HOT TEARING PREDICTIONS IN DIRECT CHILL CAST ALUMINUM AA5182 INGOTS

by

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Abstract

Hot tearing is a major industrial problem because it affects both the quality and productivity of the casting process. These defects form in the last stage of the solidification sequence, in a region where the grains are surrounded by a continuous film of liquid and so they cannot sustain tensile strain. The application of strain can cause the metal to fail in the region between the dendrites, creating a hot tear.

Previously, many researchers have focused their attention on hot tearing in cylindrical billets. With the development of a coupled thermal – stress finite element model of the aluminum Direct Chill (DC) ingot casting process at UBC, both hot tearing susceptibility and criteria validation can now be investigated in ingots of rectangular cross-section using the temperature, stress, and strain fields predicted by this process model. Two hot tearing criteria have been implemented into the DC ingot casting model: Pellini’s Total Strain criterion, in which the strain accumulated during solidification is compared to a critical value, and the RDG Hot Tearing criterion, in which the magnitude of the pressure drop in the liquid due to volumetric shrinkage and mechanical loading is an indication of hot tearing susceptibility.

To investigate the hot tearing predictions, two castings have been simulated - a non typical hot cast, and a non typical cold cast. It is the cold cast which is prone to hot tearing. The strain fields in the cold cast clearly show that there is a high accumulation of tensile strain during solidification on the rolling face, just above the ingot lip. This is also the region where hot tears are observed industrially.

There is good agreement between the hot tearing predictions made by the Total Strain criterion, and industry observations. Using the strain predictions from the cold cast simulation, this criterion predicts a region prone to hot tearing in the middle third of the rolling face, in the start-up phase. In this critical region, the plastic strain accumulated during solidification exceeds the ductility limit at a temperature of about 575°C. Both further up the ingot, and out towards the edge of the rolling face, the plastic strain accumulated during solidification was less than the ductility limit. The conditions in the hot cast were such that the predicted plastic strain during solidification on the rolling face was also always less than the ductility limit. In contrast, there is poor agreement between the hot tearing predictions made by the RDG criterion and industry observations. In both the hot and cold cast simulations, the region predicted to be most susceptible to hot tearing is the steady-state phase, where hot tearing was not observed to occur.
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Chapter 1: Introduction

1.1 Aluminum Alloys: An Overview

Often forgotten by many is that the development of society has occurred not only through war or ideology change but through scientific discovery, specifically advancements in materials science. This is readily apparent as those in the social sciences tend to name important eras after new material discoveries. Both the Bronze Age and the Iron Age denote eras when the most important aspect of the time period was the ability to refine these materials from their respective ores. One metal and alloys system extremely important for the current age is aluminum. Because of its unique properties, especially relating to strength and weight, aluminum has found uses in areas encompassing beverage containers to airplanes and space exploration. No other metal has proved to be as versatile as aluminum during the age of industrialization[1].

1.1.1 Properties of Aluminum Alloys

The progress of aluminum from a rare and exotic metal in the mid 19th century to today's generic lightweight metal has been phenomenal. World aluminum usage has grown to over 20 million tonnes per year since industrial production began in the late 1880's and annual volumetric production now exceeds all other metals combined, with the exception of steel[1]. Aluminum's phenomenal growth can be attributed to its surprising versatility, achieved through the wide range of physical and mechanical properties that can be created from within its alloy family. These properties include:

- Weight, as aluminum has a density less than one third of steel
- Corrosion resistance to various environments including food
- High Strength
- Good workability, and formability
- Heat treatable, creating products with a wide range of ductility and strength.

This broad range of properties means that aluminum has found use in most areas of manufacturing, including transportation, construction and packaging[2].

Aluminum alloys can be divided into two major categories: casting alloys and wrought alloys. Wrought alloys are generally used for further fabrication; such as rolling or drawing, while casting alloys are used to form cast parts and have compositions to increase the flow
characteristics of the liquid melt. Each category can be further subdivided based on the mechanism of property development. Some alloys are heat treatable, based on second phase solubility. The process of solution heat-treating, quenching and precipitation will dramatically alter the microstructure and thus the mechanical properties of these types of alloys. Other alloys may also undergo work hardening during mechanical reduction to increase the strength of the finished product. Most useful aluminum alloys are highly alloyed as the secondary elements play an extremely important role in enhancing the material's final mechanical properties.

1.1.2 Production of Aluminum Alloy Sheet Products

Among the common metals, industrial production of aluminum is a relatively new process, as it began only in the late 19th century when Charles Hall (USA) and Paul Héroult (France) independently discovered the electrolytic reduction reaction of alumina dissolved in molten cryolite in 1886. Essentially the same process is still used today at all aluminum smelters in the world to extract the metal from its oxide. Once refined, the aluminum, now in a liquid state, is solidified (a process referred to as casting) and then sent down the production path to form a semi finished product as per Figure 1.1.

Figure 1.1 - Operations in Semi fabricated Products Plants.
One of the major semi fabricated products is aluminum sheet, used in the manufacture of a large number of products, such as beverage containers, airplane parts and furniture. Production of aluminum sheet involves a number of manufacturing steps including casting and homogenization to form an ingot, hot and cold rolling to the desired thickness and then heat treating to ensure that the correct mechanical properties are met. Of these processes, the casting step is particularly challenging because solidification is a highly complex process and defects within the ingot, such as porosity, surface roughness, and cracks, will greatly affect the productivity and quality of the downstream manufacturing processes.

1.2 Direct Chill (DC) Casting Process for Aluminum Sheet Ingots

1.2.1 Overview of the Direct Chill Casting Process

For the past fifty to seventy-five years, the semi-continuous Direct Chill (DC) casting process has been used extensively in North America for the industrial production of aluminum ingots and billets. This casting technology has become the industry standard practice because of its relative simplicity: the process consists of an open, water-cooled mold, a vertically adjustable bottom block and cooling water, as shown in Figure 1.2. At the beginning of the process, the bottom of the mold is plugged by the bottom block. Superheated liquid is poured into the mold at a predetermined filling rate and begins to cool as the bottom block and mold conduct heat away from the liquid metal. Once the liquid reaches a certain height within the mold, the bottom block and cast ingot exit the base of the mold and are lowered into the casting pit. Superheated metal is continually added to the top of the mold so that casting can continue until the ingot has reached a predetermined length. During the casting process, the portion of the ingot protruding from the bottom of the mold is cooled by water sprays impinging on the ingot surface immediately below the mold. The water then continues to flow down the face of the ingot, where it drains into the bottom of the casting pit. The source of the water is a series of holes drilled into the base of the mold. Once the ingot has reached a predetermined length, typically between four and ten metres, the process is stopped and the ingot cooled. It is then removed from the casting pit, the bottom block is repositioned in the mold and the process is restarted.
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Figure 1.2 - Schematic of the aluminum ingot Direct Chill casting process\(^\text{[3]}\).

The casting process can be divided into three distinct stages\(^\text{[3],[4]}\): start-up phase, steady-state phase and end phase. In the start-up phase, the liquid pool profile and thermal fields evolve over time as the mold and bottom block heat up. During the steady state phase, the base of the ingot is slowly lowered into the casting pit but the thermal fields remain relatively constant with respect to the spatial position of the mold. In the end phase, casting has stopped so the thermal fields again change relative to the mold as the ingot is allowed to cool.

1.2.2 Quality Issues in DC Cast Aluminum Sheet Ingots

Although the aluminum DC casting process has been successfully employed since the 1930's, significant quality issues still remain in producing defect-free ingots consistently. As reported by Grandfield and McGlade\(^\text{[5]}\), dimensional control, hot tearing and cold cracking are the major sources of process-related defects in ingot casting. Research has shown that these defects can be attributed directly to the cooling conditions during the start-up phase of the process\(^\text{[6]}\). A schematic diagram showing the location of these defects is shown in Figure 1.3.
Initially, while the ingot surface is contained within the mold, it is cooled relatively slowly. Immediately after the ingot exits the bottom of the mold, under certain process conditions, the water sprays rapidly cool the surface of the ingot such that the surface temperature is lowered below about 150°C, while temperatures near the centre will still be above the solidus temperature. These large thermal gradients will induce thermal strains (and stresses) within the ingot, causing the base of the ingot to bow and pull away from the surface of the bottom block. This base deformation is known as butt curl. The butt curl begins to evolve when the secondary cooling water first hits the ingot surface during the start-up phase, causing the base of the ingot to bow upward and the vertical sides to be displaced inward. Once the cast length has reached between 0.5 and 0.75 m, the rate of butt curl rapidly diminishes as the ingot enters the steady-state cooling phase. Unfortunately, the macro deformation remains, and often causes the butt of the ingot to be outside the acceptable dimensional tolerances. This material must be removed, by cutting off this end of the ingot, before downstream processing can occur. Mold, bottom block design, and optimized cooling conditions have been found to reduce the amount of deformation linked to butt curl[3].

Large amounts of thermal strain can also lead to the formation of cracks. If they develop above the solidus temperature, the cracks are referred to as being 'hot tears'. If they develop
below the solidus, the cracks are referred to as being 'cold cracks'. Hot tears form when the partially solidified material is subjected to a tensile stress and there is insufficient liquid feeding to fill any newly formed gaps\textsuperscript{[7]}. This defect has been linked to high casting speeds, bottom block design, thermal gradients, alloy chemistry, and variability in cooling conditions during the start-up phase. Cold cracks form below the solidus temperature with damage thought to propagate both between and through the solidified grains. These cracks form in the typical manner either through high temperature ductile yielding or fracture\textsuperscript{[8]}. It has been found that reducing the casting velocity decreases the risk of hot tear formation because it directly governs the strain applied to the mushy zone, and the liquid pressure drop\textsuperscript{[9]}. Unfortunately, low casting speeds will increase the likelihood of cold cracks to form, and also reduce productivity.

One of the challenges to improving product quality in the DC casting process is that there are only a limited number of parameters within the process that will affect the stress – strain and cooling conditions. These variables include the geometry of the bottom block, the metal pour temperature and flow rate, the ingot withdrawal rate (casting speed), and the spray water intensity. In addition to water flow rate, proprietary cooling systems have been developed to control the ability of the spray water that cools the surface of the ingot, such as Alcoa's CO\textsubscript{2} injection system, Wagstaff's Turbo process, and Alcan's Pulse Water technique. The casting recipe refers to the variables that can be adjusted during the DC casting process: metal pour temperature and flow rate, cooling water temperature and flow rate, and casting speed. For a given bottom block geometry and spray water conditioning system, the casting recipe is chosen to limit the amount of hot tearing, cold cracking and butt curl, while maximizing volume throughput\textsuperscript{[6]}. The casting recipe has evolved largely through trial and error and is dependent on the size of the ingot, and alloy type. The geometry of the bottom block is known to have a large impact on quality, especially the formation of butt curl. Although a few attempts have been made to develop an optimized bottom block shape empirically, further work is required\textsuperscript{[3]}. 

1.2.3 Heat Transfer During Direct Chill Casting

There are three major cooling zones within the Direct Chill casting process, as per Figure 1.4\(^3\). The first regime, known as Primary Cooling, represents heat transfer between the ingot and the water-cooled mold. By the end of this cooling zone, a solid shell must be formed capable of supporting the liquid interior as the ingot is exiting the base of the mold. Heat transfer between the cast ingot and the spray of water below the mold is known as Secondary Cooling. During this stage, the cross section of the ingot solidifies and temperatures fall to a relatively low value. This region is characterized by an impingement zone, where the spray water first contacts the ingot, and a free-falling zone, where the spray water streams down the ingot surface below the impingement zone. In the impingement zone, the horizontal momentum of the water is quite high as it has just exited from the base of the mold. This causes the water to have enhanced heat transfer characteristics. In the free falling zone, the spray water has lost all horizontal momentum and tends to have reduced heat transfer characteristics. The third regime is known as Base Cooling, and represents heat transfer between the cast ingot and the bottom block. Heat transfer in this region is important in terms of the ingot quality, as good control of Base Cooling

![Figure 1.4 - Cooling Modes during DC Casting\(^4\).](image-url)
can reduce the amount of butt curl. In steady state casting conditions, about 80% of the heat is removed during Secondary Cooling and about 20% is removed during Primary Cooling\textsuperscript{[16]}. Base Cooling is a major component of the heat transfer only during the start-up phase\textsuperscript{[3]}.

There are two important phenomena that greatly affect heat transfer during the start-up phase: water ejection, and water incursion\textsuperscript{[2]}. Depending on the process parameters chosen, the surface temperature of the ingot as it exits the mold can be quite hot. If the surface temperature is too high, the spray water in contact with the ingot surface will boil. This will lead to the formation of a stable vapour film layer that can cause water to be ejected from the ingot surface at points in the free falling zone. The result is that contact between the water and the ingot is significantly reduced, decreasing heat transfer. Water incursion occurs if there is significant base distortion, or butt curl. This allows water to enter the gap formed between the bottom block and the base of the ingot, and greatly affects Base Cooling kinetics.
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1.3 Overview of the Hot Tearing Defect

It is generally believed that hot tears in aluminum DC casting start to develop at temperatures within the mushy zone, when the fraction solid is close to one\(^{11}\). Hot tearing is a major industrial problem because it affects both the quality and productivity of the casting process. If the number, distribution or size of the hot tears is too large, then the ingot must be scrapped. As can be seen in Figure 1.5, hot tears often appear quite small on the surface, when in fact they can actually propagate quite deep within the body of the ingot, as shown in Figure 1.6. In order to decrease the tendency of hot tear formation, many Al alloys must be cast at slower speeds, decreasing productivity.

![Figure 1.5 - Surface View of an ingot hot tear. The length of the crack is approximately 10 cm](image)
To solve the hot tearing problem, much effort has been put into understanding the conditions under which these tears occur. It has been found that the susceptibility to hot tearing is strongly dependent on the coherency state of the developing solid structure. The formation of hot tears has also been linked to mold cleanliness, through high frictional forces between the mold and the ingot, and the solidification range of the alloy. Hot tears can form because there is a lack of liquid feeding in the mushy zone to counteract the shrinkage associated with the phase change and/or mechanical loading of the semi-solid structure. Experiments and industrial experience have revealed that the problem is most prevalent during the transient start-up phase of the process. Moreover, it has been found that most ingot hot tears are usually found on the rolling face surface, with the crack having initiated near the ingot lip close to the center of this face, and propagated parallel to the casting direction.

In comparison with other metallurgical areas of study, hot tearing related research is much less published. Within the body of scholarly works, empirical relationships, physical models, and experimental data on hot tearing are limited. This is due to the difficulty in performing reproducible experiments within the mushy zone. Although a number of hot tearing

Figure 1.6 - Subsurface view of a hot tear. Note how far the crack extends into the body of the ingot.
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models have been developed within the last few years\cite{7,13,14,18,19,20}, there have only been a few actually compared to experimental data\cite{9,13,21}. In general, researchers give an outline of their model, do a sensitivity analysis, and determine the effects of varying the model parameters. These trends are then compared to literature without comparing actual values.

1.4 Thermal – Mechanical Modeling and Defects in the DC Casting Process

Since the invention of the DC casting process in the 1930's, most of the advancements have come through trial and error. Researchers have used their experience to hypothesize the effects of changing any of the process variables on ingot quality and then performed any required industrial experiments. However, industrial trials are generally time consuming and expensive. In the past twenty years, the emergence of powerful computers and sophisticated numerical simulation techniques has allowed for the development of off-line mathematical models of the DC casting process to predict the temperature and stress/strain fields in the solidifying ingots. These models allow researchers to better understand the DC casting process and improve industrial productivity and recovery in a systematic way as opposed to the trial-and-error approach. Within the model, the casting recipe and geometry can be adjusted and the effect of these adjustments on the temperature, stress, and strain fields can easily be identified.

Although significant improvements have occurred over the past 20 years with regard to the simulation of the casting process, there is still great difficulty in using this tool to predict with certainty which conditions will cause defects to occur. Much of this difficulty is due to the lack of good models to describe defects, especially the hot tearing defect. In the last five years, the study of hot tearing has again become popular at universities. Strong research groups have developed at European universities such as EPFL – Lausanne, Switzerland, SINTEF Materials Technology, Norway, and TU – Delft, the Netherlands. This research has been centered around hot tearing in aluminum billet castings, with little attention paid to rectangular ingots. From these groups, a number of hot tearing criteria and finite element process models have been developed. In the North American aluminum industry, much of the DC Casting product is rectangular ingot. The Department of Metals and Materials Engineering at the University of British Columbia has recently completed the development of a 3D fully coupled thermal and stress model of the start-up phase of DC cast aluminum AA5182 ingots. It is now our intention to study the formation and growth of hot tears focusing on the rectangular ingot shape.
Chapter 2: Literature Review

2.1 Introduction

Among the defects associated with aluminum DC casting, hot tearing is particularly challenging because the formation of these tears is related to both the process parameters and the individual composition of the alloy. The work on hot tearing phenomena in aluminum alloys can generally be subdivided into two distinct areas: (i) experimentally based research to study the nature of hot tear formation and growth, and (ii) development of models to predict process and composition effects on hot tearing susceptibility.

Although experiments have revealed much on the formation and growth of hot tears, the problem remains, in part, because many experiments must be performed using production size equipment. This is because the factors causing hot tearing, thermally and mechanically induced stress and strain, are dependent on the size of the ingot and the process conditions. Furthermore, performing experiments to determine if the adjustments in the process parameters have reduced hot tearing is a long and arduous task. The stress and strain fields, required for input into hot tearing criteria, are also too complex to determine through experiment alone.

The deficiencies associated with experimental work have led to the development of mathematical simulations of the DC casting process. Using the Finite Element Method, researchers have computed both the thermal fields and the stress state in the ingot during the casting process. Properly constructed and verified models are useful as they allow for the adjustment of process parameters within the framework of the model. The effect of casting recipe changes can then easily be computed by examining the new stress, strain and thermal field predictions made by the model. The effect of the casting recipe on hot tearing susceptibility can also be judged by incorporating a hot tearing criterion within the framework of the simulation.

