Experimental Observations and Analysis of Macrosegregation in Rolling Slab Ingots

by

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ABSTRACT

The increased demand for aluminum as a primary structural metal stems from a quest in automotive, aerospace, and marine industries to be more energy efficient and sustainable. This unprecedented demand drives aluminum casting methods towards increased productivity looking to, cast larger ingots faster. The unfortunate consequence of this approach is an enhanced variation of metallurgical properties over the cross section of slab ingots. Rolling slab ingots of AlCu4.5 using a typical Direct-Chill casting technique have been cast and sectioned for analysis. This alloy allowed us to compare our results with the available literature and to elucidate the marked differences in spatial variation of microstructure and composition found in radial and lateral symmetry castings. In an attempt to couple conventional theory with our results, sump and temperature profiles were measured in-situ and modeled using a commercial finite element analysis software package. The combination of experimental and modeling results indicate that the variations in the cooling parameters through the cross section are largely responsible for the spatial variances in metallurgical properties, pointing to a possible refinement of DC casting parameters.
MOTIVATION

Direct-chill casting of aluminum was developed between 1936 and 1938, nearly simultaneously in Germany (W. Roth, VAW) and in the USA (W.T. Ennor, ALCOA). The technology was roughly based on the methods for casting aluminum and copper suggested by Zunkel (1935) and Junghans (1933). The process has been classified as one of the premier developments in materials processing and has allowed aluminum to become a primary structural material. Since the days of Roth and Ennor, the process has improved dramatically, and the Direct-Chill process is the primary mode of manufacturing for rolled aluminum products.

Even with the strides in technological advancement, defects such as bleedouts, hot tears, and macrosegregation continue to plague practitioners, adversely affecting productivity. A great deal of research has been conducted in recent decades understanding the various mechanisms driving macrosegregation in DC casting. Unfortunately for researchers and practitioners alike, the forces driving macrosegregation interact in complex ways inside of castings. Different alloys, even within the same family, can have different dominant mechanisms making characterization of macrosegregation extremely difficult.

Today, computational researchers are able to approximate macrosegregation patterns in 2-dimensional models. Extending the problem into 3-dimensions adds an additional level of complexity because researchers must account for non-uniformity in metal distribution and cooling rate around the periphery of a casting. Industrially-sized experimental data for comparison is extraordinarily rare due to the large size and cost associated with processing each casting (>10 tons each). As a result, researchers have very little data to use as a benchmark for their models. In order our understanding of macrosegregation to continue to progress, there is a need for additional research in macrosegregation in DC cast rolling slab ingots.

LITERATURE REVIEW AND BACKGROUND

THE PHYSICS OF ALLOY SOLIDIFICATION

Many phenomena play important roles during alloy solidification depending on the length and/or time scale. In this section, only a few basic concepts, which are particularly relevant to the topic under consideration, are discussed.

In contrast to a pure substance which undergoes a phase change isothermally, alloys solidify over a temperature range, in which the solid and liquid co-exist in thermodynamic equilibrium. Figure 1 shows a typical equilibrium phase diagram for a fictitious eutectic-forming binary system A-B at constant pressure. At temperatures above the liquidus lines, a single liquid phase exists as a solution of constituents A and B. Liquid and solid phases coexist in equilibrium over a range of temperatures between the liquidus and solidus lines, up to the eutectic point. During the solidification of a binary alloy, the two phase region separates fully solidified and melt regions, and is known as the mushy zone.
At the eutectic point, a three phase mixture of a liquid phase and two solid phases designated α and β, exists.

When an initially superheated alloy cools, nucleation of small crystals occurs at a temperature slightly below the liquidus temperature associated with its composition. As the temperature of the solid-liquid interface decreases, the compositions of the solid and liquid interface continually change. As indicated in Figure 1, on the left side of the eutectic (hypoeutectic alloys), the solubility of species B is lower in the solid than in the liquid, and the rejection of species B from the solid leads to an increase of the B concentration in the liquid at the interface. In most practical casting processes, this condition of phase equilibrium at the solid-liquid interfaces is met. However, deviations from the phase diagram can take place due to capillarity, pressure, and kinetic effects (at high solidification rates)\textsuperscript{1,2}

The formation of a solid is basically governed by the temperature and species concentration at the interface. However, the development of interface shape is a more complicated issue, involving stability and interface curvature considerations. It is of utmost importance in solidification modeling to take into account these microscopic interfacial features, because they ultimately determine the microstructure, which strongly affects the mechanical properties of the material. There are essentially two basic growth morphologies that can exist during alloy solidification. These are dendritic and eutectic morphologies\textsuperscript{1}. Dendrites grow with a very large specific surface area and irregular solid-liquid interface. In the macroscopic sense, such growth is virtually irresolvable. Hence, the mushy zone, which is comprised of solid dendrites and interdendritic liquid, is treated as a porous solid structure which is saturated with interdendritic liquid. The growth of eutectics is much simpler than that of dendrites because the eutectic grain maintains a simple geometric interface shape. Generally, both morphologies develop together.

Due to the low mass diffusivity relative to the thermal diffusivity for metal alloys,
Figure 1: Sample phase diagram for fictitious binary eutectic system

The formation of the aforementioned microscopic structures is mainly dependent on the species concentration gradient on each side of the solid-liquid interface. On the liquid side of the interface, most of the solute is rejected from the solid, forming higher interfacial concentration than in the liquid away from the interface. The difference between the concentrations at the interface and nearby bulk liquid is usually referred to as the solutal undercooling (Figure 2). In other words, the actual temperature in the liquid is below the liquidus temperature corresponding to the interfacial liquid concentration. The actual temperatures of the interface and bulk liquid can also be different from each other, and the difference is referred to as the thermal undercooling. The thermal undercooling is usually relatively small in metal alloys, because of the high thermal diffusivity in metal alloys, and thus, since the solute gradients are confined to the small region near the interface, convective influences on the solute transport have been neglected traditionally. However, vigorous convection can severely alter the microscopic concentration and temperature profiles, and hence the movement and shape of the interface can be different from the traditional approach.¹
Figure 2: Solutal undercooling in binary alloys. (a) Phase Diagram. (b) Temperature profile at interface.
MACROSEgregation

During solidification of aluminum alloys, solute is partitioned between liquid and solid phases to either enrich or deplete inter-dendritic regions. This continual process leads to compositional variations within grains on the scale of micrometers, otherwise referred to as microsegregation. In contrast, macrosegregation refers to chemical variances over larger length scales up to the dimensions of the casting, which can be on the order of meters in industrial settings. Due to the short distances associated with microsegregation, its effects can be easily counteracted by homogenizing heat treatments post casting. Conversely, the large scale variances associated with macrosegregation mean that it cannot be mitigated by any heat treatment of a reasonable duration. Macrosegregation, in all of its forms, is generally undesirable for casting manufacturers since the compositional variance can drastically alter mechanical and microstructural properties. As a result, considerable time, effort, and money have been spent in understanding and mitigating macrosegregation in cast products.

