds to the thickness of a sessile drop of aluminium sitting on a flat substrate - the bubble can be thought of as a negative sessile drop of negative density!). This dimension is controlled by the ratio of surface tension and density (for grey cast iron the sessile drop and sessile bubble are closer to 7 mm thick). The diameter of the bubble can

of course be any size, depending on the amount of gas released by the core, and is typically 10 to 100 mm.

Size of Core Blow

A core blow can be considered analogous to a sessile drop of molten aluminium
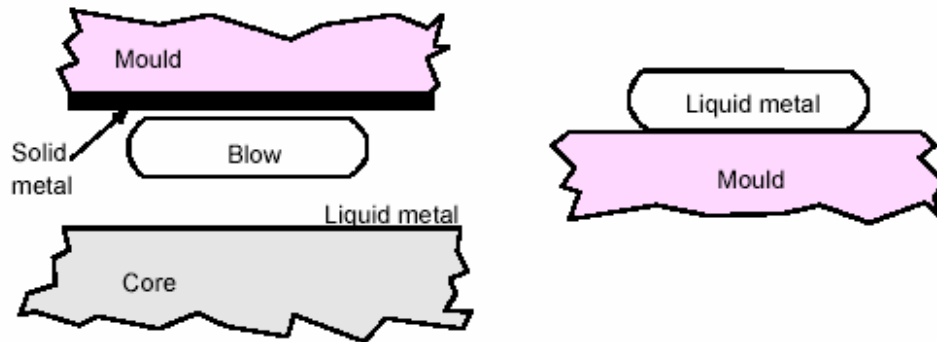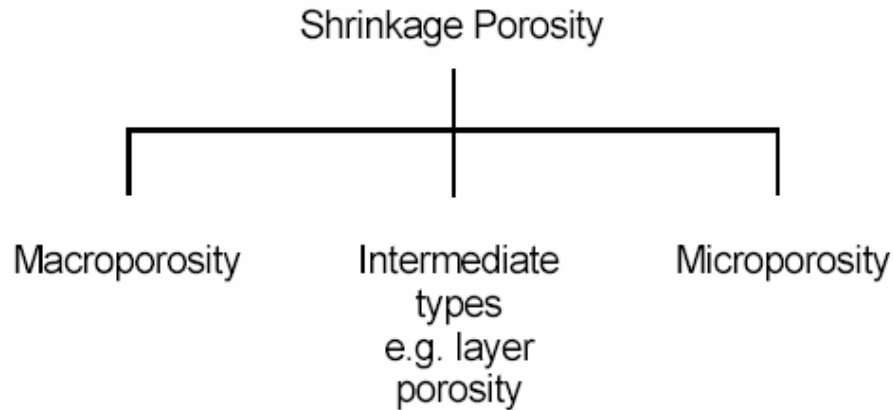
Core blow - height: ~12 mm in aluminium castings (~ 7 mm in grey iron castings)
        - diameter: 10-100 mm (depends on amount of gas released)

Characteristics of Gas Porosity Defects

Defect	Distribution	Size
Gas precipitation from solution	Uniform, apart from 1-2 mm near surface	0.05 - 0.5 mm
Air entrainment	Above ingates, especially the first ingate. Concentrated on horizontal ledges. Very close to surface. Only revealed when casting is shot-blasted or machined.	1 - 5 mm
Core blows	At a uniform distance under top of casting	Typically 100 mm diameter, 10 mm thick

Figure 3207.00.10 summarises the various gas defects and emphasises that they differ significantly in distribution and size. It is most important to bear

such differences in mind when trying to diagnose defects in castings, because, clearly, the remedies will be very different in each case.

Shrinkage Porosity

Macroporosity   Intermediate   Microporosity
                 types
                 e.g. layer
                 porosity

The second type of defect that we need now to consider is **shrinkage** porosity, which is

conventionally sub-divided into macroporosity and microporosity (see Figure 3207.00.11). In reality, there is no fundamental difference between these two forms of porosity - one gradually changes into the other as a function of the freezing range of the alloy. As we will see, it is also possible to identify intermediate types of shrinkage porosity, notably layer porosity in long freezing range alloys.

a) Macroporosity

Formation of Macroporosity (I)
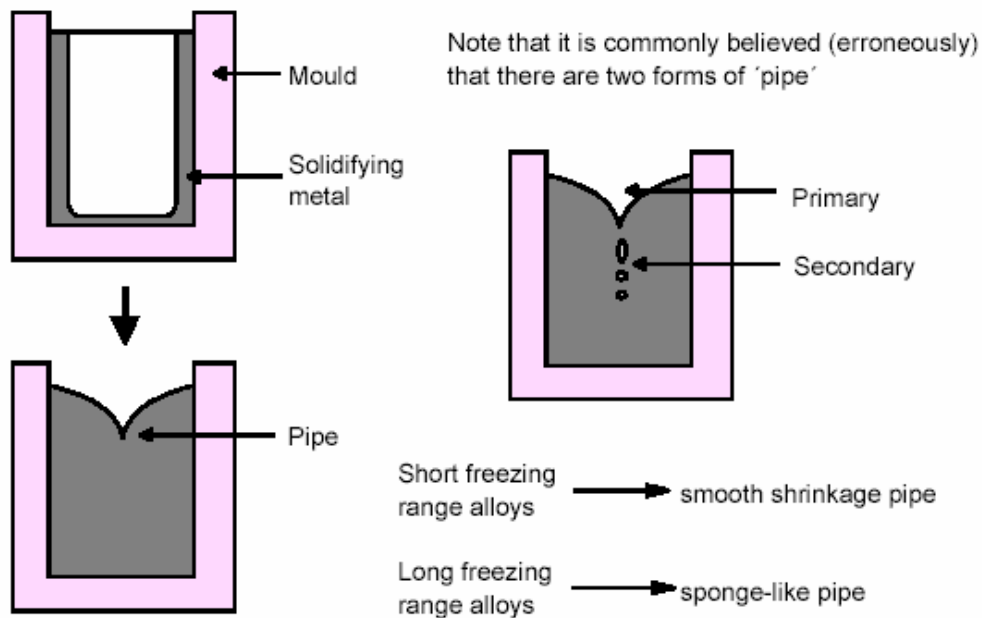
The best known form of macroporosity is the 'pipe' formed as a simple ingot of a short freezing range alloy solidifies (**Figure 3207.00.12**). Solidification starts along the walls and at a slower rate on the top surface. As progressively more metal solidifies, the volumetric contraction is compensated for by the concurrent sinking of the liquid surface, forming a smooth conical funnel or long 'tail' inside the ingot known as shrinkage pipe or piping. It was a commonly-held belief that there are two forms of pipe - primary and secondary - the latter appearing to be discrete islands of porosity below the primary pipe. This is, in fact, an incorrect interpretation of two dimensional sections of such features - the two are interconnected, constituting the same feature, and there is no distinction between them.