In the following sections in Chapter 2, the theories and criteria of hot tearing that are currently available in the published literature have been critically reviewed.
2.2 Hot Tearing Theory

2.2.1 Stress and Strain Accommodation During Solidification

Generally during solidification, the temperature gradients, solidification contraction and mechanical constraints imposed by the mold results in a distribution of stresses within the semisolid phase. At low fraction solid, the solid – liquid mixture is called a slurry, and is characterized by small solid particles in suspension that are relatively free to move about the liquid[21]. At some fraction solid, the solid grains start to interact with each other and the material begins to acquire strength. This fraction solid is called the shear coherency point. At fraction solid above this point, the mixture is called a mush, and is characterized by a developing solid network with liquid in between[14]. At low fraction solid, the mixture response to load is similar to a highly viscous liquid. In contrast, at high fraction solid, the response to load is very complex, and may include plasticity without hardening, creep, asymmetric constitutive behaviour[22],[23], and large ductility changes with temperature[24].

Clyne and Davies[13] divided the solidification process and response to stress in the semisolid phase into four stages, based on the permeability of the solidifying network: mass feeding, interdendritic feeding, interdendritic separation and interdendritic bridging. The microstructure in each of the four stages is shown in Figure 2.1.

(a) Mass Feeding

In mass feeding, both the liquid and solid are free to move in response to mechanical load and contraction stress as the microstructure consists of solid grains floating in liquid. Mass feeding is generally associated with fraction liquid corresponding to the slurry zone and near the shear coherency point in the mushy zone. There is a lower limit of fraction liquid below which mass feeding is inhibited by grain size and shape.

(b) Interdendritic Feeding:

In interdendritic feeding, the solid is no longer free to move in response to mechanical load and so the remaining liquid has to flow through the dendritic network to assist in stress relief.

(c) Interdendritic separation:

At very high fraction solid, there is a continuous and interconnected liquid film surrounding the dendrites. Stress is accommodated by the formation of pores and hot tears since there is
still little interlocking or bridging of the dendrites and the permeability of the mush is too low to allow for much liquid flow.

(d) *Interdendritic bridging:*

In the last stage of solidification, dendrite bridging occurs and the ingot has developed considerable strength. Further stress is compensated for by high temperature creep.

![Figure 2.1 - Typical microstructure in the semisolid phase corresponding to the four stages of stress relief: mass feeding (a), interdendritic feeding (b), interdendritic separation (c), and interdendritic bridging](image-url)
Interpreting the above four stages can lead to insight into strain accommodation and ductility during solidification. In the mass feeding stage, both the liquid and solid are free to move. Consequently, one would expect little resistance to flow and large amounts of ductility. At higher fraction solid, the configuration of the grains does not permit bulk movement. Instead, liquid must flow through the solid network to offset shrinkage and mechanical loading. This is because the solid does not bear much of the load as the liquid forms a mostly continuous film around each of the solid particles\textsuperscript{191}. The resistance to liquid flow through the network is called permeability, which is a complex function of dendrite size, spacing, and fraction solid. As the permeability decreases (increasing solid fraction), the ductility also decreases since there is less liquid flow. In the limit, with zero permeability (no liquid flow) but still a continuous interdendritic film, the ductility must also be close to zero since the solid still cannot bear load. This limit is the interdendritic separation stage. Once large amounts of dendrite interlocking occur, the material develops considerable strength and increased ductility.

A number of experiments have reinforced the above interpretation. In the 1940's, Singer and Cottrel showed that the ductility of Al – Si alloys drops sharply just above the solidus temperature and corresponds with a morphology change in the fracture surface from ductile to brittle\textsuperscript{251}. At the same time, Pellini conducted a series of solidification experiments on aluminum and steel plates and also showed that the ductility decreases drastically at temperatures slightly above the solidus\textsuperscript{261}. More recently, Magnin\textsuperscript{271} conducted a series of tensile experiments on an Al – 4.5 % Cu alloy at various temperatures between the solidus and the shear coherency point. The ductility results from these tensile tests appear to further validate the strain accommodation theory at intermediate and low fraction solid as the measured ductility increases with increasing fraction liquid. In another series of high – temperature tensile tests, Colley\textsuperscript{41} showed that the fraction solid transition from interdendritic separation to interdendritic bridging was only slightly dependent on strain rate.
2.2.2 Hot Tearing Mechanics

A review of the above literature on semi–solid mechanical behaviour has clearly demonstrated that the hot tearing defect occurs in the interdendritic separation stage. As shown by Rappaz\textsuperscript{[15]} using a succinonitrile-acetone alloy, the interdendritic liquid is a continuous film which cannot sustain tensile stresses, except by liquid flow. In the interdendritic separation stage, the permeability of the mush is too low for liquid to flow. Consequently, tensile stresses can cause the dendrites to separate, and ultimately forming a hot tear. The resulting crack surface is bumpy but covered with a smooth liquid layer and sometimes solid bridges that connect both sides of the crack\textsuperscript{[28]}.

Although the continuous interdendritic liquid film is known to be the cause of hot tearing, the mechanism by which hot tears form is not well understood. A number of theories exist, and fall under four broad categories: Total Strain, Strain Rate, Critical Stress and Hindered Feeding.

(i) Total Strain\textsuperscript{[26]}:

The total strain theory examines the strain accumulated at a spot in the casting during the time that it passes through the liquid feeding and interdendritic separation solidification regimes. If this strain exceeds some critical value then a hot tear is claimed to form.

(ii) Maximum Strain Rate\textsuperscript{[29]}:

In this approach, the strain rate during solidification is limited by the maximum strain rate during solidification above which fracture will occur. As long as the strain rate stays below this value, then hot tears are claimed to not form.

(iii) Critical Stress\textsuperscript{[30]}:

For the critical stress theory, a liquid filled crack is considered as a crack initiation site. The propagation of the crack is determined by a critical stress. The concept of liquid metal embrittlement has also been applied to the stress theory of hot tearing since the surface free energy between the liquid and solid (at grain boundaries) is sufficiently small to easily create hot tears with no ductility and a low yield stress.

(iv) Hindered Feeding\textsuperscript{[31]}:

The hindered feeding theory assumes that hot tears will not occur as long as there is liquid flow sufficient to offset any interdendritic separation caused by tensile loading. Often the strain rate is linked with hindered feeding and the critical strain rate is calculated above which the liquid flow rate cannot offset the rate of solid separation.
Each of these four theories has been used with some success to develop hot tearing criteria, as will be shown in the next section, 2.3: Hot Tearing Criteria. However, it is probable that hot tearing occurs because of a combination of the four theories, as opposed to one dominant theory.

Predicting and understanding the dependence of hot tearing susceptibility on composition is also a significant challenge. A number of researchers have examined hot tearing on binary alloys, and varied the composition between pure aluminum and the eutectic. These experiments have looked at hot tearing in terms of the number of cracks, their length, and the effect of cracking on current loss. Consistently, the results have demonstrated that a plot of hot tearing susceptibility with respect to composition is shaped similar to the Greek letter \( \lambda \), where the highest hot tearing susceptibility occurs at an intermediate solute concentration. It is commonly referred to as the DEL hot tearing curve. Two separate examples are shown in Figure 2.2. This type of result is expected, since hot tearing typically occurs in large freezing range alloys. Large freezing range alloys are more susceptible to hot tearing because these alloys spend more time in the interdendritic separation stage. For binary alloys that are eutectic (such as Al – Cu), the largest freezing range is at an intermediate solute concentration between the pure metal and the eutectic composition.

![Figure 2.2 - Hot Tearing Susceptibility Experiments for Al – Cu alloys as a function of Cu concentration by Clyne (a) and Warrington and McCartney (b) using the electrical current and crack length methods.](image)
2.3 Hot Tearing Criteria

Over the years, a number of criteria have been developed to investigate hot tearing. As with hot tearing theory, the major hot tearing criteria have been developed based on stress, strain, strain rate and hindered feeding parameters. Outlined below are several of the important hot tearing criteria that have been developed in the past 50 years. For a detailed review on other criteria, please refer to hot tearing reviews conducted by Katgerman\textsuperscript{29}, and Sigworth\textsuperscript{30}.

Pellini\textsuperscript{26}, was one of the first researchers to propose a hot tearing criterion. In the 1950's Pellini and his coworkers made periodic x-rays of aluminum and steel plates during solidification. From this work, they recognized that the mushy zone has limited ductility, and suggested the following three-part hot tearing criteria based on the total strain theory:

(a) Hot tears occur at the hot spot of the casting.
(b) Hot tearing is strain controlled, and occurs if the accumulated strain reaches a certain critical value.
(c) The critical strain is strain rate and temperature dependent.

This hot tearing criteria is given in Eq.(2.1):

\[
\varepsilon_{\text{total}}(T, t) \geq \varepsilon_{\text{critical}}(T, \dot{\varepsilon})
\]  \hspace{1cm} (2.1)

where \( \varepsilon_{\text{total}}(T, t) \) is the additive of the strain accumulated in the hot spot above the solidus temperature, and \( \varepsilon_{\text{critical}}(T, \dot{\varepsilon}) \) is the critical strain required for hot tear formation. If the total strain at temperature \( T \) and time \( t \) exceeds the critical strain then a hot tear will form.

Although this criterion seems to be straightforward, and simple to implement, it has largely been ignored. This is because of the great difficulty in reproducibly measuring mushy zone mechanical properties.

In 1976, Feurer\textsuperscript{29,31} examined the influence of alloy composition and solidification conditions on hot tearing. Basing his work on the hindered feeding theory, he proposed that hot tearing was a result of the inability of liquid to feed the solidification shrinkage, and developed two terms, based on liquid feeding and solidification shrinkage. The first term is the liquid volumetric flow rate per unit volume, SPV, which characterizes liquid flow in the interdendritic region. The magnitude of SPV is given by Eq. (2.2):

\[
SPV = \frac{f_l^2 \lambda_2^2 P_s}{24 \pi c^2 \eta L^2}
\]  \hspace{1cm} (2.2)

where \( f_l \) is the fraction liquid, \( \lambda_2 \) is the secondary dendrite arm spacing, \( P_s \) is the feeding pressure,
\( c \) is the tortuosity constant of the network, \( \eta \) is the viscosity and \( L \) is the length of the network. The second term characterizes the velocity of the solidification front, \( SRG \), as a function of the alloy composition and cooling rate. This term is given by Eq. (2.3):

\[
SRG = \frac{\partial \ln V}{\partial t} = \frac{1}{\rho} \frac{\partial \tilde{\rho}}{\partial t} = \frac{(\rho_o - \rho_s + akC_i) \cdot T \cdot f_i^{2-k}}{\tilde{\rho} (1-k)} \frac{m_i C_0}{(1-k)m_i C_0}
\]

where \( \rho_o, \rho_s \) and \( \rho_s \) are the densities of pure Al at the melting temperature, the liquidus and the solidus; \( \tilde{\rho} \) is the average density, \( a \) is the composition coefficient of liquid density, \( C_i \) is the composition of the liquid at the solid – liquid interface that can be related to \( f_i \) by the non-equilibrium lever rule, \( C_o \) is the alloy composition, \( k \) is the equilibrium distribution coefficient, \( m_i \) is the slope of the liquidus and \( \bar{T} \) is the average cooling rate during solidification of the primary phase. If the feeding term (SPV) becomes less than the solidification term (SRG), as shown in Figure 2.3, then there will be insufficient liquid feeding and hot tearing could occur.

Feurer defined the driving force for hot tearing as the difference between the volumetric liquid flow rate and the volumetric solidification rate, as per Eq. (2.4):

\[
DF_{\text{hot tearing}} = SPV - SRG
\]

He then calculated the driving force as a function of composition for Al – Si alloys and showed that this criterion reproduced the DEL curve for hot tearing and alloy composition. He also showed that the hot tearing driving force increases with cooling rate.

One of the drawbacks to this criterion is that it predicts the point of insufficient feeding to be at a high fraction liquid, where hot tearing is not known to occur. In a later work, Katgerman \(^{29}\) calculated the fraction liquid at the point of insufficient feeding as a function of distance from the centre of the billet for various casting speeds using the Feurer criterion. In this work, the fraction liquid at the point of insufficient feeding reached values up to 0.55 for some casting speeds, well above the interdendritic separation region. A second concern of the criterion is that if the liquid hydrostatic pressure were the only mechanism for hot tear formation, then introducing risers at the proper points in the casting should remove hot tears\(^{30}\). This solution does not decrease hot tearing susceptibility. Thirdly, the effect of mechanical loading, an important consideration for hot tearing in ingot castings, is ignored.
In 1981, Clyne and Davies\textsuperscript{131} reported the development of a hot tearing criterion based on the four stage response to stress during solidification, in which the interdendritic separation stage is most susceptible to hot tearing since it is difficult for the solid or liquid to compensate for tensile loading. The Cracking Susceptibility Coefficient, (CSC), was defined as the ratio between the time period that the casting is vulnerable to hot tearing, during interdendritic separation, and the time available for stress relief, during mass feeding, as per Eq. (2.5):

$$CSC = \frac{t_v}{t_r}$$

(2.5)

where $t_v$ is the vulnerable time period and $t_r$ is the stress relief time period. As a first approximation, the time periods chosen were: $t_v = (0.90<f_s<0.99)$ and $t_r = (0.40<f_s<0.90)$. Using the Scheil Equation, the CSC was calculated for different Al-Si alloys. The criterion was then validated by conducting a series of hot cracking susceptibility experiments on Al-Si alloys of various compositions and relating the change in measured electric current to hot tearing. It was noted that the hot tearing predictions made by the CSC criterion correlate well with their experimental results, and also with predictions made by Feurer's\textsuperscript{131} hot tearing criterion.

The CSC criterion provides researchers with a tool for assessing different alloy compositions on the basis of hot tearing. However, there are several problems with this approach.
Since it does not take into account microstructure features, the effects of grain size and morphology on hot tearing are missed. Also, by assuming that hot tearing susceptibility is the ratio of the two time intervals, the effect of process parameters is largely ignored because the criteria considers the relative time to be most important, not the absolute time that the casting spends in the vulnerable region. In fact, Drezet\cite{33} has conducted a study showing that the CSC decreases with increasing casting rate in aluminum billet castings, which is opposite of industrial hot tear observations. Although it may be a useful indicator of hot tearing in certain applications, the CSC criterion does not explicitly include the underlying mechanisms operating to produce hot tears. Furthermore, critical values relating to hot tear formation cannot be defined.

Since the introduction of the CSC criterion, a number of other parametric approaches have been developed to assess hot tearing susceptibility. The criterion by Niyama et al., originally developed for porosity, has also been used for hot tearing\cite{34} and is given by the ratio of the temperature gradient at the solidification front and the cooling rate, as per Eq. (2.6). Flender\cite{35}, looked at the size of the hot spots within the casting, and defined a criterion as per Eq. (2.7), with $L$ defined as the length of the hot spots and $u$ the rate of thermal contraction. However, the CSC criterion has been the most successful of the parametric approaches to hot tearing.

$$CSC_{Niyama} = \frac{G_{sf}}{\sqrt{T}} \quad (2.6)$$

$$CSC_{Flender} = \frac{(T^2 \cdot L)}{(G \cdot u)} \quad (2.7)$$

In 1999, Rappaz, Drezet and Gremaud\cite{7} proposed a strain-rate based hot tearing criterion relating the pressure changes within the mushy zone to hot tearing. During solidification, liquid flow must compensate for the volumic shrinkage due to the liquid to solid phase transformation and the change in volume due to mechanical deformation of the dendritic network. If the net volumic change is negative, there is an influx of liquid into the mushy zone. This flow must in turn result in a pressure differential (driving force), which manifests itself as a pressure drop, as demonstrated in Figure 2.4. At high fraction solid, the pressure drop in the liquid can be large, leading to cavitation of voids. The presence of voids then presumably acts to weaken the structure, causing the further concentration or localization of strain and the eventual formation of a hot tear.

The RDG criterion is based on the pressure drop contributions related to both the deformation of the solid, and the volume change associated with the solid – liquid phase change. In the absence of pores, the relative pressure in the mushy zone, $P_{mushy}$, is given by Eq. (2.8):
Figure 2.4 - Formation of a hot tear during solidification resulting from strain and the inability of liquid to reach areas of high fraction solid. The associated pressure drop is also shown[()].

\[ P_{mush} = P_{pgh} - (\Delta P_{mech} + \Delta P_{shrink}) \]  

where \( P_{pgh} \) is the metallostatic pressure, \( \Delta P_{mech} \) is the mechanical pressure drop, and \( \Delta P_{shrink} \) is the shrinkage pressure drop. If \( P_{mush} \) is greater than some critical value, \( P_{critical} \), a void may form.

To calculate the deformation and shrinkage pressure drop, Darcy's Law was used to relate a mass balance on a small volume of material within the mushy zone and the pressure drop needed for the liquid to feed at the velocity specified by the mass balance, as per Eq. (2.9):

\[ f_i v_i = -\frac{K}{\mu} \frac{dP}{dx} \]  

where \( v_i \) is the velocity of the liquid, \( K \) is the permeability of the mush and \( \mu \) is the liquid viscosity. The traditional Carman – Kozeny equation was used for permeability, as per Eq. (2.10):

\[ K = \frac{\lambda_2^3}{180} \frac{(1-f_s)^3}{f_s^2} \]  

where \( \lambda_2 \) is the secondary dendrite arm spacing. Making the assumption that the liquid velocity is in the direction of the heat flow only, while the solid deformation occurs perpendicular to the heat flow, then the mass balance can be described by Eq. (2.11):

\[ \frac{\partial}{\partial t} (\rho_i f_i + \rho_s f_s) + \frac{\partial}{\partial x} (\rho_i f_i v_{i,x}) + \frac{\partial}{\partial y} (\rho_s f_s v_{s,y}) = 0 \]  

where \( v_{i,x} \) and \( v_{s,y} \) are the liquid and solid velocities in their respective directions. This type of
mass balance has been used in earlier works\textsuperscript{36,37} to predict porosity, but not for hot tearing. In these earlier works, the solid deformation term, $\frac{\partial}{\partial y}(\rho_s f_s v_{x,y})$, was set to zero. The RDG criterion included the solid deformation term by relating it to the transverse strain rate, $\dot{\varepsilon}_p = \frac{\partial v_{x,y}}{\partial y}$, and then defined a hot cracking sensitivity (HCS) as per Eq. (2.12):

$$HCS = \frac{1}{\dot{\varepsilon}_p^{\text{max}}}$$  \hspace{1cm} (2.12)

where $\dot{\varepsilon}_p^{\text{max}}$ is the largest allowed strain rate before hot tears form for a given critical pressure.