Perhaps some of the first observed effects of microsegregation come from wootz steel blades, forged for nearly a millennium in the Middle East and Asia. The distinct watering pattern on such blades was found to be due to the segregation of primary carbide formers such as chromium and vanadium to interdendritic spaces and grain boundaries. Such carbides remained during subsequent thermo-mechanical processing, and revealed their beautiful patterns when etched. The first documented observations of macrosegregation come from medieval Europe. V. Biringuccio, an Italian foundryman, described large scale segregation in his bronze gun barrels in 1540. Thirty years later, L. Ercker an Austro-Hungarian chemist, published observations of liqutation in his precious metal castings. Additional observations of liqutation, now known largely as inverse segregation, were made into the 19th century by researchers again working with precious metals for currency applications, where purity and uniformity are crucial. With the invention of the Bessemer process, bulk liquid steel could be produced reliably and macrosegregation studies began on steel castings. As early as 1866, Russian metallurgists noted that the degree of macrosegregation depended on the size of the casting, and that low melting point constituents tended to agglomerate in the center of castings. As Hall and Héroult developed their processes for electrochemically extracting aluminum, aluminum went from being a precious metal, to a key metal in the industrial age. Correspondingly, large scale castings of aluminum were made and macrosegregation was once again observed. The pioneering work of G. Masing, W. Claus, S.M. Voronov, and W. Roth initiated research into macrosegregation in aluminum billets and ingots.

Up until the mid 1960’s the predominant cause of macro segregation was believed to be solute buildup at the tips of dendrites. A simple solute mass balance was developed to describe solute rejection in alloys now known as the Gulliver-Scheil segregation law. Gulliver derived these equations by considering the mass balance of infinitesimal crystal layers occurring in a decreasing temperature gradient. His result was in the form of an infinite product, and this awkward construction caused it to be overlooked until Scheil and Schemer re-derived the equations in a more familiar differential form. In the mid 1960’s, Flemings and coworkers coupled solid-state diffusion and a solutal fourier number which allowed them to estimate the amount of rejected solute for a given solidification condition. Flemings and coworkers went on to revolutionize our understanding of macro segregation by discovering the importance of convection within the mushy zone, and derived the basic equations describing
interdendritic flow\textsuperscript{21-32}. The magnitude of this research can be seen in the numerous improvements in casting research, as well as the cross-disciplinary understanding in areas as diverse as solidifying magmas and sea ice. Since the publishing of this research, our understanding of macrosegregation has expanded and the current understanding is that the fundamental cause of macro segregation is the relative movement or flow of segregated liquid and solid during the solidification process. We can further break down this advection into the following causes present during casting:

1. Shrinkage induced flow that feeds the volume change of liquid and solid under cooling.

2. Flows induced by buoyancy caused by thermal and solutal liquid gradients. Though sometimes grouped together as Thermal-Solutal convection, these effects may be additive or deleterious depending on thermal gradients and how rejected solute influences the liquid density.

3. Forced convection due to inlet conditions, gas bubble motion, magnetic fields, stirring, or vibration.

4. Redistribution and preferential sedimentation or suspension of free crystals or dendrite fragments that previously nucleated heterogeneously on a wall or in the melt.

5. Mechanical deformation due to external forces or induced by thermal boundary conditions.

As intricate as each of these causes may be, real casting conditions will include several if not all of them at some point in the casting process. Dominant effects can change drastically depending on process, and even amongst alloy families using the same process. For this reason, macrosegregation continues to be an active area of research especially amongst numerical researchers of casting processes\textsuperscript{33-35}. The complex interplay between driving forces makes each cause even more difficult to analyze experimentally, and papers regarding such experiments are generally focused on describing processes, experimental conditions, and results. However, because any result could be caused by one or more of the aforementioned causes many results are indefinite and difficult to repeat. In general, macrosegregation can be divided into two general patterns: normal and inverse segregation. In alloys with a partition coefficient (K defined as slope of liquidus/slope of solidus) less than one, normal segregation refers to the case where the hot zone is enriched in solute carried from the solidification interface by convection or shrinkage induced flow. Inverse segregation, as is usually the case for DC casting forms an opposite structure where the hot center is depleted of solute.

MACROSEGREGATION IN DC CASTING OF ALUMINUM ALLOYS

DESCRIPTION OF THE DC PROCESS

The Direct Chill (DC) casting process has been used commercially since the 1930’s for the production of non-ferrous billets and ingots for further processing\textsuperscript{36-39}. DC casting is a semi-continuous process used extensively in the aluminum industry to produce ingots and blooms from a wide range of aluminium alloys for subsequent rolling into sheet products as well as cylindrical billets for extrusions and forgings\textsuperscript{39-41}. Typical for the cast products are 200 mm in diameter for extrusion billets and 1500x500 mm for rolling ingots.
A schematic diagram of the DC casting process during steady-state casting is shown in Figure 3: DC Casting apparatus and cooling regions during casting. At the start of the process, a starter block is partially inserted into a water-cooled copper or aluminum mold. The starter block is initially positioned slightly above the lower lip of the mold to close off its bottom. The mold is then filled with molten metal until the desired mold metal level is reached, then the bottom block is gradually lowered into a casting pit carrying with it the solidifying ingot. Cooling water circulates in the water manifold of the mold and heat is transferred from the liquid metal to the mold. The removal of heat through the mold walls is called primary cooling.\textsuperscript{39,42,43}

The starter block continues to be lowered at the desired casting speed and more melt is poured into the mold to maintain a constant metal level in the mould. Once the semi-solid shell of the ingot leaves the bottom of the mold, the emerging ingot surface is impinged directly by cooling water jets that exit the mold bottom through a series of holes or slots to further cool the casting. After the direct spray of water impinges on the surface, it runs down the outer perimeter of the ingot and into the pit. The direct contact between the cooling water and the ingot surface is known as secondary cooling, and is responsible for the largest amount of heat extraction during steady-state casting. The casting process continues until the desired length of the ingot has been reached.

\begin{figure}[h]
\centering
\includegraphics[width=\textwidth]{diagram.png}
\caption{DC Casting apparatus and cooling regions during casting}
\end{figure}
While we have already elucidated the root cause of macrosegregation to be the relative movement of solid and liquid phases, why this relative motion is significant and disrupts the homogeneity of the casting is the difference in composition between the two phases. Therefore it is not much of a surprise that the magnitude and pattern of segregation in a casting is closely tied to the partition coefficient $K$, which is defined as the ratio of the slope of the liquidus over the slope of the solidus. The majority of alloy elements present in commercial aluminum alloys are hypoeutectic in composition and have a partition coefficient $K<1$. A typical macro segregation pattern for DC cast aluminum alloys has been schematically represented Figure 4. Characteristics of this structure include a central region depleted of solute adjacent to a solute-rich region at the mid thickness. The subsurface is noticeably depleted of solute, while the surface is rich in solute. This is generally referred to as the “W” segregation pattern.