It should also be appreciated that there is a difference in appearance only between the shrinkage porosity found in short and long range freezing alloys. In a short freezing range alloy, a shrinkage cavity will take the form of a shrinkage pipe, which can have a mirror-smooth finish (in common with

most of the forms of gas porosity!). In a long freezing range alloy, the shrinkage pipe takes on the character of a sponge, in which the appearance on a polished section is of separate, isolated interdendritic microporosity (i.e. the cuspoid morphology found for all other types of shrinkage). This again is an illusion of the sectioning technique. The defect is actually a macropore which is traversed by a forest of dendrites, and leads to widespread misinterpretation on a transverse section as an array of separate micropores.

Formation of Macroporosity (II)
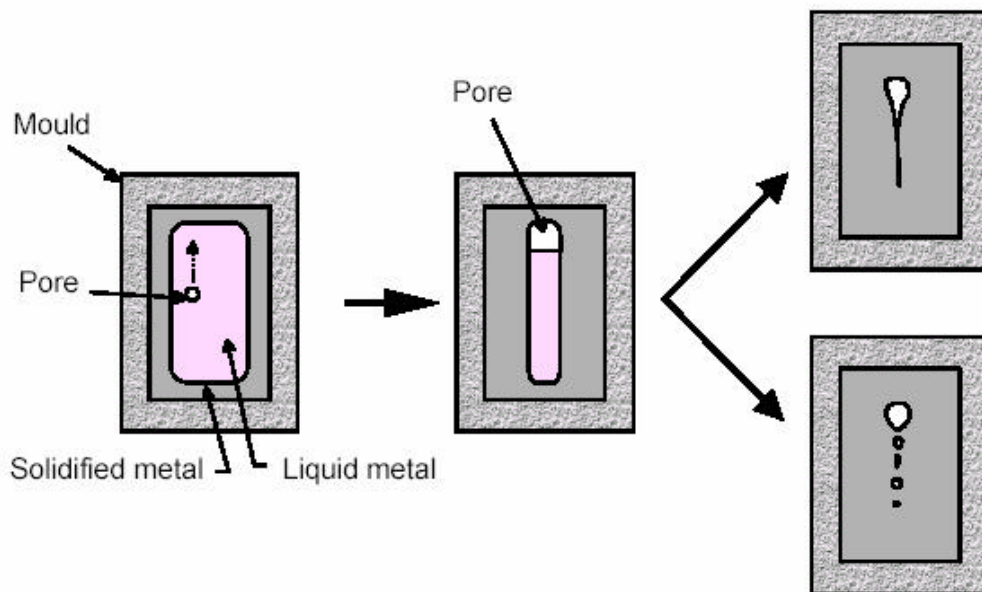
It is also of interest to consider where a single isolated area of macroporosity occurs (see Figure 3207.00.13). It is a common mistake to assume that it will be located in the thermal centre of the isolated region. This is certainly not the case. This shows a totally enclosed ingot solidifying in a mould. A pore could be nucleated anywhere in the entrapped liquid, and will in fact float upwards until it reaches the top of this enclosed volume. The advance of the front at the top of the entrapped liquid volume is locally retarded by the pore

which, when coupled with the geometry of the casting, leads to a long tapering extension to the cavity formed in the casting. If this tubular cavity is not completely straight, it can easily be misinterpreted as being isolated areas of 'secondary pipe' on a cut section. It should also be clear that this simple shape of shrinkage pore reflects the simple shape of the casting. As the shape of the casting becomes increasingly complex, the shrinkage porosity will become correspondingly more complex. So, the characteristic of macroporosity is that it is located towards the centre of a casting, although normally above the thermal centre. It is associated with the geometry of the casting, and usually lies along the centreline of symmetrical castings. As a result, it is also known as centreline porosity, or centreline shrinkage.

b) Microporosity

I would now like to turn to microshrinkage porosity (**Figure 3207.00.14**). This is particularly a problem in long freezing range alloys and/or when the temperature gradient is low. These conditions create an extensive and uniform pasty zone which is favoured by:

. metals of high conductivity, such as aluminium alloys;

. high mould temperatures, as in investment casting;

. thermal conductivity of the mould, as in sand, investment or plaster low moulds.

Microporosity

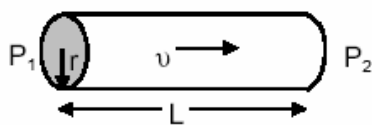**Flow through the pasty zone**

Promoted by:
- Alloys with long freezing range
- Low temperature gradients
  - high metal thermal conductivity;
  - high mould temperature;
  - low mould thermal conductivity.

