MOULD TAPER FOR HIGH SPEED
CONTINUOUS CASTING OF STAINLESS
STEEL BILLETS

BY

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Abstract

Over the past three decades, continuous casting has emerged as a dominant steel production technology. Global competition and customer expectations are driving the mini-mills to improve billet quality and increase productivity through increases in casting speed. Although high speed casting trials are being carried out at many mini-mills, there is still a lack of fundamental understanding about the influence of casting speed on mould heat transfer. At the core of the technology is the water-cooled, oscillating copper mould. Mould interaction with the billet, both thermal and mechanical, governs billet quality and productivity. Heat transfer under high speed casting conditions needs to be determined in order to design the taper of the mould.

The main objectives of this study were: to quantify the thermal-mechanical response of the mould in the casting of austenitic stainless steel and martensitic stainless steel with mathematical models; to evaluate the mould-billet interaction using mathematical models; and to provide practical recommendations for optimum mould taper design in high speed billet casting. Measurements were conducted on an operating billet casting machine at Atlas Steel to determine mould-wall temperature profiles for different steel grades, different casting speeds, mould flux types, oscillation frequency, off-center nozzle operation. The trial also involved data acquisition on mould displacement, casting speed, metal level. Samples of the billets cast at the trial were collected and process variables were recorded. An inverse heat conduction model was utilized to determine mould heat flux from measured mould wall temperatures and existing mathematical models were
utilized to investigate mould/billet interaction and mould taper using the heat flux as input.
Results from plant measurements, mathematical models and billet sample evaluation were
used to correlate mould thermal response with transverse and longitudinal depressions,
oscillation mark depths for different steel grades.

It was found from this work that mould heat flux is strongly influenced by casting
speed with mould flux lubrication. Increasing casting speed will consistently increase the
heat flux while the shell thickness will be reduced owing to the reduction of the residence
time of the strand in the mould.

This study has also shown the effect of casting conditions such as steel grade,
superheat and casting speed on mould taper requirements. Optimum mould tapers based
on the evaluation of the interaction between the mould and the billets are recommended
for austenitic and martensitic stainless steel billet casting.
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LIST OF SYMBOLS

$A, B$ internal dimensions of the mould (mm)

$C_p$ specific heat (J kg$^{-1}$C$^{-1}$)

$C_{pw}$ specific heat of water (J kg$^{-1}$C$^{-1}$)

$d_w$ width of cooling-water channel gap (m)

$D_H$ hydraulic diameter (m)

$f_s$ fraction of solid

$G$ rate of energy generation or consumption (W/m$^3$)

$h_c$ heat transfer coefficient at the mould-powder/air interface (W m$^{-2}$C$^{-1}$)

$h_r$ effective heat transfer coefficient due to radiation (W m$^{-2}$C$^{-1}$)

$h_w$ heat transfer coefficient at the mould/cooling-water interface (W m$^{-2}$C$^{-1}$)

$H$ enthalpy (J/kg)

$H^{\text{new}}$ nodal enthalpy at time $t+\Delta t$ (J/kg)

$H^{\text{old}}$ nodal enthalpy at time $t$ (J/kg)

$\Delta H$ latent heat (J/kg)

$I$ identity matrix

$k$ thermal conductivity (W m$^{-1}$C$^{-1}$)

$k_c$ thermal conductivity of continuous phase (W m$^{-1}$C$^{-1}$)

$k_{cu}$ thermal conductivity of mould copper walls (W m$^{-1}$C$^{-1}$)

$k_d$ thermal conductivity of discontinuous phase (W m$^{-1}$C$^{-1}$)
\( k_{\text{eff}} \) effective thermal conductivity (W m\(^{-1}\) °C\(^{-1}\))

\( k_w \) thermal conductivity of cooling-water (W m\(^{-1}\) °C\(^{-1}\))

\( L \) mould internal side-length (m)

\( L_w \) length of the cooling-water channel (m)

\( L^0 \) initial nodal length (m)

\( \Delta L \) billet shrinkage (m)

\( \text{Pr} \) Prandtl number

\( q \) mould heat flux estimated from inverse analysis (W m\(^{-2}\))

\( \mathbf{q} \) vector of estimated mould heat fluxes

\( q^0 \) initial mould heat flux for inverse analysis (W m\(^{-2}\))

\( \mathbf{q}^0 \) vector of initial mould heat fluxes

\( q_e \) convective heat flux losses at top of mould-powder (W m\(^{-2}\))