![Figure 4: Schematic representation of a typical "W" macrosegregation pattern found in DC casting of Aluminum alloys](image)

**Characteristic Width**

Not all elements react with aluminum in a eutectic reaction. Elements such as Titanium and Chromium exhibit a peritectic reaction with aluminum and thus have a partition coefficient $K>1$. Structurally, this causes these alloys to have macro segregation patterns inverted from Figure 4. Since macrosegregation cannot be removed from downstream processing, the shape and magnitude of these undulations is very important to the acceptability of the cast product. Alloy compositions are given with a range of acceptable values, and in some cases the amplitude of the solute deviations may fall within that range. In such conditions, macrosegregation is of minimal concern. In other cases, downstream processing can distort the casting enough that the spatial variation in concentration can be ignored (foil or thin gage
sheet products). In extreme cases, and those of the most concern to industrial casters, solute variations can fall outside the acceptable range of the alloy specification, and cannot be ignored due to mechanical processing (Plate). It is the final case that has motivated the bulk of the research into macrosegregation, and is the source of additional research by R&D centers to mitigate its effects.

1. SHRINKAGE INDUCED FLOW

In aluminum alloys, the volumetric shrinkage upon solidification is on the order of 6-8 vol%. As a consequence of this shrinkage, liquid metal flows into the mushy zone driven primarily by metallostatic pressure. While this shrinkage flow velocity may be small in magnitude (mm/s), it is significant because it occurs immediately before solidification, consequently its effects are immediately locked in place and are irreversible. The effects of shrinkage- induced flow occur due to the relative composition difference between the solid dendrites and the surrounding liquid. If one considers a representative volume element of the mushy zone (see Figure 5) where there is an increasing temperature gradient in the positive X direction, solidification then progresses in the negative X direction. As solidification progresses, more and more solute is rejected to the liquid causing a composition gradient increasing in the negative X direction. As shrinkage induced flow moves fluid in the negative X direction, traversing the volume element, it moves the most enriched solute deeper into

![Figure 5: Representative Volume Element from Mushy Zone](image)

negative X direction.
the solidifying interface. By conservation of mass, the fluid removed from the element is then replaced at the right boundary by less enriched liquid. As this volume element continues to follow the solidification interface, enriched solute is continually drawn deeper into the sump, enriching those regions. The enrichment of the deeper portions of the mushy zone come at the expense of the crystals still early in the solidification process as their surrounding solute is replaced by non-enriched liquid. In the case of DC casting, this condition occurs most notably at the surface. The rapid solidification of material at the surface rapidly draws in solute enriched fluid adjacent to the surface thereby enriching it. Consequently, the area adjacent to the surface becomes solute poor as can be seen in Figure 4.

It has been previously asserted that shrinkage induced flow occurs in a direction perpendicular to solidification. If we decompose the direction of flow into a vertical and horizontal component, the horizontal component directs flow to the casting surface, while the vertical component directs flow in the casting direction. Considering first the vertical component we can immediately see that this will not generate the negative centerline segregation present in Figure 4. Considering only vertical flow, this is nearly identical to the aforementioned steady state unidirectional solidification condition analyzed by Flemings et al.. As solidification continues in the vertical direction, a steady state criterion quickly builds up where the volume element continually accepts and rejects solute of the same respective concentrations. Consequently, there is very little macrosegregation that results due to vertical flow. The typical macrosegregation pattern is instead generated by a flow from the center to the surface, a function of the horizontal component. The resultant structure of this slow moving flow, is a net movement of solute from the center to the surface. Since no liquid can exit the control volume at the surface, a buildup of solute results. The magnitude of the resulting pileup is thus a function of the shrinkage flow, which is a function of geometric parameters, mushy zone thickness, and mushy zone permeability.

Eskin et. Al. recently proposed a scaling model to predict the volume of fluid drawn into various sections of the mushy zone during casting, and thereby provide an estimate for the macrosegregation. Figure 6 is a schematic representation of a typical solidification interface (sump) found during DC casting. Based on the aforementioned assumptions, Eskin proposes a simple scaling model that relates the slope of the mushy zone α, its thickness Lm, shrinkage ratio β, and the unique path taken during an alloy’s solidification (coefficient A). The resulting equation is displayed below, where LH is defined to be the horizontal solute transfer distance.

\[ L_H = A C_0 L_m \beta (\sin 2\alpha) / 2 \]  \hspace{1cm} (1)

The total flux of solute for a given control volume can be estimated by taking the derivative of the above equation with respect to distance from the center. Relative macrosegregation can then be determined by dividing this result by the average composition. While this scaling analysis is rough, it provides a framework by which a researcher can begin to understand the driving mechanisms at play. More complex numerical models can incorporate shrinkage induced flow with variable permeability as a function of microstructure.
Figure 6: A Schematic representation of a typical sump during DC casting of an Aluminum Billet. Adapted from Eskin (39)

2-3. CONVECTION (Forced and Natural)
Natural convection in a solidifying ingot can be driven by either temperature or concentration gradients in the molten pool, or mushy zone. Due to the process parameters of DC casting, a temperature gradient is naturally set up between the ingot surface and its center. This results in a natural density gradient as well. Near the solidifying interface, the cooler liquid sinks, following the sump to the center of the ingot. As the two “wall jets” converge in the center, the resulting interaction causes the liquid to rise from the bottom to the top of the liquid pool forming a continuous flow of liquid rising from the center. The resulting motion is a rather complex balance. As the plume rises from the center of the ingot, it curves back towards the solidification interface at the surface and flows into the mushy zone parallel to the temperature gradient. As the flow passes through the mushy zone, it is eventually deflected by the coherent dendrite network. This deflected flow combines with the natural downward convection to generate flow through the mushy zone, parallel to the solidification interface. The two flows converge again at the center generating a steady state flow condition. Solute rejected from the solidifying dendrites can affect the local density of the fluid and can either magnify or mitigate many of the effects of thermal induced convection. In general however, the described motion is present and leads to positive centerline segregation as the wall jet flows along the solidifying interface in the mushy zone, drawing rejected solute to the center. As the flows converge, the less enriched fluid (typically cooler) remains at the bottom of the sump while the less enriched liquid contributes to the buoyant plume and flows upward. The result is a continued enrichment of the centerline and stabilization of the thermally induced flow regime. As stated before, the density of solute can affect the magnitude of these convective currents, consequently, one could assume this also affects the magnitude of the induced positive centerline segregation. In fact it has been demonstrated that an

Figure 7: Analytic Solution to Convection Problem in DC Casting (from 39)
addition of magnesium to an Aluminum-Copper alloy does reduce the magnitude of the centerline segregation, when only considering convective flows\textsuperscript{51}.