Feeder
Liquid
Solid
A
Uniform Pasty Zone
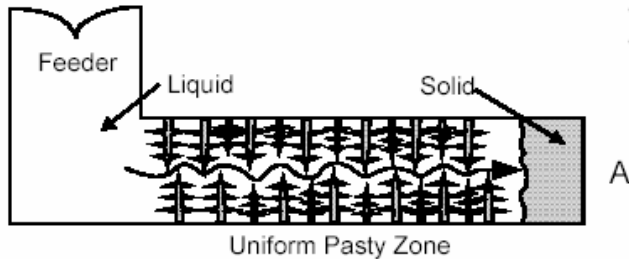
Flow through a capillary (Poisseuille)

P_1 r v P_2 L

$$P_1 - P_2 \propto \frac{L \cdot \eta \cdot v}{r^4}$$

where η is the viscosity.

For the case of a liquid metal flowing through a capillary and simultaneously freezing:

$$\Delta P \propto \frac{\eta \cdot L^2}{r^4}$$

In such cases, towards the end of solidification, there will be a 'pasty' or 'mushy' zone consisting of a forest of dendrites enclosed in the remaining liquid. This shows a simple bar-shaped casting with a feeder at one end. We shall assume that the temperature profile is such that:

1. there is a uniform 'pasty' zone over most of the length of the bar, and

2. solidification starts at the far end (A) and moves towards the feeder.

At a late stage during solidification, liquid metal from the feeder will need to flow through the pasty zone to compensate for the contraction as an increasing amount of solid metal is formed at A. As the contraction tries to pull liquid metal through the pasty zone, it imposes a tensile force upon the liquid. One analogy for this is to imagine that a long elastic rope is being pulled through a forest of trees; it can be imagined that the tortuous route leads to friction along the length of the rope. The elastic rope stretches increasingly towards the direction of the applied pull. This is analogous to

the liquid, which is stretched elastically, and experiences an increasing tensile stress as it progresses through the dendrite forest of the pasty zone. This situation can be analysed in various ways. The easiest approach is to assume that the pasty zone is uniform, and then use the famous equation by Poisseuille which describes the pressure gradient required to cause a liquid to flow along a capillary. This shows a capillary of radius **r** with pressures of P1 and P2 at each end over a distance of L through which a liquid is flowing at a volumetric rate of J per second. Fairly simple calculus can be used to show that:

$$P_1 - P_2 \propto \frac{L \cdot \eta \cdot \upsilon}{r^4}$$

where η is the viscosity. This clearly shows that the resistance to flow is critically dependent on the size of the capillary. In the case of a liquid metal flowing through a capillary, it is simultaneously solidifying and therefore slowly closing the channel. An approximate solution of this situation gives

$$\Delta P \propto \frac{\eta \cdot L^2}{r^4}$$

This shows that the pressure drop by viscous flow through the pasty zone is still very sensitive to the size of the flow channels. L is the length of the pasty zone in the casting and is equal to the whole casting length in most aluminium castings since the thermal conductivity is high and the temperature gradient consequently low. The dendrite arm spacing, DAS, is a measure of the interdendritic flow channel diameter (= 2 x r).
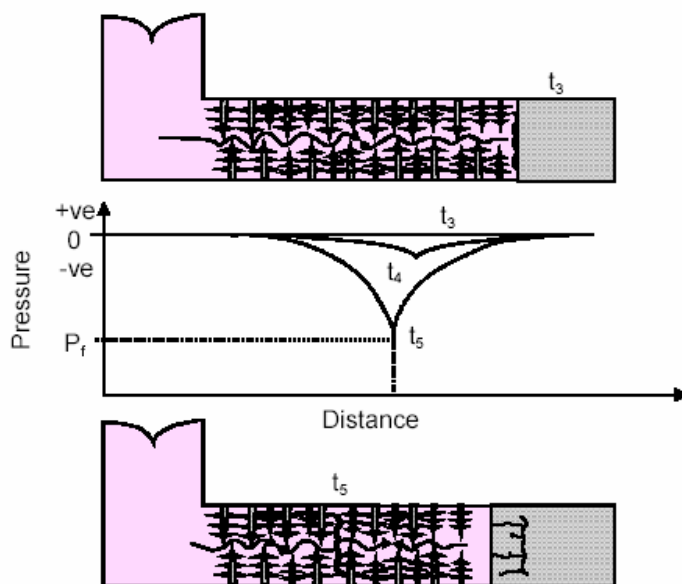
c) Layer Porosity

Figure 3207.00.15 shows a schematic view of the pressure in the liquid metal as a function of time. At time t0, everything is liquid and so there is no pressure gradient. Once solidification starts, the equation that we have just derived shows that there will be a parabolic pressure distribution through the pasty zone. As time progresses from t_1 to t_2, the gradual decrease in the capillary diameter r will increase the tensile stress in the remaining liquid.