Rappaz and colleagues calculated the shrinkage and mechanical depression pressures for various Al-Si alloy compositions under constant cooling conditions and found that the calculated depression pressures followed the so-called DEL curve, as shown in Figure 2.5. They also found good correlation between the HCS predictions made by the RDG criterion, the Clyne and Davies CSC criterion, and experimental data.

![Figure 2.5 - Comparison of HCS criterion to CSC criterion and Spittle and Cushaway measurements. Mode 1 – constant cooling rate during solidification Mode 2 – solidification, constant heat extraction\textsuperscript{7}](image)
One of the main difficulties in using this criterion is knowledge of the value for the critical pressure term, $P_{\text{critical}}$. In their work, Rappaz et al. estimated the critical pressure to be 2.0 kPa, which seems to have been an arbitrary assignment. Also, the effect of total strain is ignored as it is assumed that hot tears will not form if the velocity of the liquid is large enough to compensate the shrinkage and strain rate effects. Lastly, the critical point for hot tearing has been assumed to be void formation. Since voids may also lead to micro porosity, it is difficult to separate the micro porosity and hot tearing defects using the RDG criterion.

Braccini et al.\cite{38}, Grandfield et al.\cite{181}, and Suyitno et al.\cite{39}, have all extended the RDG hot tearing criterion to other scenarios. In a work published in 2000\cite{38}, Braccini used a similar approach as Rappaz\cite{71} to calculate a pressure drop within the mushy zone. However, two changes were made during the derivation relative to the RDG criterion. Firstly, instead of the Carman-Kozeny equation for permeability, an expression based on percolation theory was used as Braccini noted that the grain geometry in aluminum castings lends itself well this approach. The permeability expression is given in Eq. (2.13):

$$K = \frac{1}{32} \left(1 - f_s\right)^3 \left(f_c - f_s\right)^{\mu} \lambda_1^3$$

(2.13)

where $f_c$ is the coalescence fraction solid, $\lambda_1$ is the primary dendrite arm spacing, and $\mu$ is the viscosity. Secondly, an expression for critical pressure, based on the wetting angle, $\theta$, and the liquid-vapour interface surface energy, $\sigma_{lv}$, was proposed as per Eq. (2.14):

$$P_c = \frac{4 \cos \theta \sigma_{lv}}{(1 - f_s)\lambda_1}$$

(2.14)

To predict hot tearing tendencies, the hot cracking sensitivity (HCS) was calculated under various conditions. Based on shear and tensile experiments, Braccini also proposed a rheological model for the mushy zone, including fraction solid and viscoplastic effects as per Eq. (2.15):

$$\sigma_{\text{max}} = \sigma_0 \exp(\alpha f_s) \exp\left(\frac{Q}{RT}\right) \left(\dot{\varepsilon}^p\right)^m$$

(2.15)

where $\sigma_m$ is the current maximum stress in the mushy zone, $m$ is the strain rate sensitivity coefficient, $Q$ is the activation energy, and $\sigma_0$ and $\alpha$ were determined from experimental data.
In 2001, Grandfield\textsuperscript{[18]} looked at applying the RDG criterion for an equiaxed structure, and at determining a physical basis for the hot tearing critical pressure. He reasoned that if the depression pressure from shrinkage and strain rate effects could overcome the capillary pressure than a hot tear would form and so made the capillary pressure at the base of the dendrites the basis for $P_{\text{critical}}$. The RDG criterion was also derived for equiaxed grains by approximating the grains as spheres surrounded by liquid. The depression pressure during solidification was calculated for magnesium AZ91 under both equiaxed and columnar geometry. Using a typical casting recipe, it was shown that hot tears would be expected for columnar grains, and hot tears would not be expected for equiaxed grains, as is found in industry. Grandfield demonstrated that the equiaxed grain structure allowed the liquid permeability to remain much higher at higher fraction solid, reducing the shrinkage and mechanical depression pressures within the mushy zone. Of note in this work is that it was assumed that the interdendritic liquid does not affect hot tearing for equiaxed grains. In contrast, the original RDG criterion was based on the interdendritic liquid in columnar grains. Also, the shrinkage and mechanical depression pressures predicted by Grandfield were four orders of magnitude larger than the values found by Rappaz et al.\textsuperscript{[7]} under similar cooling conditions. No explanation for this discrepancy was given.

In 2002, Suyitno et al.\textsuperscript{[39]} developed a criterion to differentiate between void formation leading to microporosity and void formation leading to hot tearing. As per the RDG criterion, they began with a mass balance on a volume element in the mushy zone, but replaced Darcy's Law for the liquid feeding term with Feurer's\textsuperscript{[31]} term for liquid volumetric flow rate, SPV. It was assumed that the voids form as perfect spheres. The volume fraction decrease over time caused by shrinkage and strain was then compared with the liquid volumetric flow rate term for various strains, as shown in Figure 2.6a. The intersection of the curve represents the point of porosity formation. To determine if the pore would translate into a hot tear or remain as microporosity, the Griffith approach was used to calculate the relationship between void diameter and the minimum stress for crack propagation, as per Eq.(2.16):

$$d_{\text{crit}} = 4\gamma_e \frac{E}{\pi \sigma_m^2}$$ (2.16)

where $d_{\text{crit}}$ is the crack length, $\gamma_e$ is the surface tension of the liquid metal, $\sigma_m$ is the minimum stress and $E$ is Young's modulus. If the minimum stress becomes equal to the stress in the mush, any pore diameter larger than $d_{\text{crit}}$ will propagate as a hot tear. The stress in the mushy zone was
calculated using the rheological model proposed by Braccini\textsuperscript{[38]}, and the pore diameter was calculated based on the strain rate and the total solidification time.

Using this formulation, Suyitno, showed, in Figure 2.6b, that the porosity without hot tearing window is quite large during solidification with low strain rates. However, at higher strain rates, it becomes quite easy to propagate hot tears. The Suyitno criterion also captures quite nicely the benefits of a multi-stage hot tearing criterion, incorporating both nucleation and growth effects, and including a strain rate and stress component. Missing from this work is a discussion on the choice of the effective feeding pressure, $P_s$, used in the liquid volumetric flow rate term, since this variable highly influences the results.

In 2001, Lahaie and Bouchard\textsuperscript{[19]} devised a hot tearing criterion based on the idea that the strength of the liquid trapped between grain boundaries is the limiting factor for hot tearing. The critical hot tearing value is the stress necessary to separate two grains bonded by capillary force. An idealized microstructure was given, as shown in Figure 2.7. The microstructure consists of a hexagonal solid geometry surrounded by uniform thin films of liquid such that $h \ll a$, where $h$ is the film thickness before deformation and $a$ is a function of the grain size. From this idealized geometry, the capillary pressure is given by the Young and Laplace equation modified for the idealized hexagonal microstructure, as per Eq. (2.17):
Figure 2.7 - Idealized Microstructure used by Lahaie and Bouchard\textsuperscript{119}. 

\[ \sigma_{cap} = \frac{4 \gamma_{liq}}{3h} \left( 1 + \left( \frac{f_s^m}{1 + f_s^m} \right) \epsilon \right)^{-1} \]  

(2.17)

where $\gamma_{liq}$ is the surface tension of the liquid, $\epsilon$ is the current strain, $f_s$ is the fraction solid and $m$ is a microstructure parameter, which is 1/3 for columnar and 1/2 for equiaxed structure. Lahaie and Bouchard also proposed a physical model for the deformation of a semisolid body based on the idealized hexagonal geometry. In this model, deformation is caused solely by viscous flow or dilatation of the liquid. Although strain is implied, actual deformation of the solid does not occur. As the strain in the body is increased, the applied stress is compared to the capillary pressure. Once the applied stress is greater that the capillary pressure, hot tears will propagate within the mush.

The calculated hot tearing stress and strain are shown in Figure 2.8 for columnar and equiaxed grains. The stress predictions made by this criteria include: higher fracture stress with increased fraction solid and decreased grain size, which qualitatively agree with published experimental knowledge, our understanding of the effect of grain size on strength, and the large strength differences between liquids and solids. The predicted fracture strain using this criterion linearly decreases with fraction solid and is independent of grain size. These predictions are opposite to experimental observations, which have shown that the ductility increases with fraction solid\textsuperscript{41} at high fraction solid. It is also highly questionable that fracture strain is independent of grain size. In their work, Lahaie and Bouchard did a sensitivity study on their results but did not compare their work to available experimental data, nor did they discuss how the strain predictions made by this criterion compare against conventional results.
Figure 2.8 - The calculated fracture stress and fracture strain of a semisolid Al alloy for columnar and equiaxed grains as a function of fraction solid (a) and grain size (b) at $f_s = 0.99$.

In a series of works, Mo and colleagues\textsuperscript{[14],[40],[41]} have developed a two-phase (solid + liquid) material model incorporating both hot tearing mechanisms: shrinkage driven melt flow and thermally induced deformation. The mushy zone is modeled as a porous metallic material saturated with liquid and relates the viscoplastic strain rate tensor to the percent cohesion of the mushy zone, after Martin et al.\textsuperscript{[42]}. In this model, the pressure drop in the liquid, and the state of stress in the solid is calculated during solidification.

In conjunction with the two-phase model, Mo\textsuperscript{[40]} has suggested an Integrated Critical Strain (ICS) criterion to quantify the hot tearing susceptibility, as per Eq. (2.18):

$$ICS = \begin{cases} 0 & \text{if } \frac{p_l f^{\text{coal}}}{p_c} > p_c \\ \int_{\bar{p} = \bar{p}_l}^{\bar{p} = \bar{p}_r} \theta \sqrt{w_1 \text{tr} \left( \dot{\varepsilon}_s^p \right)^2 + w_2 \left( \dot{\varepsilon}_s^p \right)^2} dt & \text{otherwise} \end{cases}$$

where $p_l$ is the liquid pressure calculated by the two-phase model, $f^{\text{coal}}_s$ is the fraction solid for grain coalescence, $p_c$ is the critical pressure for void nucleation, $\theta$ is a Dirac function with a value of 1 if the mean stress is positive and 0 if the mean stress is negative, $\dot{\varepsilon}_s^p$ and $\dot{\varepsilon}_s^p$ are the volumetric effective deviatoric and viscoplastic strain rates, and $w_1$ and $w_2$ are weighting functions. As long as the liquid pressure is greater than the critical value, there is sufficient liquid feeding and no hot tears will form. Under this condition, the ICS index has a value of zero. If the liquid pressure falls below the critical value, liquid feeding becomes reduced.
Further straining of the metal will lead to the growth of pores into hot tears. Under this condition, the ICS index can be calculated by integrating the viscoplastic strain rate in the region bounded by the start of insufficient melt feeding, when the pressure in the liquid is smaller than the critical pressure, and the fraction solid where the metal has sufficient ductility to prevent hot tearing formation, $f_s^{cool}$.

As with the Suyitno criterion\cite{39}, this approach presents a sophisticated multi-stage approach to hot tearing. The ICS criterion calculates pore nucleation based on the liquid pressure drop, and relates pore growth to the viscoplastic strain. Unfortunately, this criterion struggles to define the critical values of $p_c$ and $f_s^{cool}$ and the weighting functions $w_1$ and $w_2$. The actual values chosen seem arbitrary at best. Moreover, the ICS index does not define critical values for hot tear formation. Instead, only the relative hot tearing susceptibility of one set of process parameters as compared to another set of parameters can be discussed.
2.4 Hot Tearing and Finite Element Modeling

The increase in computing power over the last few years has prompted the development of large finite element models of various casting processes. Since the goal of these models is process improvement, it is only natural to incorporate hot tearing criteria into their formulation. Drezet\cite{33} implemented the RDG criterion into an axi-symmetric FE model of the aluminum billet DC casting process. The liquid pressure drop was calculated in the start-up and steady-state regions. It was shown that the highest pressure drop occurs in the centre of the billet and that the area of high pressure drop is quite large in the start-up region, and narrow in the steady-state region. The effect of casting velocity was also assessed. From Figure 2.9, it is quite clear that higher casting velocities lead to higher pressures drops at the centre of the billet. As the depression pressure relates to the susceptibility to crack initiation, one can conclude that during start-up, the centre region of the billet is most prone to hot tearing, and strongly affected by casting velocity. This correlates well with industrial experience.

Rindler\cite{34} implemented the RDG criterion modified by Grandfield\cite{18} to an FE model of steel welding. The calculated strain rate in the mushy zone was compared to the critical strain rate calculated by the modified criterion for two alloys systems: austenitic stainless steels, which are prone to hot tearing, and structural steels, which are not prone to hot tearing. Both simulations were run using the same welding recipe. The results are shown in Figure 2.10. It is

![Graph showing steady-state depression pressure as a function of radius for several casting velocities.](image)

\textbf{Figure 2.9} - Steady-state depression pressure calculated using the RDG criterion as a function of radius for several casting velocities\textsuperscript{[33]}. 

Figure 2.10 - FE simulated strain rate and hot tearing critical strain rate for austenitic stainless steel alloy CrNiMo18-14-3 (a) and structural steel alloy S335 J2G1W (b). The BTR box outlines the region vulnerable to hot tearing \((0.02 < f_i < 0.2)\).\(^\text{[34]}\)

It is clear that the FE-simulated strain rate of the austenitic steel exceeds the critical hot tearing strain rate in the vulnerable region and thus hot tears are prone to occur. Within the vulnerable region of the structural steel, the critical hot tearing strain rate is always higher than the calculated strain rate and hot tears are much less likely to form.

To verify the ICS criterion\(^{[41]}\), Mo and coworkers incorporated the two-phase model into an axi-symmetric FE model of the aluminum billet DC casting process. The ICS index was then calculated for five different casting recipes consistent with experimental billets cast to produce hot tears. The computed ICS along the centreline of the billets, as shown in Figure 2.11, has a maximum just above the bottom block before decreasing to a steady state value, confirming that the hot tearing susceptibility is largest in the start-up region. The relative ranking of these five plots, in terms of their susceptibility, is T1, T3, T2, T4, T5. Experimentally, hot tears were found in trials T1, T3 and T2, with T1 having the highest casting velocity and being most severely cracked. Hot tears were not found in T4 and T5. Thus, the ICS index rankings match well with the experimental results.
Phillion and Cockcroft\cite{43} implemented the RDG criterion into a 2D FE model of aluminum ingot DC casting. The 2D slice was taken perpendicular to the rolling face, at the centre. Both the mechanical and shrinkage depression pressure was calculated in the start-up and steady-state regions. In this work, the depression pressure was calculated directly over the solidification distance. The model predictions seem to be consistent with the hot tearing behaviour observed industrially, as the highest depression pressure was observed near the surface of the ingot, just above the lip of the bottom block.
2.5 Summary

In this chapter, theory and criteria relating to hot tearing have been discussed. The physics and basic phenomena that lead to hot tearing are generally well understood: hot tearing occurs at high fraction solid when there is a continuous interdendritic film of liquid. In this stage of solidification, the mushy zone has little strength and no ductility.

Researchers have proposed a variety of mechanisms for hot tear formation: critical stress or strain rate, total strain, and hindered feeding. These mechanisms have led to the formulation of multiple hot tearing criterion. A number of these criteria have been reviewed in this chapter, with a summary of the advantages and disadvantages given in Table 2.1. However, a quantitative criterion that will predict hot tear formation under general conditions is still not available. Based on this literature review, it is clear that hot tearing is a complex phenomena that is difficult to predict. To model hot tearing, both a hot tearing criterion and the thermal and stress – strain fields present during the DC casting process are required.

<table>
<thead>
<tr>
<th>Hot Tearing Criterion</th>
<th>Alloy com position</th>
<th>Microstructure</th>
<th>Mechanical Loading</th>
<th>Critical Value</th>
<th>Hot Tear Growth</th>
</tr>
</thead>
<tbody>
<tr>
<td>Pellini</td>
<td>+</td>
<td>+</td>
<td>+</td>
<td>+</td>
<td>-</td>
</tr>
<tr>
<td>Feurer</td>
<td>+</td>
<td>-</td>
<td>-</td>
<td>+</td>
<td>-</td>
</tr>
<tr>
<td>Clyne &amp; Davies</td>
<td>+</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Rappaz</td>
<td>+</td>
<td>+</td>
<td>+</td>
<td>+</td>
<td>-</td>
</tr>
<tr>
<td>Braccini</td>
<td>+</td>
<td>+</td>
<td>+</td>
<td>+</td>
<td>+</td>
</tr>
<tr>
<td>Grandfield</td>
<td>+</td>
<td>-</td>
<td>+</td>
<td>+</td>
<td>-</td>
</tr>
<tr>
<td>Suyitno</td>
<td>+</td>
<td>-</td>
<td>+</td>
<td>+</td>
<td>+</td>
</tr>
<tr>
<td>Lahaie &amp; Bouchard</td>
<td>-</td>
<td>+</td>
<td>+</td>
<td>+</td>
<td>-</td>
</tr>
<tr>
<td>Mo</td>
<td>-</td>
<td>-</td>
<td>+</td>
<td>-</td>
<td>+</td>
</tr>
</tbody>
</table>
Chapter 3: Scope and Objectives

3.1 Scope of the Research Programme

One of the main defects in aluminum DC casting is hot tearing. The purpose of this work was to investigate and predict hot tearing phenomena in aluminum ingot DC castings. This knowledge has then been used to implement both Pellini's Total Strain criterion and the RDG hot tearing criterion into a finite element model of the ingot DC casting process, and to assess the validity of the hot tearing predictions made by these criterion under different casting recipes.