A similar type of analysis can be conducted for forced convection driven by external fields or flow entrance conditions. Chu and Jacoby of Alcoa demonstrated that ingots cast using the bi-level pour method (where metal enters through a downspout and distribution bag) were characterized by 15% more centerline segregation than the hot-top method. The difference in macrosegregation was attributed to the difference in flow pattern introduced during the two methods\textsuperscript{53}. Further investigation indicated that an optimization of the metal entrance condition could achieve a significant reduction in centerline segregation, by directing the metal downward instead of towards the periphery\textsuperscript{54}. Instead of modifying convection conditions, this result was attributed the intense dendrite fragmentation occurring at the periphery using the standard method. These fragments then nucleated unique crystals and were washed to the center of the ingot. By imposing a downward flow condition, the excessive fragmentation is reduced and the volume of sedimenting grains is reduced (see below).

By introducing an external field, such as those imposed during ElectroMagnetic Casting (EMC), further modifications to macrosegregation patterns can be achieved. Inherent to the process are shallower sump depths, and more convection leading to a more homogenous liquid. The shallower sump depth alone should decrease the magnitude of shrinkage induced flow alone (see above). By fine tuning the frequency of stirring, Zhang et al were able to produce a 7075 billet with nearly no macrosegregation\textsuperscript{55}. The small diameter of this billet has brought up some skepticism regarding the applicability of this technique regarding large scale castings, since such a result has yet to be produced on a rolling slab ingot. However, the fact remains, that significant reduction in macrosegregation can be achieved through precise forced convection.

4. MOVEMENT OF SOLID GRAINS (Equiaxed)

The motion and preferential sedimentation of equiaxed crystals is one of the major contributing factors to negative centerline segregation. From the early days of DC casting (1940’s), the idea of preferential sedimentation was proposed as a major mechanism of macrosegregation. However, this proposal was rejected early in the literature because large floating grains (“daisies”) could not always be observed in ingots with negative centerline segregation. However, numerous experiments illustrated the correlation between so-called “duplex structure”, where a mixture of coarse-cells and dendrites are found in the center of a casting, and inverse segregation\textsuperscript{53,56-62}. The idea was loosely re-instated in the 1980’s, and after a relatively contentious debate amongst the modeling community, Vreeman et al were able to use a volume of fluid model to predict inverse segregation due to the presence of sedimenting crystals\textsuperscript{63}. As was discussed previously, the likely mechanisms for solid transport are the convective currents inside of a solidifying ingot. As the equiaxed crystals are transported from the slurry region of the mushy zone by sediment transport, they are drawn to the center of the ingot. However, due to their increased density in comparison to the liquid, they are generally not conveyed upward with the convective plume. Instead, a steady state system generates where grains are preferentially deposited in the center, and then frozen into the coherent network. Other mechanisms have also been proposed to help explain inverse segregation, such as the so-called “avalanche” model. In this model, since the slurry zone forms at steep
angles in the center of the ingot, the topmost crystals instead form a continuous avalanche towards the center of the ingot simply because the angle of repose is too steep for these topmost crystals to pack effectively\textsuperscript{64}. Whatever the mechanism, the sedimenting grains fall into a region of intense undercooling where two planar solidification interfaces are converging. Thus, their solidification times are often much longer than typical grains and this effect can be seen often by their dendrite arm spacing being significantly larger than surrounding crystals\textsuperscript{56}. In addition to the grain transport mechanism explaining the origin of these crystals, other proposals have also been set forth. Instead of simple grain transport, it is proposed that as convective currents interact with the coherent network or columnar grains, dendrites or dendrite arms can be fractured off. These dendrite arms then travel with the fluid and serve as nucleation sites for grains further along in the solidification process. It is this mechanism that could explain the lack of observable sedimenting grains (“daisies”). Since the crystals are formed around a pre-nucleated embryo they will adapt a growth rate depending on their current environmental undercooling instead of that of the original embryo. Thus, the direct observation of sedimenting crystals would not be a unique criteria for free moving crystals. One must also consider the possibility of remelting as well and local coarsening behavior. If a crystal from a cooler edge region is transported to a warmer center region, there is the possibility that the crystal could begin to remelt, thereby destroying its thermal history. In addition to the fragmentation and direct transport methodologies, it has also been proposed that grains could nucleate on the liquid surface and settle to the bottom, or homogeneously nucleate in the intensely undercooled region where the interfaces interact, thereby increasing the volume fraction of primary-phase crystals\textsuperscript{55}.

While generally accepted as a mechanism for negative centerline segregation, some doubts have been expressed as to how much sedimenting grains may contribute to the overall macrosegregation pattern\textsuperscript{66}. This was partially due to an inability to experimentally track down the source of solute-lean material in experimentally cast ingots. Mainstream ideals stated that coarse-celled grains were solute poor, while finer grains were solute rich\textsuperscript{53,56}. This conclusion was most likely reached by direct observation, where coarser celled grains are present in the center of a casting, where inverse segregation occurs; whereas, finer material is located at the periphery and is typically more enriched in solute by comparison. Eskin et al used a microprobe analysis to verify this fact\textsuperscript{56}. However, it does raise the question of the sequence of events, as to whether the solute lean region caused the coarse cells, or whether the coarse cells, lacking a solute envelope to transport material enriched in solute (as seen in fine dendrites), result in inverse segregation.

This area of macrosegregation continues to be an area of study. While it is generally accepted that the sedimenting of crystals results in negative centerline segregation, models that can verify this experimental reality rely on the presupposition that floating crystals exist. Insitu observation is nearly impossible in a real as-cast situation, which means that a suitable alternative system would have to be used to conclusively identify the mechanisms and forces driving sedimenting grains.

5. DEFORMATION OF THE SOLID

Although the deformation of the coherent network is typically associated with macrosegregation in the continuous casting of steels, results indicate that even small deformations (2\%) resulting from thermal
stresses can have a large influence on macrosegregation\textsuperscript{57}. Regarding DC casting, this result most prevalently occurs in the surface regime\textsuperscript{68}. This effect is prevalent in the air-gap region of the mold in between primary cooling and secondary cooling. As the metal re-heats the regions between crystals begin to re-melt since they are the most enriched in solute. Here, a combination of effects forms the surface and subsurface regimes seen in Figure 4. As the ingot pulls away from the mold wall, there is a stress induced on the solid network promoting exudation of the inter-dendritic liquid. As this liquid is extruded to the surface, shrinkage driven flow replaces any additional voids with fresh material, thereby depleting the subsurface region of solute.