The Formation of Layer Porosity (I)
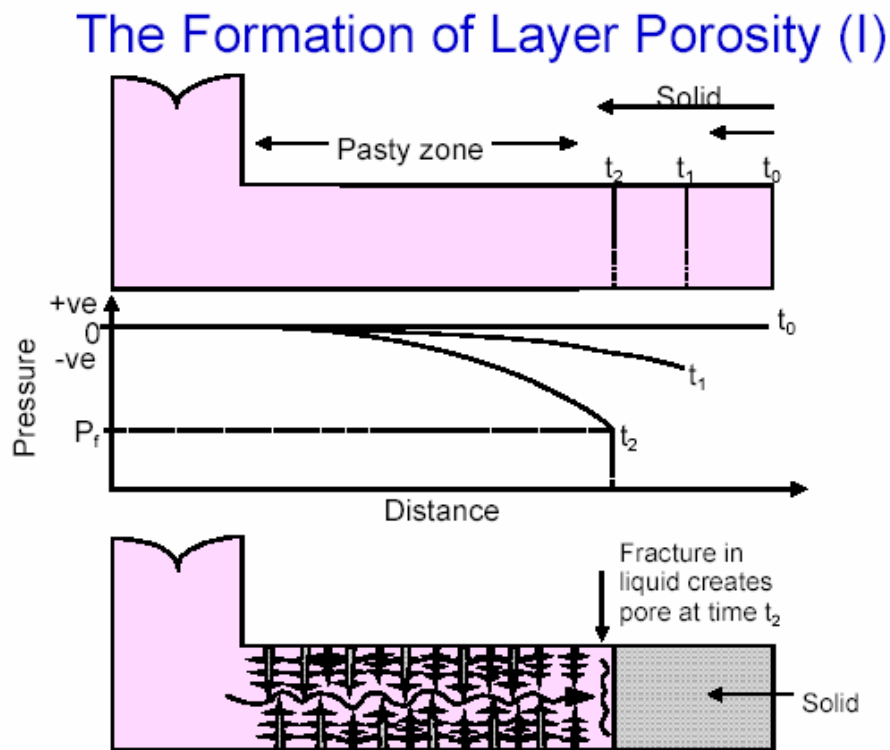
The hydrostatic tension in the liquid in the pasty zone continues to increase until, at time t2, a critical fracture pressure Pf is reached at which a pore will nucleate. This could be an isolated micropore if the shrinkage conditions are not too severe. The generation of small regions of microshrinkage porosity in long freezing range alloys is quite common. Where the shrinkage problem

is more severe however, the development of higher levels of hydrostatic tension in the liquid corresponds to higher levels of elastic energy. Once nucleated, a pore will now spread rapidly along the isopressure surface (which corresponds approximately to the isothermal and isosolid surfaces) relieving the local elastic stress. Note that the liquid moves apart, forming a break in the liquid, not through the dendrites themselves (this is in contrast to a hot tear, as we shall be discussing later). **Figure 3207.00.16**: Immediately that the pore has formed, the pressure in the liquid metal is relieved and drops to zero at time t_3. As solidification proceeds further, the contraction in the middle of the remaining liquid region is fed from both the feeder and by fluid from the surface of the newly created pore. This is a slower growth phase for the pore, extending via channels towards the region requiring feed metal. Hence a parabolic pressure gradient develops from both ends (time t_4) and after yet further solidification has taken place, the pressure in the liquid will again reach the critical level of P_f at time t_5. A further pore will then nucleate and grow.
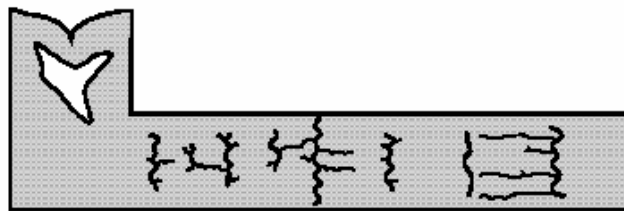

The Formation of Layer Porosity (II)

The Formation of Layer Porosity (III)

Final condition of casting

This process repeats until the whole of the casting has solidified (**Figure 3207.00.17**). Porosity therefore occurs in a sprinkled array of fine pores, or, if the shrinkage conditions are more severe, in an array of layers, as shown schematically here in the final casting. It should be reiterated that both microshrinkage porosity, and its more severe form of layer porosity, nucleate in the liquid and grow through the liquid without disturbing the dendrite mesh. The dendrites tend to bridge a region of layer porosity, effectively sewing it together with closely-spaced threads. As a result, layer porosity does not have a dramatic effect on tensile strength, and it is often quite difficult to see on a single section under the optical microscope; it has the appearance of scattered isolated pores - the joining of the pores over a large plane is not evident.

Radiographs of Nickel-Base Alloy Showing Different Forms of Shrinkage Porosity

Radiographs of 100 x 30 x 5 mm bars cast in nickel-base alloy at 1620°C in vacuum (15 μm Hg) and with mould temperatures of :

(a) 250°C

(c) 800°C

(b) 500°C

(d) 1000°C
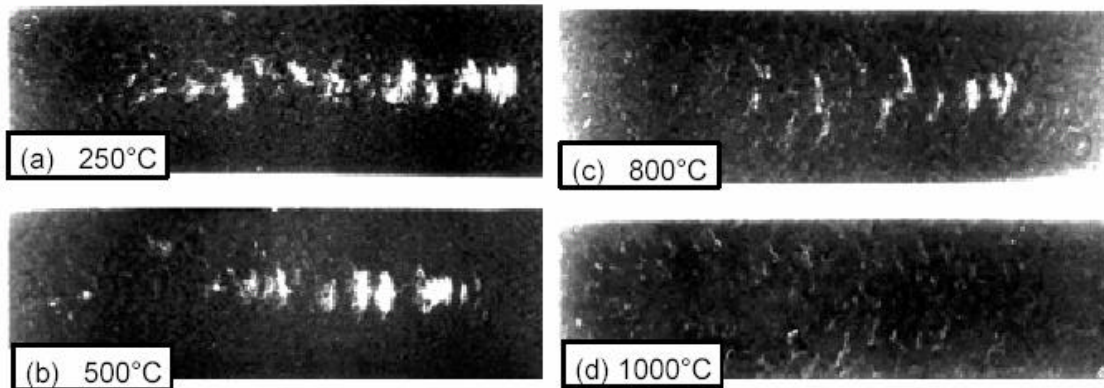
In contrast, as shown in **Figure 3207.00.18**, layer porosity can be seen quite clearly in X-rays, as long as the radiation is oriented along the plane of the porosity. This is because, of course, the radiograph effectively integrates the porosity over the depth of the section. It is also clear from this radiograph that centreline shrinkage, layer porosity, and dispersed microshrinkage are all varieties of shrinkage porosity which grade imperceptibly from one to another. In this case the conditions are changed by changing the mould temperature to move from a condition of steep temperature gradients, through shallow temperature gradients, to uniform freezing of the pasty zone. However, a similar effect can be achieved by the changing of alloys to move from short freezing range, through medium, to long. A further way in which porosity grades imperceptibly between different types is the gradual substitution of dispersed gas porosity (a kind of dispersed microporosity) for

layer porosity, or even for centreline shrinkage, as the gas content of the melt is gradually increased.