Over the years, a number of finite element models have been developed to predict the thermal, stress, and strain fields during the DC casting process. A good summary of these models has been provided by Sengupta[31]. Although the problem of hot tearing is often mentioned as the reason for developing such process models, the literature review presented in Chapter 2 – Literature Review has shown that only limited work has been done linking hot tearing criteria and process models. A few have been reported in the literature for the case of aluminum DC cast billets[14],[18],[33],[41], but little work has been done in the more complicated case of ingot castings. In the current work, hot tearing criteria were implemented into a 3D fully coupled thermal and stress finite element model of the DC casting of AA5182 ingots, developed previously at the University of British Columbia by Dr. Joydeep Sengupta and outlined in [3], [44],[45], and [46]. The hot tearing predictions made by these criteria were then discussed in terms of hot tearing susceptibility observed in industry.

3.2 Objectives of the Research Programme

In order to assess the hot tearing susceptibility during aluminum ingot DC casting, the objectives of this research project can be summarized as follows:

(1) To develop a fundamental understanding of the hot tearing phenomenon as it applies to aluminum ingot DC casting;

(2) To implement the Total Strain and RDG hot tearing criteria into the ingot casting model;

(3) To analyze the predictions made by these criteria when the model is run under different sets of casting recipes, with the aim of predicting which casting recipe produces hot tears, and the location of these hot tears;

(4) To propose a hot tearing criterion based on our knowledge of [1-3] that is both applicable and customized to the situation of rectangular ingot castings.
Chapter 4: DC Casting Model Formulation

4.1 Introduction

In order to study and model the hot tearing defect at the process level, both the thermal and displacement fields must be known. The thermal information provides knowledge of the solidification sequence, and areas of feeding difficulties. The displacement information, along with the mushy zone mechanical properties, allows one to calculate the stresses and strains in the final stages of solidification. For a number of years, the casting group with the Centre for Metallurgical and Process Engineering at the University of British Columbia has been developing a finite element model of the ingot DC casting process of aluminum alloy AA5182, a non-heat treatable aluminum alloy with nominal composition 4.5 wt% Mg and 0.35 wt% Mn. Its main usage is as the lid of aluminum beverage containers, and it is prone to hot tearing. To create this process model, the following is required:

- Calculation domain, geometry, and formulation
- Thermo – physical material properties
- The boundary and initial conditions

The commercial software package ABAQUS was chosen for this work, as it is known to be an excellent platform for solving transient, highly non-linear heat transfer and stress – strain problems. For more information on the DC casting model development and validation, refer to references [44],[45], and [46], as this algorithm was not developed during the course of this research.

4.2 Model Formulation, Domain, and Geometry

4.2.1 Analysis Formulation

The finite element prediction of the DC casting process requires the solution of a heat transfer analysis and a stress – strain analysis. For the heat transfer analysis, fluid flow has been ignored and heat is assumed to be transported by diffusion only. Latent heat was included to account for the effect of the liquid to solid phase change. The governing differential equation for the heat transfer portion of the outlined problem is given by Eq. (4.1):

\[ \nabla [k(T) \nabla T] + \dot{Q} = \rho c_p \frac{\partial T}{\partial t} \]  (4.1)
where $\rho$ is the density in kg m$^{-3}$, $c_p$ is the specific heat in J kg$^{-1}$K$^{-1}$, $T$ is the temperature in K, $k$ is the thermal conductivity in W m$^{-1}$K$^{-1}$, and $\dot{Q}$ is the latent heat of solidification in W m$^{-3}$. For the mechanical analysis, ABAQUS solves the force and momentum equilibrium for the entire geometry by simultaneously relating all of the nodal displacements and forces through the stiffness matrix. The stress-strain state can then be calculated within any element using the interpolation functions and the relationship between strain and displacement. If the material behaviour is in the elastic region, then the stiffness matrix incorporates the linear relationship between the total stress and total strain, i.e. Hooke’s law as in Eq. (4.2):

$$\sigma = [D^{el}][\varepsilon^{el}]$$  \hspace{1cm} (4.2)

where $[D^{el}]$ contains the elastic constants related to the elastic modulus and Poisson’s ratio, $[\sigma]$ is the stress tensor and $[\varepsilon^{el}]$ is the total elastic strain tensor. If the material behaviour is in the plastic region, then the stiffness matrix incorporates the relationship between the current increment of stress $[d\sigma]$, and the current increment of strain, $[d\varepsilon^{pl}]$, as per Eq. (4.3):

$$[d\sigma] = [D^{pl}][d\varepsilon^{pl}]$$  \hspace{1cm} (4.3)

ABAQUS allows for four different approaches for describing the constitutive behaviour: elastic – plastic, elastic rate – dependent plastic, elastic – creep and combined elastic – creep – plastic. As shown by McDonald et al.$^{[48]}$, the finite element stress and strain predictions are highly dependent on the choice of hardening law. In another work, Alhassan-Abu and Wells$^{[46]}$ compared the performance of these hardening laws under conditions consistent with the DC casting process to represent the constitutive behaviour of as-cast AA5182, and found that of these laws, the elastic rate-dependent plastic hardening rule was best suited for the DC casting process. Thus, the rate-dependent hardening rule was used in the current DC casting model to calculate stresses and strains in the ingot during solidification and cooling.

A fully coupled thermal – displacement analysis is required in the DC casting process. The term coupled refers to the fact that the thermal and mechanical equations are solved simultaneously. This coupling is required to link the heat transfer between the ingot and the bottom block to the magnitude of the butt curl. Other sources of coupling, such as the formation of a gap between the solidifying ingot and the mold were not explicitly considered in the current model. For a fully coupled thermal – displacement analysis, ABAQUS simultaneously solves the thermal and mechanical equations as shown in Eq. (4.4):
where $\Delta u$ and $\Delta T$ are the respective corrections to the incremental displacement and temperature, $K_u$ are submatrices of the fully coupled Jacobian matrix, and are $R$, the mechanical and thermal residual vectors, respectively. The ABAQUS software uses Newton's method to simultaneously solve the thermal and mechanical equations and a backward-difference scheme for time integration of the temperature and displacements at every Gauss point.

4.2.2 Model Domain

The analysis domain included the start-up region of the ingot (as defined in Chapter 1 - Introduction), and the bottom block. The model geometry was a rectangular ingot with a cross-section of 1638 mm by 711 mm, and a length of 1000 mm. By symmetry along the rolling and narrow face centrelines of the ingot, the problem was reduced to a quarter-section, with actual dimensions 819 mm by 355 mm. The bottom block had a similar cross-section, and a height ranging from 210 mm in the bowl of the bottom block to 305 mm at the ingot lip. The total casting time modeled was about 800 s, which allowed the centre of the ingot to cool to near room-temperature values. To simulate DC casting, a Lagrangian approach was used, and the thermal boundary conditions were moved up the domain at a rate consistent with the ingot withdrawal velocity. Horizontal layers of ingot elements were incrementally added consistent with the casting recipe based on the mold filling rate and casting speed.

4.2.3 Finite Element Geometry

The FE mesh for the model was constructed using 3300 8-noded iso-parametric coupled temperature – displacement brick elements, each with 8 gauss integration points, and 4240 nodes. 2094 elements were used to construct the ingot, each with a horizontal cross-section of approximately 80 mm in length and 30 mm in width. The height of these elements was 50 mm in the region cooled by the water sprays and 20 mm to 30 mm in the region contained by the bottom block. 1206 elements were used to construct the bottom block. These elements had a similar cross-section to the ingot elements, and a height ranging from 30 mm to 40 mm.
4.3 Thermal and Mechanical Properties of AA5182

For a fully coupled thermal – displacement finite element problem, material properties are required to describe the process of heat transfer through the system, the stress-strain response of the material, and the thermal expansion. Since temperatures in the analysis range from room temperature (25°C) to the pour temperature of the liquid (660°C), each property's temperature dependency must also be known. In ABAQUS, material property is given as tabular format. The software linearly interpolates between the tabular data to assign integration point property values over the entire temperature range. In the current DC casting model, the material properties of both the ingot and the bottom block were required. Material properties for the ingot, specified to be an AA5182 aluminum alloy, were taken from the literature. The bottom block is made from an aluminum AA6xxx alloy. As the required material properties were not available for this alloy, the bottom block was also assumed to be made from AA5182.

4.3.1 Thermal Properties

The material properties required for the thermal analysis: density, heat capacity, latent heat, and thermal conductivity, were taken from Wiskel\textsuperscript{50}, who compiled these properties from information available in the literature. These properties are presented in Table 4.1. The latent heat was assumed to evolve linearly with temperature between the liquidus and solidus temperatures. Also, the density of the metal has been assumed independent of temperature, as a temperature dependent density would alter the heat balance by increasing element mass during the analysis. Since the model transports heat through a diffusion mechanism only, the thermal conductivity values above the liquidus were increased to four times their actual value to take into account heat transport due to convection in the liquid sump\textsuperscript{51}.

4.3.2 Mechanical Properties

The mechanical properties required for the mechanical analysis include the coefficient of thermal expansion, Poisson's ratio, temperature dependent elastic moduli, and temperature and strain rate dependent plastic stress-strain curves. Since the model contains both solid and liquid phases in a single domain, the constitutive behaviour must transition from the solid concept of elastic strain, yielding and then plastic strain accumulation to the liquid concept of viscous flow without any accumulation of plastic strain. The solid to liquid transition temperature for
AA5182 was shown by Colley\cite{4} to be approximately 580°C, at a fraction solid of roughly 90\%\cite{52}. Above this temperature, the metal mushy zone properties tend towards behaviour similar to a viscous liquid. The coefficient of thermal expansion, Poisson's ratio, and elastic moduli data were taken from Wiskel\cite{50} and are shown in Table 4.2. The stress-strain data, taken from Alhassan-Abu\cite{53}, and Colley\cite{4}, is shown in Figure 4.1 and Figure 4.2.

**Coefficient of Thermal Expansion:**

The syntax in ABAQUS requires the use of a temperature dependent total thermal expansion coefficient. This thermal expansion coefficient can be imagined as the average or equivalent thermal expansion coefficient over the range of interest. It is defined as per Eq. (4.5):

$$\alpha(T) = \frac{1}{T - T^0} \int_{T^0}^{T} \alpha'(T) dT$$  \hspace{1cm} (4.5)

where $\alpha$ is the effective thermal expansion, $\alpha'$ is the temperature dependent coefficient of thermal expansion, and $T^0$ is a reference point specifying the critical temperature where the material can be assumed to have zero thermal strain. Since the material appears to behave as a liquid above 580 °C and the concept of strain accumulation does not apply, the zero thermal expansion temperature was chosen to be 580 °C.

**Elastic Modulus:**

The elastic moduli values range from 71 GPa at room temperature to 28 GPa at the solidus temperature. Moduli values at temperatures greater than 580 °C were significantly lowered to give the metal the liquid property of flow due to tensile stress.

**Stress – Strain Data:**

At this time, there is little published information regarding the constitutive behaviour of AA5182 under the conditions present in the DC casting process, especially with as-cast microstructure and in the mushy zone at high fraction solid. Alhassan-Abu\cite{53}, and Colley\cite{4}, conducted different experimental studies on the stress-strain curves below and above the solidus temperature of AA5182 at strain rates applicable to aluminum DC casting: $10^{-3}$ s$^{-1}$, $10^{-2}$ s$^{-1}$, 0.1 s$^{-1}$, and 1 s$^{-1}$. At the mechanical coherency temperature of 580 °C, the material was given a low yield stress of 1.0 MPa as the metal has no load-bearing ability at this temperature. Above this temperature, the yield stress was increased to 10 MPa. The combination of a low elastic modulus and appreciable yield stress ensure that above 580 °C, the metal will respond as a liquid and easily flow in the direction of the tensile stress without accumulating plastic strain.
### Table 4.1 - Thermal Properties of AA5182 Required for Thermal Analysis\(^{[50]}\)

<table>
<thead>
<tr>
<th>Property</th>
<th>Temperature Range (°C)</th>
<th>Equation (T in °C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Conductivity</td>
<td>T &lt; 577</td>
<td>119.2 + 0.0623 T</td>
</tr>
<tr>
<td></td>
<td>577 ≤ T ≤ 638</td>
<td>594 - 0.484 T - 0.00048 T(^2)</td>
</tr>
<tr>
<td></td>
<td>T &gt; 638(^a)</td>
<td>69 + 0.033 T</td>
</tr>
<tr>
<td></td>
<td>T &gt; 638(^b)</td>
<td>-5600 + 9.09 T</td>
</tr>
<tr>
<td>Specific Heat</td>
<td>T &lt; 577</td>
<td>897 + 0.452 T</td>
</tr>
<tr>
<td></td>
<td>577 ≤ T ≤ 638</td>
<td>-994.8 + 8 T - 0.0074 T(^2)</td>
</tr>
<tr>
<td></td>
<td>T &gt; 638</td>
<td>1097</td>
</tr>
<tr>
<td>Density</td>
<td>N/A</td>
<td>2400</td>
</tr>
<tr>
<td>Latent Heat</td>
<td></td>
<td>(Q_{LH} = 397.1\text{kJkg}^{-1})</td>
</tr>
<tr>
<td>(T_{\text{Liquidus}})</td>
<td>637°C</td>
<td>N/A</td>
</tr>
<tr>
<td>(T_{\text{Solidus}})</td>
<td>536°C</td>
<td>N/A</td>
</tr>
</tbody>
</table>

\(^{a}\) Actual conductivity at temperatures as per \([50]\).

\(^{b}\) Modified conductivity to include effect of liquid convection in the sump.

### Table 4.2 - Selected Mechanical Properties of AA5182 Required for Stress Analysis\(^{[50]}\)

<table>
<thead>
<tr>
<th>Variable</th>
<th>Temperature Range (°C)</th>
<th>Equation (T in °C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Elastic Modulus</td>
<td>T &lt; 500</td>
<td>71600 - 31.35 T - 3.452x10(^-2) T(^2)</td>
</tr>
<tr>
<td></td>
<td>500 ≤ T ≤ 577</td>
<td>308945 - 532.5 T</td>
</tr>
<tr>
<td></td>
<td>577 ≤ T ≤ 638(^a)</td>
<td>72259 - 113.1 T</td>
</tr>
<tr>
<td></td>
<td>T &gt; 638(^a)</td>
<td>100</td>
</tr>
<tr>
<td></td>
<td>T = 577(^b)</td>
<td>100</td>
</tr>
<tr>
<td></td>
<td>T = 660(^b)</td>
<td>0.01</td>
</tr>
<tr>
<td>Total Thermal</td>
<td>T ≤ 580</td>
<td>-0.0235 + 2x10(^-5) T + 4x10(^-8) T(^2)</td>
</tr>
<tr>
<td>Expansion Coefficient</td>
<td>T &gt; 580(^c)</td>
<td>0.00</td>
</tr>
<tr>
<td>Poisson’s Ratio</td>
<td>N/A</td>
<td>0.3</td>
</tr>
</tbody>
</table>

\(^{a}\) Actual Elastic Moduli for AA5182 as per \([50]\).

\(^{b}\) Modified Elastic Moduli values including effect of viscous liquid flow.

\(^{c}\) Coefficient of Thermal Expansion chosen to represent liquid properties.
Figure 4.1 - Experimental stress-strain curves for AA5182 at different temperatures (25 - 580 °C) and at strain rates of 0.00001 s\(^{-1}\) (a), and 0.001 s\(^{-1}\) (b) which were tabulated in ABAQUS\(^{[46]}\).
Figure 4.2 - Experimental stress-strain curves for AA5182 at different temperatures (25 – 580 °C) and at strain rates of 0.1 s$^{-1}$ (a), and 1.0 s$^{-1}$ (b), which were tabulated in ABAQUS$^{[46]}$. 

Strain rate = 1.0 s$^{-1}$
4.4 DC Casting Model Initial and Boundary Conditions

4.4.1 Initial Conditions

Thermal initial conditions were specified for both the ingot and the bottom block. All the nodes in the ingot were set to an initial temperature of 660 °C, the conventional pour temperature of AA5182. The initial temperature of the bottom block was set to 25 °C.

4.4.2 Thermal Boundary Conditions

Thermal boundary conditions were placed on all surfaces on both the ingot and the bottom block as shown in Figure 4.3. The interior symmetry faces ($\Gamma_2$, and $\Gamma_6$) were assumed to be adiabatic, as was the top face of the ingot ($\Gamma_4$). Cauchy – type boundary conditions were applied to the vertical exterior surfaces ($\Gamma_1$, $\Gamma_3$, and $\Gamma_7$), and the interface between the ingot and the bottom block, ($\Gamma_5$). The exterior surfaces with Cauchy – type boundary conditions require further attention as the heat transfer coefficient is highly non-linear in these regions.

Figure 4.3 - The 3D mesh for the model, with the surfaces defined for the thermal BC's[46].
**Ingot Rolling and Narrow Faces (Surface \( \Gamma_{r} \))**

On the ingot rolling and narrow faces, the magnitude of the heat transfer coefficient was varied to account for the different cooling regimes within the mold, and then in the water spray region below the mold. The water spray region can be further divided into an impingement zone, where the spray water first contacts the ingot's surface, and a free falling zone, where the spray water runs down the face of the ingot after impingement.