It becomes clear after reviewing the mechanisms driving macrosegregation, that additional research is needed to complete our understanding of how these mechanisms interplay. Unfortunately, industrial support by means of experimental data is usually alloy dependent. We have seen, especially in the case of convection driven flow, that alloy composition can have a significant effect on the mechanisms of macrosegregation. Adding in variations in casting parameters, and processes it becomes very difficult for models to reliably predict structures across the board. Furthermore, since all of the different mechanisms may have a role in some way or other in the final structure (sometimes even competitive) it is extremely difficult to isolate a single mechanism for study. Moving forward, it seems that model experts and experimental practitioners must team up in order to understand how each mechanism can affect each alloy for a given process.

**OBJECTIVES OF THIS RESEARCH**

The objective of this research is to investigate the complex interplay of the various macrosegregation mechanisms in the DC casting of a rolling slab ingot. While individual mechanisms have been painstakingly analyzed by other researchers, and numerical models have been constructed to reflect their knowledge, very little experimental data has surfaced regarding rolling slab ingots where boundary conditions are more complex than axisymmetric models. Unlike round billets, which can easily be modeled and explained in two dimensions, rolling slab ingots possess innate asymmetries what make them impossible to model in two dimensions. Adding their large size (10 tons or more), and the resources needed to investigate such castings, publications on these castings is prohibitively expensive and unwieldy. It is the goal of this research to investigate the macrosegregation patterns present in rolling slab ingots, and to provide mechanistic interpretations for the various structures observed, setting the stage for the development and analysis of a model that predicts macrosegregation in such slabs.

**EXPERIMENTAL APPARATUS AND PROCEDURES**

**INTRODUCTION**

The experiments completed during this investigation were focused on obtaining a quantitative data set regarding macrosegregation of an Al4.5Cu alloy cast in a typical rolling slab configuration. In addition to the quantitative data, additional experiments were performed to gain a more fundamental understanding of the mechanisms of macrosegregation at play in the DC casting of a rolling slab ingot.
For our alloy, we have chosen Al4.5%Cu, due to the abundant available literature. It also allows us the simplicity of a binary system, which we hope will make downstream analysis more straight-forward.

In the following sections, the experimental apparatus, diagnostic techniques, and experimental procedures, are discussed.

EXPERIMENTAL APPARATUS

Our experimental apparatus consisted of a commercial scale DC casting apparatus. An alloy of Al4.5Cu was melted in a natural gas fired 20 ton furnace. At the mouth of the furnace, the metal was inoculated with 20ppm TiB grain refiner. The metal then runs through a series of troughs made of 70% alumina and 30% silica to a run-through degassing machine. Since hydrogen is highly soluble in liquid aluminum, and much less so in the solid, solidification of hydrogen laden aluminum can lead to hydrogen porosity as the dissolved hydrogen comes out of solution upon solidification of the aluminum. The degassing machine pumps argon and chlorine through a series of spinning rotors immersed in the melt. The hydrogen migrates to the bubbles of argon and chlorine and is removed when the bubbles break the surface. From the degassing machine, the metal passes through a ceramic foam filter, which is designed to remove inclusions and other impurities present in the metal. After the filter, the metal passes through a ceramic spout outfitted with a control pin, which is capable of controlling the volume of metal flowing out of the spout, thereby controlling casting speed. A standard fiberglass combo bag was used to distribute metal evenly and quiescently into an open-top WAGSTAFF LHC™ mold with bore dimensions of 1750mmx600mm. The casting temperature was maintained at 35°C of superheat, while the casting speed was steadily ramped up to the steady state speed of 60 mm/min, and after startup the metal level was maintained at 147mm above the tang of the mold.
TEMPERATURE MEASUREMENTS

Three rakes of three thermocouples were fabricated using 28 gauge K-type thermocouple wire and stainless steel tube. The thermocouple junctions were welded together using a TIG welder under argon gas, forming a bead approximately two millimeters in diameter. Type K thermocouple wires were used due to their temperature range, cost, and availability. The thermocouple wires used were insulated using a resin-soaked fiberglass sheath. In order to prevent the formation of potentially conductive gases the thermocouple wires (20’ in length each) were placed in a 600°C oven for approximately 10 seconds. This was enough time to combust the resin, without severely degrading the fiberglass insulation. The opposite ends of the thermocouple junctions were attached to an Omega DAQ system and set to record temperature measurements every second, corresponding to each millimeter of cast length.

Once the casting had reached its steady state sump shape (~1200 mm of cast length) the thermocouple rakes were inserted into the melt one at a time. Figure 9 represents the shape of the thermocouple rakes used. When placing the thermocouples into the melt, care was taken to ensure the rake was flat. The long extension in the Z direction served as a mechanical support for the rake structure as the stainless tube was frozen into the sump. This support ensured the rake remained parallel to the liquid surface during the entire cast. Figure 10 is a representation of the location of the thermocouple rakes as they were placed into the ingot.
MACROSEGREGATION MEASUREMENTS

Once the cast was complete, the solidified ingot was removed from the casting pit and marked for sectioning. Figure 11 is a representation of the slices taken for analysis. A cross-section was taken at 1800 mm of casting length as measured from the start of the cast. This elevation ensured that we were capturing a steady state macrosegregation profile, as well as the solidification temperature profile captured by the thermocouples. An additional slice was made in the Z direction of the ingot, at the
centerline of the rolling face. This slice was used to capture the variation in centerline macrosegregation through the duration of the cast.

Once the cross sections were taken, the plates were marked for sampling. The XY cross section was divided into quarters, and 45 one-inch samples were removed from the first quadrant using a core drill. Including the dimension of the drill bit, this was the maximum number of samples that could be removed from a given quadrant. Figure 12 is a schematic representation of the locations of these removed samples. For the Z direction samples, a sample was removed each 6 inches from the center of the plate in a straight line from the bottom to the top of the ingot.

![Diagram of sample slice locations](image)

Figure 11: Schematic of Sample Slice Locations
The samples were marked using an electric pencil, identifying their location as well as the cast number. Each sample was then faced off using a machinist’s lathe to ensure flat surfaces at either end of the sample. The samples were then placed into an optical-emission-spectroscopy machine, which had been previously calibrated for this alloy family. Each sample was sparked 5 times and analyzed for copper composition. The number of sparks was maximized for the surface area of the sample available, obviously additional samples would ensure greater statistical accuracy.

Each sample was then polished using an automatic polisher on each of its flat faces (2). Each face was then etched using a separate etchant to ensure the best possible image for analysis. One face was etched with $\text{H}_3\text{PO}_4$ and one face was etched with $\text{HBF}_4$. The samples were then placed in a scanning electron microscope and images were taken of the resulting grain structure. These images were then post-processed using the standard line-intercept method for both secondary dendrite arm spacing and grain size.

ZINC SUMP

In order to better understand the shape of the solidifying interface present during the solidification of a rolling slab ingot, zinc sumps were made on an identical cast made on a separate day. For these sump profiles, 25 Kg of zinc was melted in a laboratory furnace and poured into the downspout of the casting ingot. The distribution bag, aided by the superior density of zinc compared to aluminum, then distributed the zinc relatively uniformly over the solidification surface.