Sources of Porosity

Shrinkage Porosity

Scale	Macroporosity	Intermediate Types	Microporosity
Cause	Failure of liquid feeding; i.e. 7 Feeding Rules not correctly applied	Failure of interdendritic feeding	Failure of interdendritic feeding
Short ΔT_f	Smooth shrinkage pipe	Centreline shrinkage	Dispersed micro-shrinkage
Long ΔT_f	Shrinkage sponge	Layer porosity	Dispersed micro-shrinkage

Gas Porosity

Scale	10 - 100 mm	1 - 5 mm	0.05 - 0.5 mm
Cause	Core blows	Air entrainment	Hydrogen precipitation
Short ΔT_f	No direct effect of freezing range	No direct effect of freezing range	Round micropores
Long ΔT_f	No direct effect of freezing range	No direct effect of freezing range	Interdentritic appearance of micropores

Microporosity in castings can clearly arise therefore from a variety of sources (see also **Figure 3207.00.19** and **Figure 3207.00.20**):

1. Firstly, sponge porosity is commonly mistakenly identified as microporosity. It is, of course, actually a macroshrinkage pipe resulting from insufficient liquid available at a late stage of freezing of a long freezing range alloy.

2. In the case of correctly identified microporosity, it still has a number of sources and can take a number of forms:

(a) Isolated and evenly distributed gas pores as a result of hydrogen precipitation.

(b) Isolated, but somewhat less evenly distributed, microshrinkage porosity.

(c) Centreline shrinkage porosity.

(d) Layer porosity.

The three latter types are true types of shrinkage porosity. However, as the feeding problem is progressively better addressed the last two varieties become less well developed, the distinctions gradually blurring, the shrinkage porosity becoming only a scattering of isolated microscopic pores, as variety (b) above. With very good feeding, of course, all varieties finally disappear altogether. Similarly, as we have already mentioned, as the gas content gradually increases, centerline and layer porosity gradually blur into dispersed gas porosity. Thus it is necessary to take care when categorising any microporosity; it is often the result of a number of factors involving contributions from gas and shrinkage, i.e. poor degassing and poor feeding practice. There is rarely a single cause.

Characteristics of Hot Tears

- Ragged, branching crack
- Generally intergranular
- Dendritic morphology on failure surface
- Heavily oxidised failure surface
- Often located at hot spot
- Random occurrence and extent
- Alloy specific

I would now like to turn to hot tears which are one of the most serious defects that can occur in castings. These have a number of characteristics, as shown here (see also **Figure 3207.00.21**):

. The form is that of a ragged, branching crack.

. The main and subsidiary branches generally follow an intergranular form.

. The failure surface has a dendritic morphology and is heavily oxidised prior

to any subsequent heat treatment.

. Tears are often located at hot spots.

. They can occur randomly and their extent is also variable under apparently identical casting conditions.

. Tears occur readily in some alloys, whereas others are virtually free from this problem.

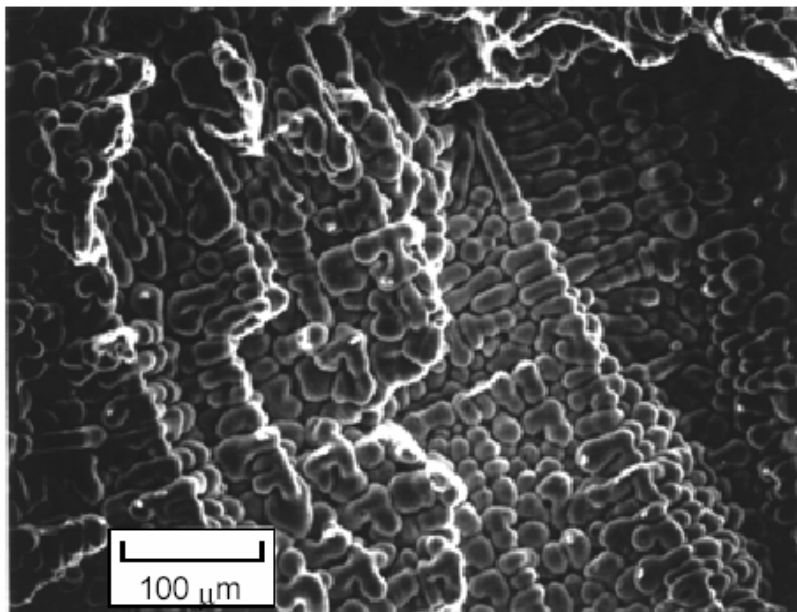SEM view of the surface of a hot tear in an Al-7Si-0.5Mg alloy sand casting

100 μm

Figure 3207.00.22 shows the surface of a hot tear in an Al-7Si-0.5Mg sand casting as viewed by a scanning electron microscope. The dendritic morphology is clearly revealed. Hot tears occur in the late stages of the pasty condition of freezing, when only a per cent or so of residual liquid remains between grains.