Within the mold, the surface of the ingot solidifies, and the heat transfer coefficient must take into account the transition from liquid to solid, and the loss in thermal contact with the mold during this phase change. The formulation of this heat transfer coefficient is described by Eq. (4.6):

\[
h = h_{\text{contact}}(1 - f_s) + h_{\text{gap}} f_s
\]

where \( h_{\text{contact}} \) represents the heat transfer coefficient when the ingot was in good thermal contact with the mold (2000 W m\(^{-2}\)K\(^{-1}\)), \( h_{\text{gap}} \) represents the heat transfer coefficient with no ingot – mold thermal contact (50 W m\(^{-2}\)K\(^{-1}\)), and \( f_s \) is the fraction solid at the location of interest. The fraction solid – temperature curve was given by a proprietary equation developed by Alcan.

As the ingot is cooled by the water sprays, heat transfer proceeds according to the boiling water heat transfer curve. Empirical proprietary correlations developed by Alcan and modified by UBC were used to calculate the transient heat transfer coefficient. These correlations are dependent on the ingot surface temperature, the ingot surface temperature at the impingement point, the water flow rate and the current vertical distance below the impingement point. A particularly unique feature of the code is the addition of water ejection due to film boiling in the free falling zone. The film boiling regime of the boiling water heat transfer curve is characterized by the formation of a stable vapour film on the ingot surface. This occurs when an area on the ingot's surface is still above the Liedenfrost temperature as it enters the free falling zone. Until the vapour film breaks down, the free-falling water is not able to wet the surface and heat transfer becomes significantly reduced between the point of vapour film formation and the base of the ingot.

A number of other features of the cooling on the ingot rolling and narrow faces were taken into account in the ingot DC casting model. For more information on these features, the implementation structure of the boundary conditions, and physical interpretation, refer to [45].
Bottom Block – Ingot Interface (Surface $\Gamma_3$)

The heat transfer at the bottom block – ingot interface consists of two separate processes, shown in blocks A and B of Figure 4.4. Firstly, the hot ingot transfers heat directly to the bottom block via an interface or gap heat conductance. As the base of the ingot deforms, it pulls away from the bottom block, creating an air gap between the two surfaces affecting the magnitude of the conductive heat transfer. At the same time, the presence of an air gap allows some of the cooling water streaming down the ingot vertical faces to flow between the ingot and bottom block enhancing convective transport to the water. The transfer of heat in the air gap from both the ingot and bottom block to the cooling water is known as water incursion.

The bottom block upper surface and ingot lower surface were coded as separate surfaces within ABAQUS. Separate boundary conditions describing gap conductance and water incursion were applied to these surfaces. Gap heat conductance represents heat flow out of the ingot and equal heat flow into the bottom block, and is described by Eq. (4.7):

\[
\text{Gap Heat Conductance} = \text{Conduction from Ingot} = \text{Conduction to Bottom Block}
\]

![Diagram showing heat transfer at the bottom block-ingot interface](image)

Figure 4.4 - Competing heat transfer processes in Base Cooling on surface $\Gamma_3$ near the centre of the ingot (Region A) and outer edges (Region B).\(^{[85]}\)
\[ q_{\text{ingot, cond}} = h_{\text{gap}} (T_{\text{ingot}} - T_{\text{bottom block}}) \]
\[ q_{\text{bottom block, cond}} = h_{\text{gap}} (T_{\text{bottom block}} - T_{\text{ingot}}) \] (4.7)

where \( h_{\text{gap}} \) represents the gap heat conductance at the interface in W m\(^{-2}\)K\(^{-1}\), and \( T_{\text{ingot}} \) and \( T_{\text{bottom block}} \) represent the integration point temperatures at the linked positions on the ingot and bottom block.

The water incursion heat transfer approximates heat lost from both interface surfaces to the cooling water in the air gap through Eq. (4.8), where \( f_{\text{water}} \) represents a factor used to account for the degree to which the interface is wetted by the cooling water, \( h_{\text{water}} \) is the heat transfer coefficient according to the boiling water heat transfer curve and \( T_{\text{sink}} \) is the far-field water temperature:

\[ q_{\text{ingot, water}} = f_{\text{water}} h_{\text{water}} (T_{\text{ingot}} - T_{\text{sink}}) \] (4.8)
\[ q_{\text{bottom block, water}} = f_{\text{water}} h_{\text{water}} (T_{\text{bottom block}} - T_{\text{sink}}) \]

At each integration point along the interface surface, the current vertical separation distance between the ingot and bottom block was computed. The heat transfer coefficient, \( h_{\text{gap}} \) and wetting factor \( f_{\text{water}} \) were ramped linearly with vertical separation, according to Table 4.3. The heat transfer coefficient \( h_{\text{water}} \) was taken from the boiling water curve as outlined in [45]. In this manner, gap conductance heat transfer was decreased to reflect the growing separation between the two surfaces, and the water incursion heat transfer was increased, as water could more easily flow in the interface.

**Table 4.3 - Lookup Table for Ingot - Bottom Block Interface Heat Transfer**

<table>
<thead>
<tr>
<th>Ingot – Bottom Block Separation Distance (mm)</th>
<th>Magnitude of HTC (W m(^2)K(^{-1}))</th>
<th>Ingot – Bottom Block Separation Distance (mm)</th>
<th>Magnitude of Wetting Factor</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>500</td>
<td>X &lt; 5</td>
<td>0</td>
</tr>
<tr>
<td>5</td>
<td>480</td>
<td>5</td>
<td>0</td>
</tr>
<tr>
<td>9</td>
<td>65</td>
<td>10</td>
<td>1</td>
</tr>
<tr>
<td>10</td>
<td>50</td>
<td>X &gt; 10</td>
<td>1</td>
</tr>
<tr>
<td>10 &lt; X &lt; 50</td>
<td>50</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
Bottom Block Exterior Faces (Surfaces $\Gamma_5$ and $\Gamma_5$)

The heat transfer coefficient on vertical faces of the bottom block was varied to account for the geometry as shown in Figure 4.5. Although industrially, the bottom block has both slanted and vertical exterior faces, the modeled exterior faces were vertical only, for simplicity. Above the bottom Point C, the bottom block is water cooled and the heat transfer coefficient was taken from the boiling water curve. Below this point, the free-falling water falls away, and the bottom block is cooled by the surrounding air. In this case, the heat transfer coefficient was assumed to be constant at 25 W m$^{-2}$K$^{-1}$. The bottom surface of the bottom block was also air cooled at a rate of 25 W m$^{-2}$K$^{-1}$.

![Figure 4.5](image)

**Figure 4.5** - Actual (a) and Modeled (b) x-z cross section of the bottom block showing areas of different heat transfer.
4.4.3 Mechanical Boundary Conditions

The displacement boundary conditions are tabulated in Table 4.4. These conditions restrict out-of-plane motion along the symmetry planes of the ingot and bottom block, and also restrict vertical separation of the ingot and bottom block at the centreline. Applying these boundary conditions avoids rigid body motion, and aids convergence issues in the mathematical solution.

Table 4.4 - Displacement Boundary Conditions for Stress Analysis

<table>
<thead>
<tr>
<th>Model Surface</th>
<th>Boundary Condition</th>
</tr>
</thead>
<tbody>
<tr>
<td>Nodes on ingot and bottom block x-z symmetry face</td>
<td>Constrained in y-direction</td>
</tr>
<tr>
<td>Nodes on ingot and bottom block y-z symmetry face</td>
<td>Constrained in x-direction</td>
</tr>
<tr>
<td>Node at the bottom centre of the bottom block</td>
<td>Constrained in all three directions</td>
</tr>
<tr>
<td>Node at the bottom centre of the ingot</td>
<td>Constrained to the bottom block via a spring element to resist separation [46].</td>
</tr>
</tbody>
</table>

4.5 Summary

The domain, geometry, material properties, and boundary conditions for the 3D coupled thermal – displacement model of the ingot DC casting process have been presented. Great care has been taken to ensure that the material has the ability to behave both as a liquid and a solid by identifying a mechanical coherency temperature. Above 580 °C, aluminum alloy AA5182 behaves as a viscous liquid and can easily deform without accumulating plastic or thermal strain. Below this temperature, the metal behaves as a solid with elastic, plastic and thermal strain response to an applied stress or change in temperature.

In D.C ingot casting, there are a number of process parameters that affect the quality of the ingot. For a fixed mold and bottom block geometry, the pour temperature, casting speed and water flow rate are the easily adjustable variables within industry. These variables are also easily adjustable in this model, and so their effect on the temperature, stress, and strain fields can be evaluated. This represents a significant and exciting advancement as it allows the simulation and analysis of multiple casting recipes as compared to expensive and time-consuming experiments.
Chapter 5: Hot Tearing Criteria Implementation

5.1 Introduction

As described in Chapter 2 – Literature Review, a number of criterion have been developed to assess hot tearing susceptibility. To be considered successful, the criterion should be able to relate the thermal and mechanical predictions produced by the ingot DC casting model to hot tearing defects observed in industry. In the current work, two criterion were used to predict hot tears in the ingot DC casting process: the Rappaz – Drezet – Gremaud\(^{[7]}\) (RDG) pressure based criterion and Pellini's Total Strain\(^{[26]}\) criterion. The RDG criterion was chosen because (a) it takes into account both the thermal and mechanical effects on the formation of hot tears, (b) it appears to have a sound mechanistic and physical basis as it is simply a solution to the mass balance equation and (c) it has been successfully incorporated previously in axisymmetric billet finite element models to predict hot tear formation in aluminum billets\(^{[33]}\). The Total Strain criterion was chosen because preliminary casting simulations using the ingot DC casting model showed an area of high plastic strain in the region vulnerable to hot tearing for some casting recipes.

In this chapter, the technique used to implement the RDG criterion into the ingot casting model is outlined. In the current context, implementation of the RDG criterion is defined to be the methodology used to calculate the mushy zone depression pressure. The Total Strain criterion requires no implementation strategy, since it is simply a comparison between the strains during solidification and some critical values.

The basic premise of the RDG criterion is that if \(P_{\text{mush}} < P_{\text{critical}}\) then hot tearing will occur\(^{[7]}\). In the absence of pores, the relative pressure in the mushy zone, \(P_{\text{mush}}\), is given by Eq. (5.1):

\[
P_{\text{mush}} = P_{\rho gh} - (\Delta P_{\text{mech}} + \Delta P_{\text{shrink}})
\]  

Unfortunately, the value of \(P_{\text{critical}}\) is unknown and not easy to estimate. The hot tearing model developed in this work will therefore use the value of the pressure drop contributions related to the deformation of the solid skeleton, and the shrinkage associated with the phase change \((\Delta P_{\text{mech}} + \Delta P_{\text{shrink}})\) to assess hot tearing susceptibility. The larger the value of this pressure drop, the more prone a location is to hot tearing. If the pressure drop is negative, then the mushy zone pressure is higher relative to atmospheric pressure, and hot tears should not form.
5.2 Governing Equations for Depression Pressure

To calculate the deformation and shrinkage pressure drop, a mass balance can be performed on a small volume of material within the mushy zone as per Eq. (2.10). Making the assumption that the liquid velocity is in the direction of the heat flow only (parallel to the gradient), while the solid deformation occurs perpendicular to the heat flow, then the mass balance can be described by Eq. (5.2):

$$\frac{\partial}{\partial t}(\rho_1 f_1 + \rho_s f_s) + \frac{\partial}{\partial x}(\rho_1 f_1 v_{1,x}) + \frac{\partial}{\partial y}(\rho_s f_s v_{s,y}) = 0 \tag{5.2}$$

where $v_{1,x}$ and $v_{s,y}$ are the liquid and solid velocities in their respective directions. Starting from (5.2), Rappaz et al.\(^7\) used the Darcy Equation to show that the pressure drop in the mushy zone due to deformation and shrinkage could be calculated as per Eq. (5.3):

$$\Delta P_{\text{mech}} + \Delta P_{\text{shrink}} = \beta \mu \int_x^L \frac{F}{K} \, dx + (1 + \beta) \mu \int_x^L \frac{E}{K} \, dx \tag{5.3}$$

where $\beta$ is the alloy volumetric shrinkage associated with the phase transformation, given by $\beta = \left( \frac{\rho_s}{\rho_l} - 1 \right)$, and $\mu$ is the liquid viscosity. $E$ and $F$ represent the integral of the fraction solid times the strain rate and the integral of the rate of solid evolution, as per Eqs. (5.4) and (5.5):

$$E = \int_0^x f_s \varepsilon_p \, dx \tag{5.4}$$

$$F = \int_0^x \frac{df_s}{dt} \, dx \tag{5.5}$$

where $\varepsilon_p$ is transverse strain rate, $f_s$ is the fraction solid with respect to position within the mush, and $\frac{df_s}{dt}$ is the evolution of fraction solid. $K$ is the permeability of the mush and is given by the Kozeny-Carman relationship, as per Eq. (5.6):

$$K = \frac{\lambda_2^2 (1 - f_s)^3}{180 \frac{f_s^2}{f_s^2}} \tag{5.6}$$

where $\lambda_2$ is the secondary dendrite arm spacing. The limits of integration are defined such that zero is near the solidus, $x$ is an arbitrary distance from the solidus, and $L$ is the length of the mushy zone.

The RDG criterion can be further simplified by assuming a constant temperature gradient $[G]$, cooling rate $[\dot{T}]$ and strain rate $[\dot{\varepsilon}]$ along the length of the mushy zone. In this
fashion, as was originally proposed for the RDG criterion\textsuperscript{[7],[33]}, the integrals over distance are replaced by integrals over temperature, assuming \( dx = \frac{dT}{G} \), with the limits of integration changed from spatial to thermal: \( 0 \rightarrow T_{cg}, L \rightarrow T_i \).

The effect of the transformation from a spatial to a temperature integral is to change the hot tearing criterion from spatially variant to invariant. In the spatially variant solution, the depression pressure is calculated over the actual mushy zone at a particular time increment and so the spatial variation in quantities can be captured. This allows one to incorporate the variation in the cooling rate and strain rate in the mushy zone that is predicted by the finite element model. In the spatially invariant solution, the depression pressure is calculated by assuming that the solidification conditions such as cooling rate and strain rate along the length of the mushy zone match the conditions at the solidifying location. By assuming a constant temperature gradient, both the assumed mushy zone length and temperature profile can be derived.

The invariant form has been applied extensively in the billet casting case. It has worked well in predicting hot tearing at the centreline of the billet, just above the bottom block. This form of the RDG criterion can be used in billet casting because the assumptions of constant temperature gradient, cooling rate and strain rate are more-or-less valid. Although the spatially variant form represents a major improvement over the invariant solution, it has not received much attention in the literature. In contrast to the spatially variant form, the invariant form is fairly easy to implement into a DC casting process model. Furthermore, using the invariant form requires that one use a “standard” process model containing only limited knowledge of the constitutive behaviour in the semisolid phase. This is because literature information currently exists on approximate strain rate values at the end of solidification for billet casting and this is the only strain rate value required for this solution to the RDG criterion. For the spatially variant solution, much knowledge on semisolid constitutive behaviour is required to create an accurate profile of the mushy zone strain rate. As discussed in Section 4.4 - Thermal and Mechanical Properties of AA5182, the constitutive behaviour of the semisolid material is highly temperature dependent. Only at medium and high fraction solids does the concept of plastic strain and strain rate apply, since at low fraction solid, the material behaves as a liquid. Furthermore, a well developed and validated process model is required to ensure proper coupling of the effects of temperature and stress-strain.

In ingot casting, hot tearing occurs near the surface of the ingot. At the surface, the
cooling rate is typically 15 - 30 °C s\(^{-1}\). Further inward, the cooling rate drops to less than 2 °C s\(^{-1}\).

In terms of stress and strain behaviour, the ingot surface during solidification is under tensile stress, while much of the interior is generally under compression. Clearly, in ingot casting, both the cooling rate and strain rate vary considerably along the mushy zone near the surface of the ingot. Under these circumstances, the spatially variant form of the RDG criterion should be used.

### 5.3 Calculation Methodology

In the current work, the RDG Criterion was implemented into the ingot DC cast model using the spatially variant solution to the mass balance. To determine the depression pressure according to the RDG criterion, the plastic strain rate, \(\dot{\varepsilon}^{pl}\), fraction solid, \(f_s\), and solidification rate, \(\frac{df_s}{dt}\), must be known over the mushy zone length. These parameters were derived from the ingot DC casting model simulation by outputting the temperature field and the strain rate tensor to the ABAQUS output database for each time increment. The heat flux vector was also outputted to the database, as it was needed to calculated the path for integration.

To calculate the limits of integration, the temperature dependent fraction solid curve experimentally determined by Thompson\(^{[52]}\) was used. According to this data, the liquidus temperature for AA5182 is 637 °C, and the solidus temperature is 509 °C. However, the last one to two percent of liquid solidifies over quite a large range. In this region, dendrite bridging has already occurred and hot tears are unlikely to form as the structure will support appreciable mechanical loading. Rappaz et al.\(^{[7]}\) assumed that the peak depression pressure occurs at a fraction solid of 0.98 for Al – Si alloys, and called the corresponding temperature the grain coalescence temperature, \(T_{cg}\). The same assumption was be made in the current work with \(T_{cg} = 553 \) °C for the AA5182 alloy.

A PYTHON code was written to extract the process variable information from the casting model and then to calculate the deformation and shrinkage depression pressure. PYTHON is a module based open source code that is available within the ABAQUS post-processor. The script was written to access the thermal history of every node in a sequential manner, to determine the time increment when the temperature of the node currently being processed fell below \(T_{cg}\). At that time increment, the heat flux at the current node was outputted from the database. Using the
magnitude of heat flux in the $x$, $y$, and $z$ directions at the current node, the angle of heat flow relative to horizontal was calculated. The code then projected a straight line through the mushy zone along the angle of heat flow from the current node towards the centre of the ingot, to find all the element intersections (vertex, line or face) that lay on this projection. The position, temperature, cooling rate, and strain rate at each of these intersection points was then acquired from the output database. All data points further towards the centre of the ingot than the intersection point whose current temperature was equal to the liquidus temperature were discarded. The intersection points that remain corresponded to the mushy zone of the current node. The mushy zone vector for node 2807 is shown in Figure 5.1.