After the cast, the section with the zinc poisoned sump was removed. The ingot section was then sliced into cross sections each 25 mm perpendicular to the rolling face. Each of these plates was then machined flat by a diamond fly cutter at a relatively low traverse speed to ensure a clean finish. Coffer dams to contain etchant were built using 2” aluminized tape run around the perimeter of each plate. A
solution of dilute sulfuric acid was then used to etch the surface of each plate and reveal the zinc poisoned interface. Each surface was then neutralized and the position of the interface was measured with a T-square. The coordinates of the interface from each slice were then input into a 3D drafting software package, and the rendered surface was generated.

Figure 13: Image of zinc sump after etching, note marks in the lower region used to determine interface height

THERMAL BOUNDARY CONDITION

In order to generate a numerical model for the casting conditions used during the experimental trials, an experiment was derived using that of Rappaz et al as reference[70]. During the cast used to determine the shape of the solidification interface, an additional thermocouple rake was fabricated. This rake consisted of a set of five K-type thermocouples spaced 5mm apart. Once the cast reached steady state (~1200mm) the rake was lowered into the melt pool perpendicular to the mold bore with one rake placed at the surface. The rake was “swallowed” by the casting, and temperature data was again recorded at a rate of 1 hz or 1/mm. After the casting was complete, the temperature data was analyzed and processed according to the algorithm developed by Rappaz et al in order to determine the necessary heat flux boundary condition.
RESULTS

MACROSEGREGATION

The OES composition data for copper was normalized to the furnace composition using the equation:

\[ Macroseggregation = \left( \frac{C - C_0}{C_0} \right) \times 100 \]  

(2)

The macrosegregation value was then plotted against its relative position in the first quadrant (see Figure 12) and the macrosegregation surface was generated, which can be seen in Figure 14.

![Figure 14: Surface plot of normalized copper macrosegregation. Colored by % deviation from furnace composition.](image)

For reference, the lower right corner is the center of the ingot, and the upper boundary is the rolling face. If one were to draw a line from about 650mm from the short face, directly to the rolling face and analyzed the macrosegregation of that line, one could see many of the structures present in Figure 4, namely the negative centerline segregation and the positive segregation at the mid-thickness. What is missing around the periphery of the ingot is the enriched surface and depleted subsurface region. These structures are millimeters in thickness, and our sampling was too low of resolution to be able to capture
these effects. In general, what we can see is that there is a depleted region in the center of the ingot that extends from the center-point to approximately 300 mm from the short face and is approximately 35 mm thick. A parallel enriched zone then extends to the same distance from the short face, between the depleted region and the rolling face. Its width is approximately 125 mm. Between the enriched zone and the rolling face, there is very little in terms of macrosegregation, it is nominally the same as the furnace. What is also interesting about this plot is that in the region extending from the short face, we do not see any of the “W” pattern. By symmetry, one would expect to see a similar “W” pattern extending from each of the chilled faces, thereby forming a box of enriched around a central, depleted core (see Figure 15). Instead of this enriched region, the entire area adjacent to the short face is characterized by effectively no noticeable macrosegregation.

Figure 15: Expected Macrosegregation due to symmetry arguments.

GRAIN SIZE

The grain size results were analyzed as mentioned previously in the experimental procedures. The evolution of the grain size (and dendrite arm spacing) is represented below in Figure 16. These plots trace the microstructure from the center of the short face, along a straight line from the center of the short face, to the center of the ingot. These microstructures only represent a single trace. In addition to these images, microstructures were analyzed for grain size along all of quadrant 1. The surface plot of the grain size for the ingot cross section can be seen in Figure 17.
13a: Microstructure at center of short face

13b: Microstructure 109mm from short face

13c: Microstructure 218mm from short face

13d: Microstructure 328mm from short face

13e: Microstructure 437 mm from short face

13f: Microstructure 546mm from short face
Figure 16: Microstructure of ingot sections from a line traced from the center of the short face to the centerline of the ingot. Samples have been etched with HBF4.
The surface plot for grain size (Figure 17) has several interesting characteristics. Areas adjacent to surfaces, both rolling face and short face, are characterized by the smallest grain size. Moving towards the center, there is a gradual increase in grain size. Perhaps the most defining feature of this plot is the region of extremely large grains that extends parallel to the rolling face, towards the short face. Immediately noticeable about this structure is that it is not in the center of the ingot, instead it lies in a region approximately 75mm from the centerline.

DENDRITE ARM SPACING

In addition to grain size, the same microstructures from Figure 16 were analyzed for secondary dendrite arm spacing. Once analyzed, the data was again plotted against position to form a surface plot similar to Figure 14 and Figure 17. This surface can be seen in Figure 18.
Figure 18: Surface plot of secondary dendrite arm spacing. Colored by DAS in um

Similar to the plot for grain size, the plot for secondary dendrite arm spacing is characterized by a zone of minimum spacing directly adjacent to any chill face. Moving towards the center, the dendrite arm spacing gradually increases. However, much like the plot for grain size, the region of highest DAS is not in the center, but instead is found in a region parallel to the rolling face, approximately 75 mm from the centerline (short face to center in x-direction).

ZINC SUMP

The primary goal of the zinc sump was to gain a better understanding of the solidification interface as it shaped in a rolling slab ingot. After the slices were measured (Figure 13) the entire surface was generated using Solidworks. The full profile can be seen in Figure 19. The green exterior is a transparent model of the surface of the ingot as it solidifies. Figure 20 represents half of the zinc sump, as sliced along the centerline parallel to the rolling face. Perhaps the most notable feature of this plot is the gentle curve of the solidification interface as it emerges from the short face, and then changes curvature approximately midway to the bottom of the sump. Figure 21 is similar to Figure 20, except the slice is made at 90° to the former, thus bisecting the rolling face instead of the short face. In this cross section one can clearly see the deep “V” shaped sump profile, formed when the two solidification interfaces impinge on each other, the intersecting plane is known as the “plane of thermal convergence”. It is easy to see that the curves would have executed the same smooth “S” curve as those exhibited by the short face, had the two interfaces not met. Figure 22 imposes the two cuts made in previous figures simultaneously. This is the solidification interface from quadrant 1, where the samples previously discussed were removed.
Figure 19: Full zinc sump profile as measured from a zinc poisoned ingot
Figure 20: Half of zinc poisoned sump, sliced at the centerline of the short face and parallel to the rolling face (AAA Cross section)

Figure 21: Half of zinc poisoned sump, sliced at the centerline of the rolling face and parallel to the short face (BBB Cross Section)
Figure 22: Quarter 1 of zinc sump, corresponding to solidification interface where samples were taken (AAA+BBB Cross Section)

TEMPERATURE PROFILE
The temperature profile data gathered during casting is organized into a single plot seen in Figure 23. What is worthy of note is that in all cases except the surface, the thermocouples remain isothermal for a certain period before gradually descending through the liquidus and solidus temperatures for this alloy.