Hot Tearing Model

Hot Tearing Model

1. Initial stage: hexagonal grains surrounded by liquid film.

2. Application of tensile strain leads to grain impingement and the creation of intergranular pools.

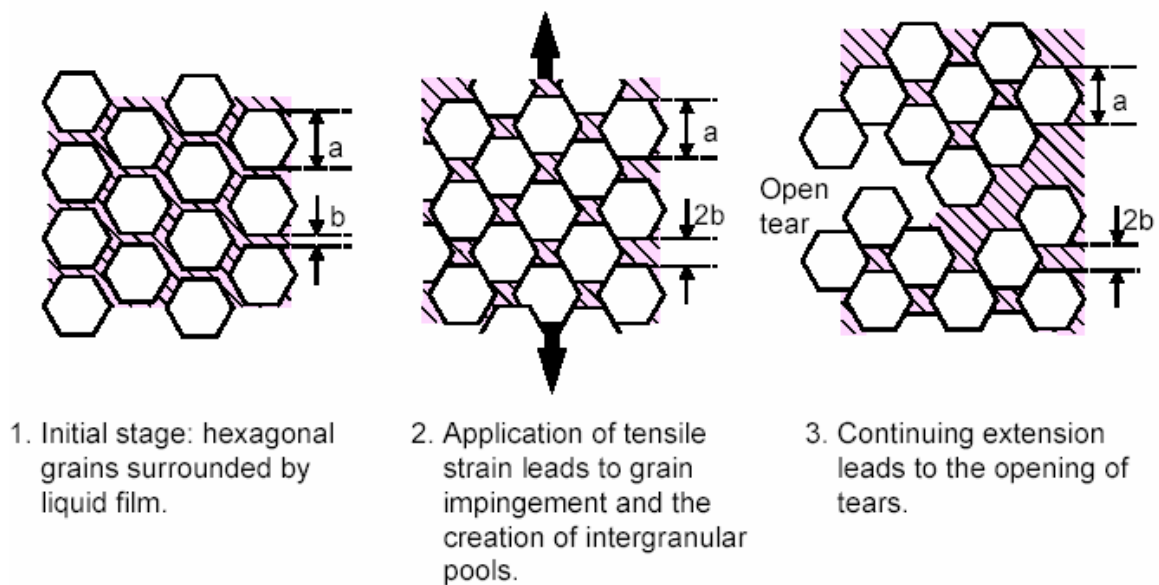3. Continuing extension leads to the opening of tears.

Figure 3207.00.23 shows a simple model of hexagonal grains of diameter a separated by a liquid film which initially has a thickness of b. At this stage, if the mixture is subjected to a tensile strain, usually as a result of the linear contraction of the casting as it cools, the grains will tend to move further apart in the longitudinal direction but come closer together in the transverse direction. At first this separation is a fairly uniform process distributed over many grains, but later becomes concentrated in one or more branching planes. This causes the residual liquid to rearrange itself somewhat and will effectively create segregation defects in the form of layers of a solute-rich low melting point constituent in

the casting. However, such features have not so far been shown to be important in their effect on mechanical or other properties. At this stage no substantial defect is caused and the grains will still be nicely cemented together by the liquid phase after freezing. If strain continues to be applied to the solidifying casting, then the ability of he residual liquid to rearrange itself to fill the volumes left by the rearranging grains is now used up. At this point the further separation of the grains draws air into the space, forming a defect known as a hot tear. The hot tear is so-called because it forms at particularly high temperatures, in the solid/liquid regime. Also, its ragged form, often with branching tributaries, is nicely described as a tear. It contrasts with tensile failures in the solid state such as quench cracks formed on quenching castings into water following solution treatment, in which the failure is a fairly straight, smooth, narrow crack.

Hot Tear Examples

This rather simple model does in fact correspond well with features seen in real castings. This shows a radiograph (**Figure 3207.00.24**) of a hot tear in an Al-6.6%Cu grain refined alloy. The dark areas are copper-rich eutectic which are the segregated areas formed in Stage 2 of the model. The white areas are the open tears.

Radiograph of a hot tear in an Al-6.6Cu grain-refined alloy
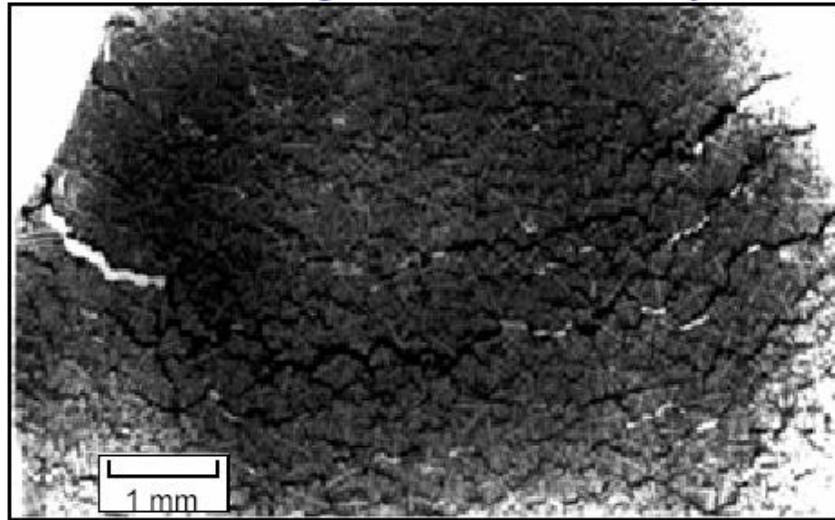
Dark areas are copper-rich eutectic.
White areas are open tears.