\( q(z) \) mould heat flux profile (W m\(^{-2}\))

\( \text{Re} \) Reynolds number

\( S \) sum of squares

\( t \) time (s)

\( t_N \) negative-strip time (s)

\( T_a \) ambient temperature (°C)

\( T^c \) calculated mould temperature for the inverse analysis (°C)

\( T^m \) measured mould wall temperatures (°C)

\( T^0 \) temperatures calculated using the initial mould heat flux \( q' \) (°C)

\( T_0 \) temperature at the cold face of the mould (°C)
$T_w$  temperature of cooling-water ($^\circ$C)

$T_{w_{\text{inlet}}}$  cooling-water inlet temperature ($^\circ$C)

$\text{CSp}$  casting speed (m/min)

$V_c$  cooling-water velocity (m/s)

$X, Y, Z$  spatial coordinates

$\eta_w$  viscosity of water (Pa s)

$\mu$  mass attenuation coefficient for gamma rays (cm$^{-1}$)

$\rho$  density (kg m$^{-3}$)

Subscripts

$i, j, k$  dummy summation indices

$l$  liquid phase

$\gamma$  $\gamma$-iron phase
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Chapter 1 INTRODUCTION

Stainless steel has been continuously cast for more than four decades.\cite{1} The first major installation in North America for the continuous casting of stainless steel slabs was established in the early 1950's at Atlas Steel in Canada.\cite{1} Over the years, the continuous casting process for the production of steel billets, blooms, and slabs has emerged as one of the most important technologies in the steel industry worldwide,\cite{2} particularly in the stainless steel sector. A revolutionary development in refining in the 1970's, namely the invention of the AOD, followed by the development of similar novel processes such VOD, paved the way for the production of high quality stainless steel at reduced cost. Stainless steel is now manufactured by the following route:

\[ \text{Steelmaking} \rightarrow \text{Refining} \rightarrow \text{Rinsing} \rightarrow \text{Continuous casting} \]

where, following melting in an electric furnace, the molten steel is refined in an AOD or VOD furnace and then stirred to homogenize the temperature before continuous casting.\cite{3-7}

The different grades of stainless steel that are continuously cast can be classified in terms of their microstructure as ferritic, martensitic, austenitic and duplex,\cite{8-11} which makes the processing more complex than that of plain carbon grades. The chemical reactivity, physical properties, and solidification modes differ widely for the various grades.
of stainless steel. Although stainless steels have been continuously cast for about four decades, the products are not entirely free from defects. Problems with respect to cleanliness, the depth and uniformity of oscillation marks, segregation and centerline porosity, surface and internal cracks, and shape defects, such as longitudinal and transverse depressions, continue to plague the industry. Although considerable progress has been made in improving the quality of semi-finished sections through technological developments such as new tundish designs, and EMS in the mould and sub-mould region, the goal of producing defect-free products on a consistent basis has not been fully realized. Driven by the need to significantly reduce costs without compromising quality, new developments will continue well into the next century and have heralded a new era in continuous casting.

In view of quality and productivity in billet casting, the importance of heat transfer in the mould and the interaction between the mould and the strand can never be overstated. Heat transfer from molten steel to the walls of the continuous casting mould is controlled, in large part, by conduction across the air gap that forms as the solid shell shrinks. The air gap accounts for as much as 80–90% of the total heat resistance to heat flow. To compensate for the billet shrinkage, the mould walls are tapered inwards; the resulting reduction in the air gap improves the rate of mould heat extraction and decreases the surface temperature of the billet at the mould exit and thereby increases the thickness of the shell. Lack of sufficient taper, additionally, can lead to the formation of off-corner cracks. On the other hand, an excessive taper can cause difficulty in the withdrawal of the strand which promotes mould wear, and in extreme cases, causes the billet to jam in the mould. The quantification of the strand-mould gap is thus a primary step toward
defining mould taper. The gap, however, is a complex function of several variables such as steel grade, casting speed, superheat etc., which renders it extremely difficult to characterize.

The earliest work published on mould tapers for casting billets is that of Dippenaar et al.\textsuperscript{[16]} From then on, the importance of mould taper design has caused increasing attention of researchers in the field of billet casting, especially for high speed billet casting.\textsuperscript{[17-19]} The calculation of mould taper has been made possible in the Billet Casting Research Group in UBC with corresponding mathematical models.\textsuperscript{[1,12,14]}

In this study, Atals Steel was chosen as the location for the plant trial because very little data is available on heat transfer during continuous casting of stainless steels. Temperature data taken from an instrumented mould have been converted to axial heat flux profiles using a mathematical model called INVERSE developed by UBC. These heat flux profiles vary with the grades of steel, casting speed and other operating parameters. Then other models have been employed to quantify mould distortion, billet shrinkage and finally, mould tapers for different grades of steel. The calculations have been compared with industrial mould taper currently in use, and recommendations have been made for new mould tapers.