Figure 23: Raw TC data gathered from Al4.5Cu Cast. Thermocouples are those along the cross section extending from the short face, parallel to the Rolling face
Combining these thermocouple measurements into a single plot, and spacing them appropriately, we can generate a temperature profile for this same cross section. This has been visualized in Figure 24. Here we can see that the rough shape of the profile is similar to that of the zinc sump (ignoring resolution difficulties). The isothermal zone corresponds to the liquid pool, which corresponds to the same isothermal regions above. The aforementioned dip in temperature corresponds to the liquidus isotherm, where the first solid crystals begin to form.

![Figure 24: Thermal profile of cross section extending from short face, parallel to rolling face](image)

DISCUSSION

It is perhaps easiest to begin the discussion with the macrosegregation profile and its most obvious feature, the depleted centerline zone. It is this region that incited the majority of macrosegregation funding throughout history. Used primarily as an aerospace alloy AA2024 is composed of 4.5% copper and several other elements to generate hard precipitates when the alloy is aged properly. When the alloy is rolled to thin enough gage, the centerline segregation can be ignored for the most part. However, when attempting to make plate stock out of AA2024, hardness measurements over the cross
section were non-uniform and traced the typical macrosegregation profile found in Figure 4\textsuperscript{71}. This variation in hardness through the cross section is generally attributed to the solute depleted zone in the center of the ingot.

Figure 25 is a superposition of the zinc sump profile for quadrant 1 viewed parallel to the rolling face, superimposed on the macrosegregation surface from Figure 14. It is important to note the axes of the two figures. While Figure 14 is drawn in the XY plane, Figure 20 is drawn in the XZ plane. By superimposing the two figures we are only able to ascertain information regarding one degree of freedom from this figure, notably an X coordinate. By marking the X coordinate of the end of the depleted zone, we can transfer the X coordinate to the zinc sump.

![Figure 25: Superimposed Macrosegregation Surface and zinc sump profile parallel to rolling face](image)

Performing an operation similar to Figure 25, Figure 26 is the same macrosegregation surface, but with the zinc sump from quadrant 1 viewed parallel to the short face superimposed. Once again, the macrosegregation surface is viewed in the XY plane, but the sump is viewed in the YZ plane. The superposition of the two images grants us the ability to discern one degree of freedom, in this case the Y coordinate of the boundary of the depleted zone. If we then take the X and Y coordinates removed from Figure 25 and Figure 26 respectively we can outline the solute depleted zone with an X and Y coordinate system as seen in Figure 27. By using the same coordinates, and plotting them on the zinc sump profiles, now properly oriented to account for gravity we can see how these lines properly line up with the zinc sump profiles, as reproduced in Figure 28. What we see here is that the region of depleted zone in the center of the ingot corresponds precisely to the lowest region of the sump, indicated in Figure 28 by red
arrows. This observation is consistent with one we would expect due to the sedimentation of primary phase grains to the bottom of the sump. This analysis makes no suppositions of the origins of the particles. Due to the convective currents present, these grains could have easily originated in the slurry zone of the mushy zone, or on the liquid surface. All that this confirms is that the negative centerline segregation forms in the lowest region of the sump, where one would expect to find precipitated grains. It is also worthy of note that no “daisies” were observed in the center region as can be seen in Figure 16. “Daisies” were typically associated with the presence of sedimenting grains historically, but with grain refined castings, such as that currently under investigation, growth of large crystals would have been restricted by the overwhelming rate of heterogeneous nucleation. Thus, the absence of observed floating grains is not a cause for concern.

![Figure 26: Superimposed Macrosegregation Surface and zinc sump profile parallel to short face](image-url)
The natural continuation of the discussion is the solute rich region found between the solute depleted center and the rolling face, seen in red in Figure 14. The Y coordinate from the previous analysis marks the end of the solute depleted center as we move towards the rolling face. However, this point also marks the beginning of the enriched zone. By imposing this line on the plots for grain size (Figure 29),
and dendrite arm spacing (Figure 30), we see that the beginning of the enriched zone corresponds to
the region of largest grain size and dendrite arm spacing. Given that, dendrite arm spacing and grain size
are driven by different mechanisms, normally one could not make a correlation between the two.
However, in the case of grain refined alloys, the distinction between grains and dendrite arms becomes
extremely difficult to discern especially in regions of slow cooling, such as the center of a solidifying
ingot. In such cases, one normally refers to these structures as cells, since it is not truly dendritic and
grains and dendrite arms are not distinguishable. Given that such conditions exist in the center of an
ingot, it is necessary to observe both the grain size and dendrite arm spacing, because in the center they
reflect more accurately a cell size and should indicate similar trends. Recalling the discussion on
shrinkage induced flow from the introduction equation 1 is the scaling law derived by Eskin et al. In his
analysis, Eskin assumes the permeability to be constant through the thickness of the billet, and groups
this into term A. In reality, the permeability of the mushy zone can be approximated by the Kozeny-
Carman relationship\textsuperscript{72-75}:

\[
K = \frac{(1-g_s)^3}{\kappa_{KC}S_v^3g_s^3}
\]

(3)

Where \(g_s\) is the solid fraction, \(S_v\) is the specific surface area of the solid phase, and \(\kappa_{KC}\) is the Kozeny-
Carman constant, which has been determined to be 5 for equiaxed structures\textsuperscript{72}. In modeling

\begin{figure}
\centering
\includegraphics[width=\textwidth]{figure29.png}
\caption{Grain Size surface plot, marked with boundary of depleted/enriched zone from macrosegregation surface}
\end{figure}
applications, permeability is analyzed as a continuum, from 0 fraction solid to 1. However, what many models fail to account for is the difference in microstructure during the evolution of a casting. While there is little data for the multitude of specific grain structures, it is generally assumed that grain refined alloys are equiaxed thus the Karman-Cozeny constant remains the same. What actually evolves during a casting is the morphology of the grain. As a grain goes from dendritic to cellular or globular, the specific surface area decreases dramatically. We can see the effect of this change in specific surface area in equation 3. \( S_p \) is an inverted square term, implying that a change in microstructure can dramatically change the permeability, even for a given fraction solid. Using the dendrite arm spacing as a representative quantity for the change in microstructure, the inverse of the dendrite arm spacing can be used to roughly illustrate the associated change in specific surface area with grain evolution. While numerically these approximations are in no way correct, the trend is correct. Depending on the magnitude of the microstructural evolution this quantity could also be squared however, for the present discussion such modifications would not modify the spatial permeability map. Making the aforementioned approximation for the change in specific surface area, and assuming an arbitrary fraction solid (0.2), the permeability can be mapped in the XY plane. Figure 31 is the permeability map for quadrant 1, using the above approximations. Comparing Figure 31 to Figure 30 and Figure 29 we can see that the line of maximum permeability corresponds to the lines of maximum dendrite arm spacing and grain size. Referring back to Equation 1, this change in permeability issues a spatial dependence on the parameter \( A \), increasing it in areas of increased permeability. The consequence of increased permeability can be obtained directly from Equation 1, such an increase results in an increase in solute transported through the mushy zone towards the solidification interface. It can be seen in comparing
the macrosegregation profile to the permeability map that the line of maximum permeability roughly corresponds to the boundary of the solute rich region (Y1 in both figures). This result is not surprising, given that the increased solute transport present inherently at this “boundary” would draw material into the mushy zone towards the rolling face. As the solid fraction increases towards the solidus isotherm, the permeability decreases correspondingly, which would cause the solute rich region to taper off in the direction of the rolling face, an effect visible in Figure 14.