It can be noted that grains can separate but remain connected by residual liquid. This shows an example of a filled hot tear in an Al-10%Cu alloy cast at 250°C superheat and which was not grain refined (see **Figure 3207.00.25**). A liquid-filled hot tear like this is often called a 'healed tear'. This is a misconception since the term 'healed tear' suggests that the tear formed in an open state, and was subsequently 'healed' by the inflow of liquid. This is not so. The liquid-filled state is the original state of the tear, which only develops into the tear if the liquid drains clear for some reason, or if the tear opens further, exhausting the ability of the residual liquid to continue to fill it.
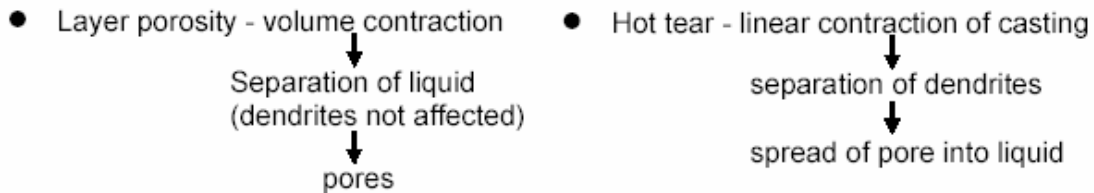
A filled hot tear in an Al-10Cu alloy, not grain refined, cast with 250°C superheat

It can also be noted that the hot tear defect is quite different to that of layer porosity (see also **Figure 3207.00.26**). In layer porosity, it is the volume contraction on solidification which drives the liquid to separate, leaving the dendrite mesh unaffected, and in place, effectively linking the sheet of pores at many points across its surface. In contrast, the hot tear forms as a consequence of the driving force of the linear contraction of the casting as it cools, which separates the dendrites first, and leads on to the spread of a pore into the liquid later. The percentage of residual liquid in the solidifying casting is critical to the development of hot tears. For those alloys in which there are large quantities of a residual eutectic, the ability of the casting to contract without danger of exhausting the supply of liquid means that such alloys are not easily susceptible to hot tearing. Such alloys include the Al-Si family and their hot tearing resistance explains the popularity of these alloys as casting alloys. In contrast, the Al-Cu family of alloys, although strong, are

subject to severe hot tearing problems as a result of the character of solidification; only a small amount of

residual liquid surrounds the grains for a relatively long period at a late stage of freezing.

Comments on Hot Tearing

1. Do not confuse hot tears with layer porosity!

- Layer porosity - volume contraction
  ↓
  Separation of liquid
  (dendrites not affected)
  ↓
  pores

- Hot tear - linear contraction of casting
  ↓
  separation of dendrites
  ↓
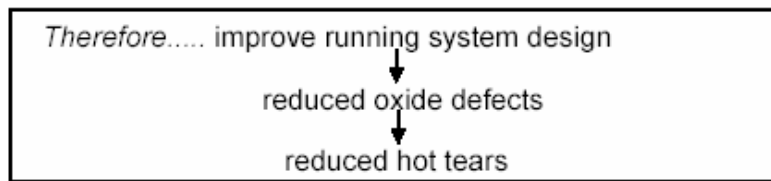 spread of pore into liquid

2. Increasing residual liquid in solidifying casting reduces hot tears!
   e.g. Al-Si - good hot tearing resistance
   Al-Cu - poor hot tearing resistance

3. Hot tears require nucleation!
   e.g. oxide films in aluminium alloys.

Therefore..... improve running system design
↓
reduced oxide defects
↓
reduced hot tears

One final aspect of hot tearing is worth emphasising. Hot tears usually have to be nucleated. If no suitable nucleus is present, then it is difficult, or impossible in some alloys, to form a tear. Since it seems that, in aluminium alloys, oxide films are excellent nuclei for hot tears (and excellent for nucleating other volume defects such as the various forms of porosity) then it follows that improving the design of the filling system and increasing the quality of the liquid metal will often cause hot tears to disappear. This non-traditional technique is recommended as the most effective method of all to deal with hot tears.

Prevention of Hot Tears:

- Alter casting design
- Chill hot spots
- Reduce constraint from mould
- Add brackets and webs
- Grain refinement
- Reduce casting temperature
- Alloying
- Reduce contracting length

There are various 'traditional' techniques for dealing with hot tearing (see also **Figure 3207.00.27**):

. It may be possible to alter the geometry of the casting to reduce stress concentrations and hot spots, for example, by providing generous radii at vulnerable sections.

. Local hot spots can be reduced by local chilling which will strengthen the metal by taking it out of the susceptible temperature range.

. There are various ways of reducing the mould strength so that it provides less constraint to the contracting casting.

. Brackets and webs can be placed across a vulnerable corner or hot spot to provide mechanical support and to enhance local cooling.

. Grain refinement should help to reduce tear initiation since the strain will be spread over a greater number of grain boundaries.

. A reduction in the casting temperature can sometimes help, probably because it reduces the grain size.

. It is sometimes possible to benefit from varying the alloy constituents within the specified composition ranges. In particular, increasing the volume

fraction of eutectic liquid may help by increasing the pre-tear extension and by decreasing the cracking susceptibility.

. Finally, it is sometimes possible to site feeders carefully so that the casting is effectively split up into a series of short lengths to reduce the strain concentration.