This study has helped understand the factors that influence mould heat transfer, especially in the casting of stainless steel, and thereby has laid the ground for the design of mould tapers for these steel grades.
Chapter 2 LITERATURE REVIEW

The mould is the heart of the continuous casting machine in many respects. Being the primary heat extraction device, it is required to solidify a shell of sufficient thickness to contain the liquid core in the submould regions. The mould also provides support for the newly solidified steel shell at a time when the shell is weak and susceptible to bulging due to ferrostatic pressure. The surface quality, shape and even internal quality of the solidified steel are profoundly influenced by the design and operation of the mould. For example, in the casting of billets, rhomboidity, longitudinal & transverse depressions are the result of mould behavior. This chapter is devoted to the mould and presents current knowledge on mould design, mould related defects, solidification of stainless steel, mould distortion with an emphasis on heat transfer.

2.1 Description of the Mould Assembly

Mould response significantly impacts the quality of continuously cast steel billets. Continuous billet casting moulds are typically square copper tubes, approximately 0.8 m in length. Internal mould dimensions range from 114 to 254 mm, with the mould wall thickness varying from 9.5 to 19 mm. Figure 2.1 illustrates the mould assembly. The mould is installed in a steel jacket which supports the mould and contains the cooling water. The mould is secured by plates that fit into slots cut in the outer surface of the tube,
near the mould top. On top of the mould, plates secure the assembly, seal the cooling water channel and serve as part of the oil distribution system. Inside the steel jacket, water baffles are typically placed within 4 mm of the mould to facilitate a high cooling water velocity. Cooling water enters the gap between the mould and water baffle at the bottom of the mould. At the top of the assembly, the water is routed to the back of the water baffle, before it exits to the cooling water system.

Billet moulds are typically used for several hundred heats; an average heat is approximately 45 to 90 minutes in duration. The moulds are then replaced because of mould wear, distortion or damage. Poor lubrication or mould taper design causes mould wear, usually at the bottom of the mould where the steel shell abrades the mould. Moulds also permanently distort with use, indicating that the thermal stresses in the mould sometimes exceed the yielding limit of the copper. Studies have shown that the mould taper slowly changes with mould use.

2.2 Heat Transfer in the Mould

Heat is extracted, from the liquid steel to the mould cooling water, through the following path:\textsuperscript{[19]}

- Convection in the liquid steel
- Conduction through the solid steel shell
- Conduction and radiation through the mould-shell gap
- Conduction through the copper mould
- Convection to the cooling water

As the billet solidifies and shell grows, heat transfer varies across the mould face, both vertically and horizontally. The mould shell gap is the largest thermal resistance to heat
transfer, particularly near the meniscus. Lower in the mould, the thermal resistance of the solid steel shell may provide a comparable resistance. Thus heat extraction lower in the mould reduces because of the increased total thermal resistance. The corners solidify more

Figure 2.1 Schematic of a typical billet mould assembly[56]
quickly because of two dimensional cooling from two faces of the mould. The corners then shrink away from the mould wall more quickly than the mid-face, contributing further to non-uniform heat transfer. Based on the examination of solidification bands, the influence of non-uniform corner cooling extends about 20 mm from the mould corners.

Mathematical models have been successfully used to quantify heat transfer between the billet and the mould. Mould heat transfer coefficients have been estimated based on changes in mould water cooling temperature; subsequently, the heat transfer coefficients were defined based on dwell time in the mould. In later studies, thermocouples were installed in industrial mould to obtain on-site temperature. Axial heat-flux profiles were then calculated from the thermocouple data, using finite-difference models. Once mould heat transfer was quantified, the corresponding heat flux could be applied to a billet solidification model. Finite difference models were again appropriate for estimating the temperature distribution in the shell and shell thickness as a function of position in the mould.

In view of heat transfer, the mould can be divided into two zones: an upper region in which heat extraction can be influenced by factors altering the gap width or gap conductivity like taper, mould distortion, lubricant type and flow rate, and a lower region of gap and shell resistance dominance in which the heat extraction can be influenced by factors like casting speed (alter the thickness) and taper (change gap).