This scenario creates a snapshot in time of the current thermal and strain rate conditions along a path in the mushy zone from the $T_{liq}$ to $T_{cg}$ for the node currently being processed, as shown in Figure 5.2. From this snapshot in time, the variation in depression pressure along the mushy zone length was determined. The profile of the solidification sequence was approximated by using Thompson's fraction solid curve for AA5182\textsuperscript{[52]}. Since most ingot hot tears develop parallel to the casting direction, the strain rate required for the RDG criterion is the transverse strain rate, $\dot{\varepsilon}_{tr}$, defined as the strain rate perpendicular to both the heat flux vector and the

![Figure 5.1 - Thermal profile of a cross section of the ingot, in the centre of the rolling face. The white arrow indicates the mushy zone path and element intersection points used to calculate the total pressure as node 2807, on the ingot surface, reached $T_{cg}$.](image-url)
Chapter 5: Hot Tearing Criteria Implementation

Figure 5.2 - Example plot of the temperature and transverse strain rate variation along the mushy zone length that was used to calculate the total pressure for node 2807 using the spatially variant solution method.

casting direction. Following the assumptions made by M'Hamdi et al.[54], only the plastic contribution to the strain rate was used. The form of the calculation is shown in Eqs. (5.7) and (5.8):

\[
\begin{align*}
\mathbf{n}_r & = [\mathbf{HF}] \times [\mathbf{CAST}] \\
[\ddot{\varepsilon}_r] & = [\ddot{\varepsilon}_x] [\mathbf{n}_r] + [\ddot{\varepsilon}_y] [\mathbf{n}_y] + [\ddot{\varepsilon}_z] [\mathbf{n}_z]
\end{align*}
\]  
(5.7)  
(5.8)

where \([\mathbf{HF}] \times [\mathbf{CAST}]\) is the cross product of the heat flux vector at \(T\), and the vector denoting the casting direction, and \([\mathbf{n}_r]\) is the vector representing the transverse direction, \([\ddot{\varepsilon}_r]\) is the plastic strain rate vector in the \(x\), \(y\) or \(z\) direction, and \([\ddot{\varepsilon}_r]\) is the transverse plastic strain rate.

The depression pressure was calculated using the trapezoid rule according to Eqs. (5.9), (5.10), and (5.11):

\[
F_i = \frac{1}{2} \left( \frac{df}{dT} \hat{T}_i + \frac{df}{dT} \hat{T}_{i-1} \right) \cdot (x_i - x_{i-1}) + F_{i-1}
\]

(5.9)
\[ E_i = \frac{1}{2} \left( f_{s,i} \dot{\varepsilon}_\mu + f_{s,i-1} \dot{\varepsilon}_p \right) (x_i - x_{i-1}) + E_{i-1} \]  

(5.10)

\[ \Delta P_i = \frac{1}{2} \left[ \beta \mu \left( \frac{F_i}{K_i} + \frac{F_{i+1}}{K_{i+1}} \right) + (1 + \beta) \mu \left( \frac{E_i}{K_i} + \frac{E_{i+1}}{K_{i+1}} \right) \right] (x_{i+1} - x_i) + \Delta P_{i+1} \]  

(5.11)

where \( x_i \) is the location of the \( i^{th} \) intersection point along the mushy zone line projection. Note that \( F_i \) and \( E_i \) are evaluated by integrating from \( x = 0 \) to \( x = L \) whereas \( \Delta P_i \) is evaluated by integrating from \( x = L \) to \( x = x_i \). This arises from the boundary conditions – i.e. the velocity is 0 at \( x = 0 \) (\( f_s = 0.98 \)) and \( \Delta P_i = 0 \) at the liquidus. In the permeability calculation, \( \lambda_2 \) was held constant, and assumed to be 100\( \mu \text{m} \). The calculated depression pressures for selected nodes were then outputted for further analysis.

5.4 Summary

In this chapter, the derivation of the RDG criterion, and its implementation into the ingot DC casting model have been discussed. In the ingot case, both the temperature profile and strain rate profile along the length of the mushy zone are non-linear, as shown in Figure 5.2. The two assumptions required for the invariant form of the RDG criterion – constant mushy zone temperature gradient and strain rate clearly do not apply. Using the spatially variant solution of the RDG criterion for ingot castings represents a major improvement over the earlier work in billet castings that use the invariant solution, since it is able to capture the large variation in the solidification conditions along the mushy zone seen in the ingot DC casting process.

A PYTHON code has been written to extract solidification and strain rate information from the ingot DC casting model and then to calculate the deformation and shrinkage depression pressure. The calculated pressures were outputted to a database, and the hot tearing susceptibility was then assessed.
Chapter 6: Hot Tearing Predictions in the ingot DC Casting Process

6.1 Introduction

With the description of an FE model of the ingot DC casting process completed in Chapter 4 – DC Casting Model Formulation, and a description of the implementation strategy for the RDG and Total Strain hot tearing criteria, completed in Chapter 5 – Hot Tearing Criteria Implementation, attention can now be returned to the problem of hot tearing in ingot castings. Our industrial partner, Alcan, has provided data for two castings; one which was cast under conditions prone to hot tearing, and the other which was not. These casting recipes will be used to relate the temperature and stress-strain fields, the RDG Criterion\(^\text{[7]}\), the Total Strain Criterion\(^\text{[26]}\), and hot tearing.

6.2 Hot and Cold Cast Recipe Information

The two sets of process parameters provided by Alcan were named 'cold cast', and 'hot cast', to reflect the amount of heat extracted by the cooling water in the start-up phase. The cold cast is characterized by a high flow rate of cooling water during the start-up phase. This promotes little or no water ejection on the ingot surface, large butt curl, and significant amounts of water incursion between the ingot and the bottom block. In the hot cast, the cooling water flow rate was reduced during the start-up phase by 25 % as compared to the cold cast (shown in Figure 6.1), and the bottom block filling time was reduced by 20 s. These changes, particularly the decrease in cooling water flow rate, have a significant effect on heat transfer during the start-up phase by promoting water ejection during secondary cooling, and reducing butt curl (which leads to little water incursion). The other process parameters were identical for the two casting recipes, and so in the steady-state phase, the thermal, stress and strain fields in the two castings should also be similar.

The predicted evolution in temperature 50 mm above the bottom block at the centre of the rolling face, and the evolution in butt curl measured by experiments\(^\text{[31]}\) for both casting recipes is shown in Figure 6.2. From Fig. 6.2(a), it can be clearly seen that the evolution in temperature during Primary Cooling is the same for both casting recipes. However, during Secondary Cooling, once the free-falling cooling zone is reached, the lower cooling water flow rate in the hot cast promotes water ejection. This decreases significantly the magnitude of the heat transfer...
coefficient, and causes the surface of the ingot to reheat. In the cold cast, without water ejection, the surface of the ingot continues to cool quickly during the free-falling cooling zone. As shown in Fig. 6.2(b), the hot casting conditions resulted in less than 8 mm of butt curl, whereas the cold cast resulted in approximately 40 mm of butt curl, as measured at the centre of the narrow face.

It is the cold cast which is susceptible to hot tearing. Using this casting recipe, hot tears are found to occur in the middle third of the rolling face immediately above the ingot lip, during the start-up phase of the casting process, and propagate in the vertical direction.

![Figure 6.1 - Normalized secondary cooling water flow rate as a function of time for the hot cast (ejection) and cold cast (no ejection) recipes.](image)
Figure 6.2 - Experimentally measured butt curl in the hot and cold castings\textsuperscript{[b]}. Note that the data for the hot cast stops at 240 s because of experimental difficulties.
6.3 Temperature and stress-strain Predictions in the Hot and Cold Casts

To examine the effect of casting recipe on the temperature and stress-strain predictions, the casting model was used to simulate casting for 800 s using the cold cast and hot cast recipes. To facilitate comparison, four locations on the rolling face surface, named Points A, B, C, and D, have been selected for outputting results from the model and are shown in Figure 6.3.

6.3.1 Results from the Cold Cast Simulation

Figure 6.4 shows a comparison of the predicted evolution in temperature and plastic strain $\varepsilon^{\text{yy}}$ at the centreline of the rolling face (Point A) and outside the middle third of the rolling face (Point B), 50 mm from the ingot lip. The $y$-orientation of plastic strain has been used since hot tears form on the rolling face, and parallel to the rolling direction. Point A is in the region where hot tears are prone to occur while Point B is outside this region. In Fig. 6.4(a), the data is presented over the full 800 s of simulation. The temperature results predicted for the two locations indicate that there is little variation in cooling across the rolling face during casting. However, there seems to be a great deal of variation in plastic strain, as Point B begins to deform in tension but then is exposed to compressive plastic strain, while Point A develops more tension and is exposed to only a small amount of compressive plastic strain. In Fig. 6.4(b), the time axis is constrained so that only the plastic strain accumulated above the solidus temperature is shown. Both Point A and Point B undergo tensile plastic strain during solidification. However, the magnitude of tensile plastic strain at Point A is approximately 40% larger than Point B.

In Figure 6.5, the predicted evolution in stress $\sigma_{\text{yy}}$ during the solidification interval is shown for both Point A and Point B. These values indicate that there is about a 15% difference in stress between these two points at the end of solidification. So, although as per Figure 6.4, the local cooling conditions at Point A and Point B are identical during solidification, both the $y$-orientation stress and strain are significantly higher at Point A. The thermal strain associated with the cooling conditions would provide a local value of stress and strain, which would be the same at Point A and Point B. Superimposed on top of the local value is the macroscopic effect of the entire body coming to equilibrium. The butt curl deformation, which bows the base of the ingot in a similar fashion to a beam in bending, would superimpose a large value of tensile stress and strain at the centre of the rolling face, and a smaller value out towards the edges, consistent with what is predicted by the model. During the solidification of Point A and Point B, the
magnitude of the butt curl increases from 0 mm to approximately 30 mm. The observations made with respect to Figure 6.4 and Figure 6.5, i.e. high stress and tensile plastic strain during solidification at Point A as compared to Point B, is fully consistent with hot tearing observed industrially, as these defects are found to form in the middle third of the rolling face of the ingot.

Figure 6.6 shows the predicted evolution in temperature (a) and plastic strain $\varepsilon_{yy}$ (b) during solidification at the centreline of the rolling face at different heights above the ingot lip (Points A, C, and D). Point A (50 mm from the ingot lip) experiences the start-up phase of the process, and is greatly susceptible to hot tearing. Point C (250 mm) is still in the start-up phase, but is less susceptible to hot tearing. Point D (500 mm) is almost in the steady-state phase and is not susceptible to hot tearing. As can be seen in Fig. 6.6(a), the cooling rate at the three heights is not the same. The plot of evolution in temperature for Point A contains a 75 s pause in cooling near 590 °C, and also a significant thermal recalescence. The thermal history at Point C and Point D, which are closer to the steady-state phase, are similar with no pause in cooling at 590 °C. Instead, the cooling rate is shallow between 590 °C and 570 °C, with only a small thermal recalescence. The pause in cooling is significant, since the temperature 590 °C corresponds to a fraction solid of about 0.85. Thus, Point A spends much time in the region vulnerable to hot tearing, while Point C and Point D do not. As can be seen in Fig. 6.6(b), the accumulation of plastic strain during solidification is quite different for the three locations. The largest tensile plastic strain during solidification occurs at Point A, and the smallest occurs at Point D. At Point C, there is a large accumulation of compressive strain, because Point C is above the neutral axis created by the global bending effect of butt curl. At Point D, tensile strain is accumulating only because of the local solidification conditions, as it is too far away from the ingot lip to be much affected by the butt curl. Furthermore, during the time that Point D solidifies, the magnitude of the butt curl does not increase at all. These observations are also fully consistent with hot tearing observed industrially as the defects are found only in the start-up phase of the process.

In Figure 6.7, the predicted variation in accumulated plastic strain during solidification as a function of both distance from the ingot lip (between Points A and D), and distance from the rolling face centreline just above the ingot lip (between Points A and B) is shown. While this graph is a little difficult to interpret due to the different length scales, there are two key observations: firstly, the accumulated plastic strain in the vertical direction reaches a maximum near Point A, before quickly becoming compressive higher up the casting, and secondly, just
above the ingot lip, the plastic strain during solidification is again largest at Point A, the centreline, and then gradually decreases moving outwards towards Point B. From this figure, it is clear that hot tears should only occur in the middle of the rolling face, in the start-up phase.

6.3.2 Comparing Temperature and Stress - Strain in the Two Casting Recipes

A comparison of the predicted evolution of temperature and strain at Point A between the hot cast and the cold cast is shown in Figure 6.8. In Fig. 6.8(a), the data is presented over the full 800 s of simulation. The temperature predictions show that in the initial stages of cooling, both Point A - Hot Cast (A-HC) and Point A - Cold Cast (A-CC) cool at the same rate. However, at about 200 s, immediately as this region exits the impingement zone, water ejection begins to occur on the rolling face of the hot cast, due to the low cooling water flow rate. The low cooling rate associated with water ejection causes Point A-HC to reheat to a temperature above the solidus temperature, and stay above the solidus for some time. Some remelting of the ingot may occur. The effect of low cooling water flow rate and water ejection on plastic strain is significant. Point A-HC initially develops moderate tension, then is exposed to compressive strain and finally is re-loaded in tension, whereas Point A-CC quickly attains a large tensile strain, which then gradually decreases over time. In Fig. 6.8(b), temperature and plastic strain in the solidification interval is examined. As can be seen from the plots of evolution in temperature, both Point A-HC and Point A-CC cool at the same rate during the entire solidification sequence. However, in contrast to the large amount of plastic strain accumulated during solidification at Point A-CC, the total plastic strain accumulated at Point A-HC is almost zero. This is caused by the butt curl, which has reached a value of ~28.0 mm when Point A-CC reaches the solidus temperature, but has only reached a value of ~6.0 mm when Point A-HC reaches the solidus temperature.

In Figure 6.9, the evolution in stress $\sigma_{yy}$ during the solidification interval is shown for both Point A-HC and Point A-CC. In Fig 6.9(a), the values during the solidification interval are shown. As can be seen, the stress predictions for the two casting recipes indicate that the tensile thermal stress during solidification is significantly larger in the cold cast as compared to the hot cast. At later times after solidification, shown in Fig. 6.9(b), the Point A in the hot cast is exposed to large tensile stresses while Point A in the cold cast is in compression. However, this region has now cooled to about 100 °C and has significant strength. The observations made in
**Figure 6.8** and **Figure 6.9** show that both stress and strain during solidification are reduced in the hot cast start-up phase as compared to the cold cast. This is because the contributing factors towards the development of stress and strain within the ingot—thermal gradient near the ingot surface, and butt curl, are also reduced in the hot cast due to the reduced cooling water flow rate.

**Figure 6.10** shows a comparison of the predicted evolution of temperature and plastic strain between the hot and cold casts at Point D, in the steady-state phase. As expected, these predictions are similar since the steady-state cooling water flow rates are the same in the two casting recipes and hence the thermal and stress-strain fields should also be similar.

**Figure 6.11** is a plot similar to **Figure 6.7**, with the addition of the “horizontal” and “vertical” variation in plastic strain at the end of solidification for the hot cast. The magnitude of the plastic strain predictions along the rolling face at 50 mm (horizontal plot) clearly shows that the strain accumulated during solidification in the start-up phase of the cold cast is much larger as compared to the hot cast on the rolling face of the ingot. The plastic strain in the “vertical” direction is also greater in the cold cast in the start-up phase of casting, around Point A. As the steady-state region is approached, the plastic strain during solidification is similar for the two simulations, because the same steady-state cooling water flow rate is being used.

### 6.3.3 Insight into Hot Tearing from the Process Model

There is high hot tearing susceptibility in the cold cast, on the middle third of the rolling face, just above the ingot lip. The evolution of temperature, stress, and strain during solidification, calculated by the ingot casting model for Point A, B, C, and D, is shown in **Figures 6.4 – 6.11**, and provides insight into why hot tearing occurs in only this relatively small region of the cold cast, not in the rest of the ingot, and also not at all in the hot cast.

The presented results have shown that for the cold cast, the magnitude of the tensile plastic strain $\varepsilon_{\text{pl}}$, during solidification is largest in the centre of the rolling face, just above the ingot lip, at Point A, because of butt curl and high cooling water flow rate. In comparison, the magnitude of the tensile plastic strain during solidification out towards the edge of the rolling face just above the ingot lip (Point B) and in the steady-state region (Point D) is quite small or zero. The plastic strain further up, but still in the start-up phase (Point C), is compressive as it is above the effective neutral axis where butt curl develops. In the hot cast, tensile strain during solidification in the region from Point A to Point B is almost zero. Further up the ingot, in the
region around Point D, the solidification plastic strain in the two simulations is comparable due to similarities in the steady-state process parameters. If it is assumed that hot tearing is a strain dominated effect, then clearly the region most susceptible to hot tearing would be in the cold cast, in the middle portion of the rolling face, just above the ingot lip. These defects are not found in the hot cast since there is very little plastic strain accumulated during solidification in the start-up phase.

Figure 6.3 - A diagram of one half of the rolling face of the ingot, marked with the locations of Points A, B, C, and D. The mesh is also shown.
Figure 6.4 - The cold cast simulated evolution in temperature and plastic strain $\varepsilon_{yy}^{pl}$ 50 mm above the ingot lip at the centreline of the rolling face (Point A) and 0.64m towards the edge (Point B). Data for the entire simulation (a) and above the solidus temperature (b) is presented.
Figure 6.5 - The simulated evolution in stress $\sigma_{yy}$ during solidification 50 mm above the ingot lip at Point A and Point B using the cold cast recipe.
Figure 6.6 - The cold cast simulated evolution of temperature (a) and plastic strain $\varepsilon_{yy}^{pl}$ (b) during solidification at the centreline of the rolling face for Point A, Point C and Point D.
Figure 6.7 - The variation in total plastic strain $\varepsilon_{yy}^p$ at the end of solidification as a function of position on the rolling face for the cold cast. To aid in interpretation, the location of Points A, B, C and D are shown.
Figure 6.8 - The simulated evolution in temperature and plastic strain $\varepsilon_{yy}^{pl}$ at Point A for both the cold cast and the hot cast. Data for the entire simulation (a) and above the solidus temperature (b) is presented.
Figure 6.9 - The simulated evolution in stress $\sigma_{yy}$ at Point A for both the cold cast and the hot cast.
Figure 6.10 - The simulated evolution in temperature and plastic strain $\varepsilon_{yy}^{pl}$ at Point D for both the cold cast and the hot cast.