Cold Cracks

Cold Cracks

- ❑ Form at temperatures below the solidus
- ❑ Straighter and smoother than hot tears
- ❑ Transgranular or intergranular
- ❑ Can be oxide-free (if formed at low temperature)

- ❑ Sources of stress:
  - differential cooling
  - mould/core restraint
  - phase transformation
  - heat treatment

- ❑ Prevention:
  - reduce stress raisers
    - avoid abrupt changes of section
    - eliminate oxide defects
  - reduce mould/core restraint
  - eliminate or use alternative heat treatments

The final type of defect that we will consider are **cold cracks** (**Figure 3207.00.28**). By definition, these form at some temperature below the solidus and often considerably higher than room temperature. Whereas hot tears tend to be rather ragged in nature, cold cracks are straighter and smoother. Cold cracks can be either transgranular or intergranular, depending on the relative strengths of the grains and the grain boundaries. Whereas hot tears are always oxidised, cold cracks can be free from oxidation if they are formed at relatively low temperature and the castings not subsequently heat treated. Stress is required for a crack to nucleate and

grow and there are a number of possible sources. Firstly, the casting continues to suffer strains as it cools to room temperature. This arises simply from the differential cooling rates of different parts of the asting. In addition, of course, the mould or cores may resist the contraction of the casting to such a degree that dangerously high stress is generated within the casting. Stresses can also arise from the volumetric changes associated with phase transformation(s) as the casting cools to room temperature, one example being the transformation from delta ferrite to austenite in steels. However, a greater danger from stress during cooling occurs later, if the casting is subject to solution or homogenisation heat treatment. In the case of aluminium castings, the quench from the solution treatment temperature causes the most severe stress to which the casting is subjected, and tensile failure may occur during the quench. This is because the strain on cooling from near the melting point is approximately 1.3 % for aluminium. (This is of course the pattern-making contraction for a freely contracting casting.) A contraction strain of this magnitude is more than ten times greater than the strain to produce plastic yielding in the casting. Thus the corresponding stress comfortably exceeds the initial yield point, and is not far from its ultimate tensile strength. On occasions the tensile strength of the alloy is exceeded during the quench, and the casting fails by cracking.

Since the residual stress is hardly relieved at all by normal ageing, most heat treated castings are put into service with dangerously high internal stresses. Many in-service failures by cracking can be attributed to residual internal stress, although this is often overlooked by metallurgists looking for more obvious external symptoms!

Some of the ways of preventing cold cracking should be obvious. Firstly, any action to reduce stress concentrations should be beneficial. For example, abrupt changes in section should be avoided by re-designing the casting or

by replacing sharp angles with generous radii. Internal defects such as oxide films in aluminium castings can act as nuclei for cold cracking. In addition, if the defect is in fact a *folded* film this is, in effect, a crack in the casting which will simply propagate when a stress is applied. A well designed running system should reduce or eliminate oxides and, in turn, have a beneficial effect on cracking. Secondly, it is sometimes possible to select alternative mould and core materials to reduce constraint, so long as this is not at the expense of, for example, reduced dimensional stability or a poorer surface finish. It is also recommended that serious consideration be given to eliminating solution treatment and quenching if at all possible, choosing some alternative treatment for the attainment of strength. **Figure 3207.00.29**: If quenching cannot be avoided, then it requires to be controlled. Hot water quenching is hardly any improvement over cold water quenching. The best combination of rate and safety is afforded by polymer quenchants. There is some evidence that polymer quenching may lead to more reproducible mechanical properties, as shown by these results for a sand-cast Al-7%Si-0.5%Mg alloy. However, often forced air cooling is sufficiently rapid to retain a significant amount of the value of the quench, but is sufficiently even to reduce the stress build up almost to zero. Mechanical properties are, of course, not so high when air quenching is used. However, the reduction in internal stress means that the casting actually performs more reliably in service.

Reduction in Quenching Stresses

Rates of cooling of a 20 mm diameter aluminium bar when quenched by various means from 500°C
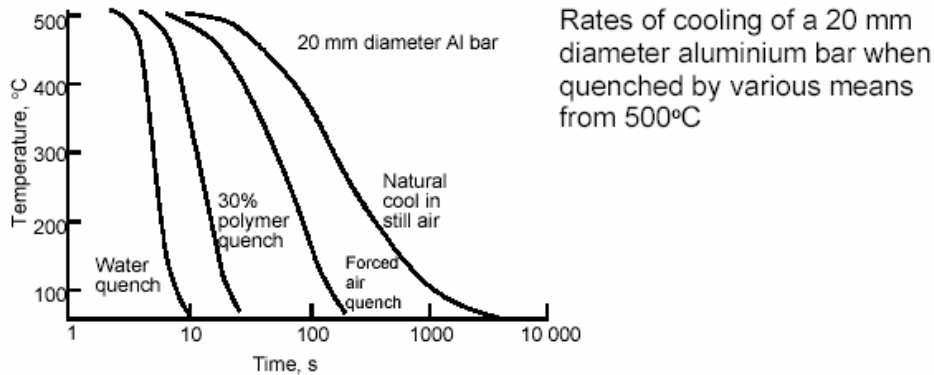
Effect of quenching medium on ductility

Quenching Medium	Elongation, %	
	Mean +/- 2.5 σ	Minimum
Hot water (70°C)	4.73 +/- 2.72	2.01
Cold water	6.47 +/- 1.67	4.80
Water-glycol mixture	5.81 +/- 0.96	4.85

Conclusions

In conclusion, castings unfortunately can contain defects which may render them unsuitable for service, resulting in higher costs and/or lower profits for the production foundry and delivery delays to the customer. Some defects may not always be found prior to service - in fact, some *cannot* be found using normal non-destructive techniques - and there is always a danger that service stressing may cause a pre-existing defect to propagate, leading to premature failure. This lecture has provided an introduction to the nature and origin of major solidification defects in castings. Emphasis has been placed on how such defects can be diagnosed correctly and, more importantly, eliminated from the outset by using correct foundry techniques. Quality-conscious foundries are increasingly recognising that all parts of the

production process must be properly controlled if the industry is to maintain its position of being a leading supplier of metallic components.