![Permeability map](image)

Figure 31: Permeability map calculated from dendrite arm spacing. Maximum permeability has been parked by line Y1

The final discussion point remaining is the absence of a solute rich band adjacent to the short face of the casting. The previous analysis could be used to explain the absence of solute, due to the lack of cells large cells found in this region. While it is true that the maximum cell size is located at the border of the depleted zone and the enriched zone, the superior cell size in that location is due to the steeper slope of the solidification front, and thus slower cooling rate. Figure 17 and Figure 18 both indicate that even with a shallower sump profile, and slower solidification rate, there continues to be a gradual transition to a larger cell size moving along the centerline from the short face to the center. Figure 14 however, does not predict a correspondingly diminished enriched zone in this area. In order to understand this phenomenon, a numerical model was generated in order to understand the complex fluid dynamics at play in a solidifying ingot.

ANSYS Fluent was used as a computational fluid dynamics package (CFD) to construct a numerical model of the solidification process. FLUENT uses a finite-volume method (FVM) to solve the governing equations for fluid flow. FVM discretization is based on an integral form of the partial-differential equations (PDE) to be solved in the given problem. The PDE is written in a form such that it can be solved for a given unit or cell. The domain to be analyzed is previously discretized into a series of cells or units, which are the aforementioned cells used to solve the PDE. The resulting solution matrix usually involves fluxes from one cell to another, and thus flux calculation is very important to the FVM.
Additional literature on the structure of ANSYS code and the finite volume method can be found elsewhere, and the reader is directed to other sources for this information\textsuperscript{77}.

The model constructed in the FLUENT software package incorporated the necessary fluid parameters for analysis of a phase change calculation, they have been summarized below in Table 1. Fluent is capable of analyzing solidification phenomena, by accounting for the gradual transition from liquid to solid by using a Scheil equation interpretation based on temperature. In order to accurately model macrosegregation, variations in permeability and solute rejection would need to be accounted for. Since the subject of this investigation was simply an understanding of the fluid dynamic conditions, the momentum sink term generated by the fraction solid was judged to be sufficient. The mold geometry size was specified to be 650mmx1750mm. The casting speed was set at 60mm/min, and the mass flow inlet condition was calculated to reflect the change in liquid/solid density with zero mass accumulation.

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Table 1: Thermophysical Properties of Al4.5Cu Alloy

Boundary conditions for the solidifying ingot were specified using the heat flux data obtained and analyzed per the experimental procedure and the algorithm developed by Rappaz et al\textsuperscript{70}. This data has been represented below in Figure 32.
Initially, it was assumed that the heat flux condition was uniform around the perimeter of the ingot, the resulting solution predicted bleedouts at the short face.

Given the improbability of this solution, an additional experiment was performed in order to specify the boundary conditions at the short face. It was determined that the shape of the heat flux boundary was identical to that found in Figure 32, but multiplied by a factor of 1.35. That is, the ratio of cooling between the short face and the rolling face is 1.35. The corrected model was solved, and a plot of temperature profiles (colored by temperature) was generated with velocity vectors (scaled by velocity) using the software’s post-processor. This result is found below in Figure 33. A cutout section has been brought out for ease of discussion.

Bringing the discussion back to the absence of an enriched region adjacent to the short face, we can use the results of the model to understand this result. As can be seen in the figure below, the molten metal exits the combo bag and immediately extends along the surface towards the short face. As the metal stream impacts the short face, a portion of the flow redirects in the Y direction (into the page) and then back along the rolling face in the positive X direction (not seen). The balance of the flow redirects in the z-direction, scouring the solidification interface and driving fluid parallel to the sump. The result of this flow is twofold, first, that this scouring motion is a potential source of sedimenting grains. As the fluid scours the solidification interface, dendrite arms as well as grains in the slurry zone are removed from the mushy zone and transported with the fluid. It is then possible that such grains grow and
preferentially sediment to the center, forming the inverse centerline segregation. The second result, is that as the fluid flows in the Z direction, parallel to the solidification front, the rejected solute is drawn with it. The consequence of this action is that even with an increased permeability, the material drawn into the mushy zone through shrinkage flow is nominal in composition, instead of enriched. Thus, as the ingot progresses, no solute rich region is formed adjacent to the short face. The absence of such a flow on the rolling face therefore does not inhibit the formation of an enriched zone.

Figure 33: CFD solution with temperature profiles, and fluid velocity vectors scaled by relative velocity.

CONCLUSION
Macrosegregation in the DC casting of rolling slab ingots has been investigated, using a combination of experimental and computational methods. A macrosegregation profile through the cross section of an
industrial rolling slab ingot, shows that macrosegregation in such castings cannot be properly illustrated through a line plot. The complex interplay between fluid flow and precipitation becomes magnified in such large scale. In a first modeling attempt of our results, we showed that it is possible for the fluid dynamic conditions of the mold to affect grain and solute distributions within a casting. We have also shown that the non-uniform heat transfer conditions from the short face to the rolling face can contribute to a layered structure within rolling slab ingots. This information indicates that cooling conditions within the mold could continue to modify macrosegregation profiles. More work is underway to understand the effects of fluid flow within the mushy zone, its effects on grain and solute transport, and how such phenomena can affect macrosegregation.

SUGGESTIONS FOR FURTHER WORK
The complex interplay between the various mechanisms of macrosegregation in rolling slab ingots, make a complete analysis very difficult to formulate, as many mechanisms have conflicting effects. That said, there are numerous groups currently working on modeling codes to be able to predict macrosegregation structures for a variety of alloys. While this work is ultimately very useful to the industrial practitioner, models being able to adapt microstructure based on cooling condition and adapt permeability relations reflecting this are still too computationally expensive to attempt on the large scale. In the mean time however, there is a significant gap in the literature regarding the understanding of precipitated grains. Most notably where they originate from, but also how any movement does occur. Grain motion inside of the mushy zone could be on a continuum basis, where those sliding down the solidification front are immediately replaced on the upper end of the slope, or it could be the result of a cataclysmic event such as an avalanche. It is answers to these questions that would most notably improve models immediately, and are worth additional evaluation.
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