Figure 6.11 - The variation in total solidification plastic strain $\varepsilon_{yy}^{pl}$ as a function of position on the rolling face for both casting recipes.
6.4 Hot Tearing Predictions Using a Total Strain Criterion

The Total Strain Criterion is simply a means of assessing the amount of strain accumulated during solidification over some critical interval. To use this criterion for the ingot DC casting of aluminum alloy AA5182, the value of the critical hot tearing strain must be known. Although some research has been conducted to determine the constitutive behaviour of alloys in the mushy zone, much of this work has been compression\textsuperscript{[22]} and / or shear experiments \textsuperscript{[23]}, whereas the Total Strain Criterion requires information about the tensile constitutive behaviour. There are many challenges in performing these experiments: achieving constant and known temperature, retaining the as-cast microstructure, confidence in the fraction solid with temperature relationship, and measuring the strain and strain rate.

Colley\textsuperscript{[55]}, has recently measured the tensile properties of DC cast AA5182 at temperatures between 500 °C and 580 °C, at a range of strain rates between approximately $10^{-2}$ s$^{-1}$ and $10^{-4}$ s$^{-1}$. The constitutive behaviour is shown in Figure 6.12, which plots the stress (a) and ductility (b) as a function of temperature. As can be seen from the stress results, AA5182 exhibits significant tensile strength at temperatures as high as about 570 °C - 575 °C, depending on the strain rate. Above this critical temperature, a sharp decrease in the tensile strength occurs. For strain, the results show that the alloy loses much of its ductility by about 540 °C, and at a temperature between 565 °C – 570 °C, the magnitude of the ductility becomes insignificant. Thus, there exists a temperature band between about 540 °C and 575 °C in which the alloy contains some strength but little ductility. Furthermore, the stress – ductility results also show that the mushy zone constitutive behaviour of AA5182 is extremely sensitive to strain rate.

The work by Colley satisfies the requirements for a hot tearing critical strain: tensile ductility measurements in the mushy zone, and D.C ingot as-cast microstructure. Figure 6.13 shows a plot of the variation in plastic strain $\varepsilon^p$, with temperature for the cold cast at Point A and Point B, and the hot cast at Point A, along with a plot of the strain to failure with temperature measured by Colley for a strain rate of 0.002 s$^{-1}$. Point A and Point B in the cold cast were chosen because they accumulate large plastic strains during solidification, while Point A in the hot cast was chosen for comparison purposes. Also, the strain rate of 0.002 s$^{-1}$ was chosen for its relevance to ingot DC casting. From these results, it would appear that the cold cast conditions at Point A have led to total strains that are similar to or slightly larger than the ductility limit at temperatures above 575 °C. At Point B and Point A-HC, the total strains are below the ductility limit at temperatures above 575 °C. All the predicted total strains below 575 °C are less than the
ductility limit of the alloy at all three points. Therefore, the Total Strain Criterion approach seems to show conclusively that hot tearing will occur in the cold cast, in the middle third of the rolling face just above the ingot lip, and will not occur in the hot cast.

However, a number of issues should be raised with respect to this work: firstly, the ductility data measured by Colley is at constant strain rate and temperature whereas the plastic strains predicted by the model represent strains accumulated over a range of strain rate and temperature; secondly, the ductility measurements were only made at six temperatures between 500 °C and 580 °C, so more data is needed; and thirdly, the underlying assumption with the constitutive data is that the fraction solid – temperature relationship during reheating is the same as during cooling. These issues highlight the need for a much better understanding of the constitutive behaviour of alloys in the mushy zone. Performing further constant temperature experiments would help to better define the decrease in ductility at high fraction solid, and the temperature at which the ductility becomes small (currently 540 °C). This temperature is the lower limit for hot tearing susceptibility. Also missing is constitutive data relating to the properties above 580 °C. This would help define the upper limit for hot tearing, and the temperature at which mass feeding can begin to accommodate tensile strain.
Figure 6.12 - The variation in tensile strength (a) and ductility (b) for AA5182 with temperature in the mushy zone at various strain rates [55].
Figure 6.13 - Comparison between the predicted variation in strain with temperature and measured strain-to-failure data by Colley\textsuperscript{[55]} (strain rate = 0.002 s\textsuperscript{-1}) at Point A and Point B for the cold cast and Point A for the hot cast.
6.5 Hot Tearing Predictions Using the RDG Criterion

Recall that the RDG criterion calculates a pressure drop based on a mass balance in the liquid, and is an integration of the plastic strain rate and solidification rate conditions in the mushy zone, from the solidifying point straight back towards the fully liquid region. The theory is that with larger pressure drops comes increased pore formation, and ultimately, hot tearing. To examine the effect of casting recipe on hot tearing susceptibility using the RDG criterion, the temperature and plastic strain-rate predictions from the hot and cold cast simulations were used. The results at the critical regions of the rolling face are again discussed with the respect to the behaviour predicted at Points A, B, C, and D.

6.5.1 Results from the Cold Cast

The depression pressure in the cold cast was calculated for a number of nodes on the rolling face of the ingot. The variation in total ($\Delta P_{\text{total}}$), thermal ($\Delta P_{\text{shrink}}$), and mechanical ($\Delta P_{\text{mech}}$) depression pressure between Point A and Point B in the start-up phase is shown in Figure 6.14. In Fig. 6.14(a), all three depression pressures are plotted so that their relative magnitudes can be compared. From these predictions, a number of key observations should be made. Firstly, $\Delta P_{\text{total}}$ appears to be fairly constant along the rolling face, as the standard deviation of the values between Point A and Point B is approximately 13% of the mean value. The same observation can be made for $\Delta P_{\text{shrink}}$, with a standard deviation of only 10% of the mean. Secondly, the value of $\Delta P_{\text{mech}}$ is quite small, only about 10% as compared to $\Delta P_{\text{shrink}}$. Thirdly, $\Delta P_{\text{mech}}$ appears to also be fairly constant along the rolling face. In Fig. 6.14(b), only the mechanical depression pressure is shown. From this plot, it appears that $\Delta P_{\text{mech}}$ is fairly constant in the middle one-half of the rolling face, from Point A out to a distance of 0.5 m, and then increases towards the edge of the rolling face.

Figure 6.15, shows fraction solid and cooling rate profiles (a) and strain rate profile (b) along three vectors in the mushy zone of the cold cast. Mushy A defines a vector beginning at Point A, and continuing in a direction defined by the thermal gradient until the liquidus temperature is reached. The time increment corresponds to the time when Point A is at a fraction solid of 0.98. Mushy B, and 0.3 m, define similar vectors, beginning at Point B, and 0.3 m from the rolling face centreline. This figure can be used to provide further insight into the cold cast depression pressure on the rolling face just above the lip. Fraction solid and cooling rate are the
critical parameters for $\Delta P_{\text{shrink}}$, and the predicted mushy zone profiles at Point A and Point B locations are shown in Fig. 6.15(a) to be similar. This agrees with the results in Figure 6.4 that showed that the evolution in temperature at Point A and Point B was approximately the same. Due to these similarities, one should expect the same type of results in Figure 6.14 – i.e. little variation in $\Delta P_{\text{shrink}}$ for a series of nodes on the rolling face between Point A and Point B. Strain rate is the critical parameter for $\Delta P_{\text{mech}}$. As can be seen in Fig. 6.15(b), the mushy zone strain rate for Mushy A, 0.3 m and Mushy B can be divided into three regions: no strain rate at small $f_s$, compressive strain rate at medium $f_s$, and tensile strain rate at large $f_s$. At low fraction solid, the constitutive behaviour of the semisolid is a viscous liquid, as described in Chapter 4, and so elastic and plastic strain accumulation is not possible.

Compressive and tensile strain rates exist in tandem because of thermal expansion and the deformation of the ingot as a whole. Whenever large bodies cool quickly, the strain related to the thermal gradient causes the surface to be in deformed in tension, and the interior, in compression. From the results predicted, it can be seen that the tensile strain rate at the surface of the ingot is the same for Point A, 0.3 m, and Point B. However, further into the mushy zone, the the compressive strain rate is much less for Mushy B as compared to Mushy A and 0.3 m, (which have the same compressive strain rate). The difference in compressive strain rate between Mushy A and Mushy B is caused by the macroscopic effect of the ingot coming to equilibrium. For Mushy A and 0.3 m, the effect of butt curl is being superimposed on the local thermally driven stress-strain conditions. Mushy B is in proximity to the corner of the ingot, and so both the butt curl and the corner are superimposing global effects on the local stress-strain conditions. This leads to a decreased compressive strain rate in the mushy zone. In terms of $\Delta P_{\text{mech}}$, it was observed in Figure 6.15 that $\Delta P_{\text{mech}}$ is fairly constant in the middle one-half of the rolling face, and then increases towards the edge of the rolling face. This matches well with the mushy zone strain rate plots in Fig. 6.14(b), as the area under the curve would be approximately the same for Mushy A and 0.3 m, but would be larger for Mushy B since the negative area in the compressive portion would be much smaller. The integration of larger areas will lead to larger depression pressures.

The observations made for Figure 6.14 are troubling with reference to hot tearing in the start-up phase. Since $\Delta P_{\text{shrink}} >> \Delta P_{\text{mech}}$, the results suggest that high hot tearing susceptibility above the ingot lip occurs predominantly because of the volumetric shrinkage associated with the phase change, not any mechanical loading. Hot tearing must be associated with the mechanical
loading term, because porosity is commonly linked with the shrinkage term. Furthermore, from the accumulated solidification plastic strain results shown in Figure 6.7, it was expected that $\Delta P_{\text{mech}}$ would decrease from Point A to Point B. The results show that the magnitude of $\Delta P_{\text{mech}}$ is relatively constant in the middle one-half of the rolling face (Point A to 0.3 m) and then increasing in proximity to the ingot corner (Point B and beyond). Thus, the mechanical depression pressure would have the largest effect on hot tearing susceptibility at the edge of the rolling face, and less of an effect in the middle of the rolling face, where hot tearing is prone to occur.

The variation in depression pressure at the centreline of the rolling face between Point A (start-up) and Point D (steady-state) is shown in Figure 6.16. A number of observations can be made from these pressure predictions. Firstly, the largest value of $\Delta P_{\text{total}}$ is at 0.2 m from the ingot lip. Secondly, the value of $\Delta P_{\text{total}}$ seems to be reaching a steady-state value between 0.3 m and 0.5 m from the ingot lip. Thirdly, the steady-state value of $\Delta P_{\text{total}}$ is much larger than $\Delta P_{\text{total}}$ immediately above the ingot lip. Fourthly, $\Delta P_{\text{shrink}}$ is much larger than $\Delta P_{\text{mech}}$. Lastly, the value of $\Delta P_{\text{total}}$ at Point C does not match the rest of the results. These observations, which clearly demonstrate that the region of high depression pressure is the steady-state phase, are not consistent with hot tearing, since hot tears form almost exclusively in the start-up phase. In Fig. 6.16(b), only the mechanical depression pressure is shown. This plot shows that $\Delta P_{\text{mech}}$ is positive just above the ingot lip, where hot tears are known to occur, then becomes negative further into the start-up phase. The negative sign means the relatively large mechanical depression pressure at 0.2 m and Point C term is hindering hot tear formation by decreasing the value of $\Delta P_{\text{total}}$. This portion of the curve has similar trends to the plot of total plastic strain accumulated during solidification, “vertical” direction, shown in Figure 6.7. As the steady-state phase is reached, $\Delta P_{\text{mech}}$ becomes positive again. The observations in the start-up phase are consistent with hot tearing as hot tearing occurs just above the ingot lip. However, the largest positive value is at Point D, in the steady-state phase.

Figure 6.17 (a) shows the predicted variation in cooling rate and fraction solid in the mushy zone for Point A, Point D, and a point 0.2 m from the ingot lip. The cooling rate curves at the three points show that at the surface of the ingot, the cooling rate is very high whereas further into the mushy zone, the cooling rate is quite low. Furthermore, there is significant difference in the surface (convective) cooling rate between the three points, as $T_1 = -16 \, ^\circ C \, s^{-1}$ and $T_{0.2m} = -32 \, ^\circ C \, s^{-1}$. The cooling rates at Point 0.2 m and Point D are equal. This is because the cooling
water flow rate in the start-up phase is increasing with time, and causes the cooling rate to increase from Point A to Point 0.2 m. At 0.18 m above the ingot lip, the flow rate becomes constant and so the cooling rate also reaches a steady-state value. In terms of fraction solid, Fig. 6.17(a) shows that the variation in \( f_s \) along the length of the mushy zone is fairly similar between the three points. From these predictions, one would expect that the magnitude of \( \Delta P_{shrink} \) would be smallest at Point A and would be larger at Point 0.2 m and Point D, with Point 0.2 m and Point D having similar values. However, the results have shown that \( \Delta P_{shrink} \) at Point 0.2 m is almost double the value at Point D. This would suggest that there are some problems in the methodology for calculating the depression pressure.

The depression pressure at Point C does not match the trends for the rest of the rolling face. This observation supports the theory that the depression pressure is being calculated incorrectly. In Figure 6.17 (b), the variation in fraction solid and \( \Delta P_{shrink} \) in the mushy zone is shown for Point 0.2 m and Point C. This plot helps to determine why the value of \( \Delta P_{shrink} \) at Point C is only half of the value of \( \Delta P_{shrink} \) at Point 0.2 m. The results from Fig. 6.17(a) indicated that cooling rate, fraction solid and thus \( \Delta P_{shrink} \) at Point 0.2 m and Point C should be approximately the same. However, as shown in Fig. 6.17(b), there is a large difference in the calculated \( \Delta P_{shrink} \). The fraction solid curves in Fig. 6.17(b) show that the pressure difference between Point 0.2 m and Point C is artificial, and occurs because the sub model calculating the depression pressure only has a few data points to make the integration. In both Point 0.2 m and Point C, four points along the mushy zone were used to calculate the depression pressure as the model can only interpolate strain rate and temperature at element intersections (faces, edges and corners). Note the difference in fraction solid between Point 0.2 m and Point C at the arrow, where the integration problem arises. In the solidification sequence proposed by Thompson, upwards of 30 % of the solid is formed in the first 5 °C below the liquidus temperature. So, although the temperature difference between Point 0.2 m and Point C is less than 1 °C, the fraction solid at Point C is 0.12 and the fraction solid at Point 0.2 m is only 0.01. This greatly affects the permeability in the mushy zone given by the Kozeny-Carman equation, and in turn will affect the depression pressure calculation. If more points were available along the mushy zone to make the integration, the fraction solid along the mushy zone would have better resolution and then both Point C and Point 0.2 m would contain integration points at \( f_s = 0.01 \) and \( f_s = 0.12 \).

The results presented with respect to Figure 6.17 lead to the conclusion that the
maximum in $\Delta P_{\text{shrink}}$ seen in Figure 6.16 at 0.2 m is probably artificial in nature. Once the cooling water flow rate reaches a constant value, $\Delta P_{\text{shrink}}$ should be similar at all locations further up the rolling face. It will also always be much larger than $\Delta P_{\text{mech}}$. This means that $\Delta P_{\text{shrink}}$ will increase through the start-up phase until it attains a constant value in the steady-state phase. The larger implication is twofold. Firstly, since $\Delta P_{\text{shrink}}$ will always be greatest in the steady-state phase, due to the higher cooling water flow rate, it cannot be used to accurately predict the location of hot tears in ingot DC casting, since the results will give the wrong trends. Secondly, the size of the mesh greatly affects the hot tearing susceptibility. Thus, for a coarse mesh, a number of points along the rolling face must be used to find an average value of depression pressure at any particular height above the ingot lip.

6.5.2 Comparing the Hot Cast and the Cold Cast Depression Pressures

Figure 6.18 shows a comparison of shrinkage and mechanical depression pressure along the rolling face 50 mm above the ingot lip (a) and a comparison of the total depression pressure at the centreline of the rolling face at various heights (b) for both the hot and cold casting recipes. In Fig. 6.18(a), the predictions clearly show that the magnitude of both $\Delta P_{\text{shrink}}$ and $\Delta P_{\text{mech}}$ is larger in the start-up phase of the cold cast as compared to the hot cast. The large value of $\Delta P_{\text{shrink}}$ in the cold cast start-up is caused by the high cooling water flow rate. This leads to the rolling face being cooled at a rate of 17 °C s$^{-1}$, whereas in the hot cast, the surface cooling rate is only 7 °C s$^{-1}$. Because of water ejection and limited butt curl, the plastic strain in the mushy zone of the hot cast is quite small. Hence $\Delta P_{\text{mech}}$ is also small. In Fig. 6.18(b), the trends indicate that in the start-up phase, the cold cast $\Delta P_{\text{total}}$ is much larger than the hot cast $\Delta P_{\text{total}}$, and in the steady-state phase, $\Delta P_{\text{total}}$ is of similar value for the two casting recipes. As shown in Figure 6.1, the cooling water flow rate for the two casting recipes is the same during steady-state. Also, as shown in Figure 6.10, the evolution in temperature and plastic strain at Point D is the same in the hot and cold casts. It then would then be expected that the steady-state depression pressures would also be similar.
6.5.3 Insight into Hot Tearing from the RDG Criterion

The RDG criterion has been applied to the hot cast and cold cast simulations. With respect to the cold cast, the predicted depression pressures do not demonstrate why hot tearing occurs only in the middle third of the rolling face in the start-up phase. As shown in Figure 6.14, $\Delta P_{\text{total}}$ seems to be constant across the rolling face just above the lip. As shown in Figure 6.16, $\Delta P_{\text{total}}$ in the steady-state phase is much larger than in the region susceptible to hot tearing. Both of these results occur because the magnitude of $\Delta P_{\text{shrink}}$ is much larger than the magnitude of $\Delta P_{\text{mech}}$. One underlying assumption with the RDG criterion is that porosity and hot tearing are related. If this were the case in ingot DC casting, then in addition to hot tearing, porosity should be an issue on the rolling face surface, especially in the steady-state region because $\Delta P_{\text{shrink}}$ is so large. Industrial experience has shown that porosity is not an issue for ingot casting. Since the RDG criterion predicts pore formation, a second interpretation could be that the number of hot tearing nucleation sites is greatest in the steady-state phase, but as shown in Figure 6.7, there is little stress-strain effects to produce a hot tear. Although the number of pores is less in the start-up phase, there is much more tensile strain during solidification. These interpretations suggest that only the term $\Delta P_{\text{mech}}$ be used. However, $\Delta P_{\text{mech}}$ is constant in the start-up phase in the region prone to hot tearing, and does not give any insight into why hot tears are most prone at the centreline. Furthermore, the steady-state value of $\Delta P_{\text{mech}}$ is much larger than in the start-up phase. Using only the $\Delta P_{\text{mech}}$, one would predict the highest hot tearing susceptibility to be in the steady-state phase. Using the term $\Delta P_{\text{mech}}$ to predict the location of hot tears in the cold cast works only in the start-up phase, as $\Delta P_{\text{mech}}$ is positive just above the ingot lip and then becomes significantly negative further into the start-up phase. This would predict that in the start-up phase, hot tears would occur only in the portion just above the ingot lip. With respect to the hot cast – cold cast comparison, the RDG criterion has been successful in describing the variation in hot tearing susceptibility between the hot cast and the cold cast in the start-up phase, but only because the of higher cooling water flow rate in the cold cast. As shown in Figure 6.18, the magnitude of $\Delta P_{\text{total}}$ in the start-up phase is much larger in the cold cast as compared to the hot cast.

The implementation of the RDG criterion into the ingot casting model also highlighted that the mesh resolution in the ingot DC casting model has an effect on the depression pressure predictions. Since the model did not provide enough data points to properly integrate the variation in fraction solid and permeability along the length of the mushy zone, it is difficult to say whether the observed trends in $\Delta P_{\text{total}}$ are accurate.
Figure 6.14 - Depression pressure variation 50 mm above the ingot lip along the rolling face (between Point A and Point B) for the Cold Cast Simulation. Note the difference in y-axis scale between plot (a) and plot (b).
Figure 6.15 - Variation in solid fraction and cooling rate (a) and strain rate (b) in the mushy zone related to Point A and Point B for the cold cast simulation.
Figure 6.16 - Depression Pressure Variation at the centreline of the rolling face at various heights from the ingot lip for the Cold Cast. Points A, 0.2 m, C and D are marked. Note the difference in y-axis scale between plot (a) and plot (b).
Figure 6.17 - Variation in fraction solid and cooling rate at Point A, Point 0.2 m and Point D (a), and variation in fraction solid and thermal pressure at Point 0.2 m and Point C (b) for the cold cast simulation.
Figure 6.18 - Depression Pressure Variation along the rolling face 50 mm above the ingot (a) and at the centreline of the rolling face at various heights from the ingot lip (b) for the Hot Cast and Cold Cast Simulations. Points A, B, C and D are marked.
6.6 Towards a Two-Stage Hot Tearing Criterion

Up to this point, the hot tearing defect has been looked at in two distinct ways: using strain accumulation during solidification in Sections 6.3 and 6.4 and also using a pressure drop within the liquid, in Section 6.5. Both methods have their drawbacks. The total strain method requires much knowledge of the ductility – temperature – strain rate relationship in the mushy zone. This information is very hard to measure. Furthermore, defect nucleation is completely ignored. To predict hot tearing using the pressure drop method requires the elusive critical pressure for pore formation. The underlying assumption is that once pores are created they will develop into hot tears. This completely ignores damage accumulation and defect growth. Moreover, the current implementation of the RDG criterion does not work for ingot castings, since $\Delta P_{\text{shrink}}$ is quite large at the end of solidification yet no porosity is seen. Also, $\Delta P_{\text{mech}}$, the term most likely responsible for hot tearing (as opposed to porosity), is much smaller than the term $\Delta P_{\text{shrink}}$. What this means is that the RDG criterion can be used to determine the regions which might be susceptible to hot tear / porosity nucleation, but some other method is needed to quantify the the growth of these initiation sites into fully developed hot tears.

The key issues in predicting hot tear formation is one of quantifying nucleation and then accumulation of damage in the material in the critical range of temperature (fraction solid) where there is a combination of sufficient mechanical strength to bear load and yet limited ductility. The combination of the RDG criterion and a total strain criterion would provide further insight into hot tearing. This would be somewhat similar to the ICS criterion, recently developed by Mo et al.\cite{56}, and described in Chapter 2 – Literature Review. For AA5182, the ICS criterion cannot be used since neither of the key parameters: critical pressure, $P_c$, and fraction solid for grain coalescence, $f_{\text{coal}}$, are known. However, one could use the RDG criterion to look at the tendency for hot tear initiation in different casting recipes, and then use the total strain criterion to determine which regions would form large hot tears from the nucleation sites.

6.7 Damage Accumulation during Tensile Loading in the Semisolid State

A preliminary investigation was conducted into the accumulation of damage in the critical range of temperature. In this work, a series of mushy zone tensile experiments were performed using the apparatus developed by Colley\cite{55} to measure stress – strain behaviour of aluminum alloys in the semisolid state. This apparatus consists of a modified Instron mechanical testing
machine connected to a Gleeble Thermomechanical Simulator. A low force (4.5kN), load cell was used to enable accurate measurement of the low loads associated with the partially solidified state. A digital camera and zoom lens were used to acquire a series of high-resolution images of the specimen diameter during testing, from which the instantaneous dimetral strain was calculated.

The experimental work has been carried out on aluminum alloy AA5182, supplied in the as-cast state by the Arvida Research and Development Centre (ARDC) of Alcan International. A diagram of the sample geometry is shown in Figure 6.19.

In the current work, two experiments were performed using this apparatus. In each experiment, the load was applied at a cross-head speed of approximately 0.085 mm s$^{-1}$ and the heating rate was 1.5 °C s$^{-1}$. For the first experiment, a set of tests was performed at varying temperatures (510 °C to 545 °C) to compare the fracture surface of these samples to an actual ingot hot tear fracture surface. In the second experiment, a set of tests was performed at 520 °C to measure the accumulation of damage at various plastic strains. In these tests, each sample was loaded in tension for a different length of time, and then stopped. The longest time was the time necessary to fracture the sample.

![Diagram of sample geometry](image)

Figure 6.19 - Schematic of sample geometry with both axial direction (a) and cross-section (b) shown.
6.7.1 Fracture Experiments

The fracture tests were conducted at four temperatures: 510 °C, 520 °C, 535 °C, and 545 °C. Fracture was investigated using high magnification SEM to determine the effect of temperature on the fracture surface. The SEM micrographs for each fracture surface are shown in Figure 6.20 and Figure 6.21. As can be seen in Fig. 6.20(a), the fracture surface at 510 °C is rough and contains clearly outlined grains. Thus, fracture appears to have taken place under conditions consistent with the material being a fully solidified structure. In Fig. 6.21(a) and Fig. 6.21(b), the fracture surfaces are smooth and rounded. This suggests that some liquid was present during the 535 °C and 545 °C tests, and that fracture occurred in the intergranular region as this is where the first liquid appears during any heating process. The most interesting fracture surface was formed at 520 °C, and shown in Fig. 6.20(b). In this micrograph, fine whiskers can be seen to have been created during the fracture process. An EDX analysis revealed that these whiskers were mostly aluminum, and not some complex precipitate forming due to melting of the liquid during fracture.

The yield stress and strain to failure as a function of temperature measured during this experiment is shown in Figure 6.22. These results show that while the alloy AA5182 has significant ductility at high temperatures when fully solid, the dramatic loss in ductility in the mushy zone at high fraction solid occurs at a temperature almost immediately above the solidus temperature. At 535 °C and 545 °C, the measured ductility was zero, as the sample broke right after yielding of the alloy. According to Thompson\textsuperscript{521}, the fraction solid of the samples varies from a fraction solid of one at 510 °C to a fraction solid greater than 0.98 at 545 °C. Clearly, the presence of liquid films has a great effect on the alloy's ductility, as even only the small amounts of liquid seen at 535 °C and 545 °C cause the material to become brittle. Furthermore, these two samples fractured without the accumulation of any damage, as optical micrographs of the cross section of these samples did not reveal the presence of any porosity. In contrast, the yield stress results demonstrated that the yield stress of AA5182 is much less susceptible to the presence of liquid films as it decreased only slightly with increasing temperature for the four trials.
6.7.2 Damage Accumulation Experiments

The accumulation of damage during tensile loading in the semisolid phase was assessed by performing four tensile tests at 520°C. This temperature was chosen because it was the highest temperature that one could measure ductility in the sample during testing. In these tests, each sample was loaded in tension for a different length of time, and then stopped. The four times were 20 s, 50 s, 75, and 85 s. In the first three tests, the experiment was stopped before the sample fractured. Unfortunately, the amount of liquid at this temperature is quite small. According to Thompson, the fraction solid of AA5182 at 520°C is greater than 0.99. This makes it very difficult to say for certain that liquid is actually present in the sample at this temperature. From Figure 6.20, the fracture surface at 520°C was shown to contain aluminum whiskers. This could be evidence of small pockets of liquid only beginning to form.

Each sample was sectioned and polished to measure the porosity of each sample in the region of maximum strain. The measurements are given in Table 6.1, along with the strain in necked region, and the yield stress. As can be seen in these results, there is fairly significant porosity growth in the sample before fracture. The sample loaded for 20 s had accumulated damage of ~2 %, and the sample that fractured (after 85 s) had accumulated ~16 % porosity in the vicinity of the crack. The most surprising results were at 50 s and 75 s, since the sample loaded for 50 s accumulated 20 % more strain than the sample loaded for 75 s, yet the porosity measured at 75 s was significantly larger as compared to 50 s. These results suggest that microstructure and second phase particles have a great effect on porosity and fraction solid upon reheating. Obviously the solidification conditions in the sample held for 50 s were not the same as the other three tests.

<table>
<thead>
<tr>
<th>Test Time (s)</th>
<th>Yield Stress (MPa)</th>
<th>Measured True Strain</th>
<th>Porosity</th>
</tr>
</thead>
<tbody>
<tr>
<td>20 s</td>
<td>19.8</td>
<td>0.09</td>
<td>2%</td>
</tr>
<tr>
<td>50 s</td>
<td>19.7</td>
<td>0.25</td>
<td>5%</td>
</tr>
<tr>
<td>75 s</td>
<td>18.7</td>
<td>0.21</td>
<td>14%</td>
</tr>
<tr>
<td>85 s</td>
<td>18.3</td>
<td>0.38</td>
<td>16%</td>
</tr>
</tbody>
</table>
6.7.3 Insight into Damage Accumulation during Solidification and Hot Tearing

These preliminary results reveal two important characteristics about damage accumulation and hot tearing. Firstly, in the presence of liquid films very little damage must accumulate before fracture, as no ductility was measured at the test temperatures of 535 °C and 545 °C. Secondly, in the region where ductility is measurable, the amount of damage accumulated before failure can be significant. However, as shown in the tests at 50 s and 75 s, the formation of porosity as a function of strain is quite dependent on the local microstructure. This makes it difficult to link porosity to strain and ultimately, hot tearing.

It also must be mentioned that there is a major assumption inherent in these experiment – the conditions obtained on reheating samples up to the mushy zone are identical as the conditions obtained on cooling from the liquid. It is very difficult to assess if this assumption is valid. There may be a number of complex precipitates or inclusions that become trapped during solidification. Although they have little effect on stress – strain during cooling, upon reheating, these inclusions may act as crack initiation sites in the presence of liquid.

Clearly, further study on damage accumulation and constitutive behaviour in the mushy zone is needed to better predict the nucleation and growth of hot tears.
Figure 6.20 - SEM micrograph of the sample fracture surfaces for tests performed at 510°C (a) and 520°C (b).
Figure 6.21 - SEM micrograph of the sample fracture surfaces for tests performed at 535°C (a) and 545°C (b).
Chapter 6: Hot Tearing Predictions in the ingot DC Casting Process

Figure 6.22 - The variation in tensile yield strength and strain to failure for AA5182 with temperature.
Chapter 7: Summary and Conclusions

7.1 Summary and Conclusions

Hot tears form in the last stage of the solidification sequence, in a region where the grains are surrounded by a continuous film of liquid and so they cannot sustain tensile strain. Application of strain causes the metal to fail along the path of the liquid film, creating a hot tear. Over the years, a number of researchers have investigated hot tearing, and proposed a number of criteria in which the susceptibility to hot tearing is evaluated based on alloy composition, thermal gradient and the development of stress and strain within the mushy zone. However, because of the difficulty in reproducing the mushy zone conditions, these criteria have generally been validated using sensitivity analysis and literature data only. With the earlier development of a coupled thermal – displacement finite element model of the aluminum ingot DC casting process at UBC, both hot tearing susceptibility and criteria validation can now be investigated in aluminum ingots using the temperature, stress, and strain fields predicted by this process model.

Two hot tearing criteria have been implemented into the ingot DC casting model: Pellini’s Total Strain criterion, and the RDG Hot Tearing criterion. To investigate hot tearing, the FE model was run under two casting scenarios: a cold cast, which is prone to hot tearing in the start-up phase, and a hot cast, which is not prone to hot tearing. A comparison of the predictions made using these criteria to hot tearing observed industrially lead to the following conclusions:

- The predicted stress and strain fields can be used to explain hot tearing susceptibility in the start-up phase of the process. In the region just above the ingot lip, there is high accumulation of tensile plastic strain during solidification, which is caused by the superposition of the macroscopic butt curl effect and the local temperature gradient conditions. This tensile plastic strain is largest at the centreline of the rolling face. In comparison, in the steady-state phase where there is no butt curl effect, the plastic strain accumulated during solidification is small despite larger temperature gradients due to the increased cooling water flow rate.

- There is good agreement between the hot tearing predictions made by the Total Strain criterion, and industry observations. Using the strain predictions from the cold cast simulation, this criterion predicts a region prone to hot tearing in the middle third of the rolling face, in the start-up phase. In this critical region, the plastic strain accumulated during solidification exceeds the ductility limit at a temperature of about 575 °C. Both
further up the ingot, and out towards the edge of the rolling face, the plastic strain accumulated during solidification was less than the ductility limit. The conditions in the hot cast were such that the predicted plastic strain during solidification on the rolling face was also always less than the ductility limit.

• There is poor agreement between the hot tearing predictions made by the RDG criterion and industry observations. In both the hot and cold casts, the RDG criterion predicted the highest hot tearing susceptibility to be in the steady-state phase of the process. Furthermore, the shrinkage depression pressure ($\Delta P_{\text{shrink}}$) was much larger than the mechanical depression pressure ($\Delta P_{\text{mech}}$) and so it dominates the hot tearing susceptibility interpretation. Generally, $\Delta P_{\text{shrink}}$ is used to predict porosity during solidification. $\Delta P_{\text{shrink}}$ is largest in the steady-state phase because the cooling water flow rate, and hence the cooling rate, is also largest in the steady-state phase. In terms of $\Delta P_{\text{mech}}$, most of the mushy zone is under a compressive strain rate with only the low temperature or high fraction solid regime in tension. This makes the mechanical depression pressure term small or negative, which implies that hot tearing is not affected or is even hindered by this term.

• The RDG criterion is capable of differentiating the hot tearing susceptibility between the hot and cold casts in the start-up phase, because the cooling water flow rate is quite different between these two casting recipes. In the cold cast start-up, with high cooling water flow rate, $\Delta P_{\text{total}}$ was quite high. In contrast, in the hot cast start-up, with low cooling water flow rate, $\Delta P_{\text{total}}$ was almost zero.

• The size of the mesh used in the ingot DC casting process model is too coarse for a proper calculation of the depression pressure, since it does not capture sufficient resolution in the spatial gradient in fraction solid along the mushy zone.
7.2 Recommendations for Future Work

The conclusions drawn from this research have been based primarily on the assumption that hot tearing susceptibility is controlled by a nucleation stage (RDG criterion) or a damage accumulation stage (Total Strain criterion). Although the Total Strain criterion seems to work best for ingot castings, what is actually needed is a two-phase hot tearing criterion, including both defect nucleation and damage accumulation. This would allow the investigation of situations more complex than the hot and cold casts, as both of these casting recipes were extreme and not standard industry practice.

The critical hot tearing strain used in this study was based on a limited number of experiments in the mushy zone. More tests should be done, to improve resolution in the data and the repeatability. Furthermore, the critical hot tearing strain is measured at constant strain rate and temperature and a method is required to link these values with the actual process of strain accumulated over a range of strain rate and temperature.

Since the mesh resolution is too coarse in the ingot DC casting model to fully resolve the spatial variation in fraction solid along the mushy zone, research should be initiated to either refine the mesh of the ingot casting model, or to develop a code within the RDG hot tearing subroutine to interpolate the temperature between element intersections.

Finally, to more fully validate the Total Strain criterion, it should be implemented in other aluminum alloy systems that are both more susceptible and less susceptible to hot tearing than AA5182. This would be worth pursuing as it would show whether the Total Strain criterion can easily differentiate between alloys which are prone to hot tearing and alloys which are not, or if a more complicated, two-phase criterion is needed.
References


47. Gadala, M., MECH 515 Class Handouts, 2003.