2.2.1 The Influence of Chemical Composition on Heat Transfer

Singth and Blazek demonstrated a heat transfer dependence on carbon content. Overall heat transfer was a minimum for 0.1 pct. carbon steels and was relatively constant
for grades above 0.3 pct. carbon. The surface of billets containing approximately 0.1 pct. carbon are rough, characterized by wrinkles and indentations. The low heat flux associated with casting these steels was attributed to this rough surface. A small decrease in heat extraction was also seen with high carbon steel (0.85 pct. carbon) compared with medium carbon grades. Grill and Brimacombe\textsuperscript{22} correlated the 0.1 pct. carbon heat transfer minimum with the lower limit of peritectic phase change, where the $\delta$ to $\gamma$ solid shrinkage is the greatest. The wrinkled surface is likely associated with a phase change instability of shrinking, gap formation and reheating. Grill and Brimacombe\textsuperscript{22} presented the following mechanism:

1. The solidifying shell in contact with the mould cools quickly and transforms from the $\delta$ to $\gamma$ phase.

2. The outer surface shrinks more than the inner surface, which is still $\gamma$ phase, causing inward bending of the shell.

3. The surface in the gap reheats because of reduced heat transfer across the gap, shell strength reduces locally, and ferrostatic pressure deforms the shell back toward the mould wall. The resulting surface is wrinkled, and retains its shape as the shell cools and strengthens.

Some researchers\textsuperscript{40} pointed out that austenite stainless steel with a ratio of Ni/Cr range from 0.55 to 0.60 have the minimum heat flux. The thickness of the shell is not uniform and the tendency to form depressions is high. The Si content has no significant effect on the average heat flux of the mould. It is said that in low carbon steel, the Al and Cr content has no significant influence on the average heat flux. The average heat flux increases with the Ni and S contents.
2.2.2 Mould lubricant

Two types of lubricant are employed in continuous casting: oil and mould powder. Oil lubricants lubricate the mould by wetting the mould wall; they then partially break down owing to the high temperature, and contribute to the atmosphere in the gap. Mould powders, on the other hand, simply melt and wet the steel, with the extent of wetting being controlled by interfacial forces. The difference in behavior between the two types gives rise to different patterns of heat extraction per unit weight, although the total heat transfer is very nearly the same.

When casting with oil lubrication, many gases exist in mould-shell gap in addition to nitrogen, including shrouding gases and components of combustion. The composition of this atmosphere is believed to impact heat transfer through the mould-shell gap. The pyrolysis of oil at the meniscus creates hydrogen, which has a thermal conductivity seven times greater than air.

Singh and Blazek\textsuperscript{[23]} carried out experiments in a continuous casting mould with horizontal water passages. They found, using mould powder as lubricant, that:

In the case of low-carbon steel (0.10\%C) the heat transfer increased just below the meniscus but decreased thereafter.

In the case of high carbon steel (0.40\%C) the heat transfer was significantly lower over the entire mould length.

2.2.3 Casting Speed

Many investigations have observed an increase in heat flux \textsuperscript{[24]} and mould wall temperatures \textsuperscript{[25]} with an increase in casting speed, as can be seen from Figure 2.2. Singh
and Blazek\textsuperscript{[23]} have reported that the increase in heat flux was less for a 0.10 carbon steel than for a 0.40\% carbon steel.

![Mould flux as a function of distance from top of mould for different casting speed\textsuperscript{[23]}](image)

Figure 2.2 Mould flux as a function of distance from top of mould for different casting speed\textsuperscript{[23]}

The average overall heat transfer coefficient has also been noted to increase with casting speed\textsuperscript{[61,62,63]} Although a definite relationship was not formulated it has been suggested that the increase is linear\textsuperscript{[26]} Hills\textsuperscript{[27]} has proposed that the functional relationship between dwell time (\textasciitilde distance below the meniscus/ casting speed) and heat transfer coefficient varies from mould to mould. Subsequently, Brimacombe et al\textsuperscript{[28]} have
demonstrated that the mean heat transfer coefficient can be related to the dwell time as follows, for different moulds:

\[ h_m = 1.696 - 0.0162 t_m \]

where \( h_m \) is in kW/m\(^2\)°C and \( t_m \) is in seconds.

Despite the increase in heat transfer rate with casting speed, it is important to note that the specific amount of heat extraction, J/kg\(^1\), decreases, resulting in a net decrease in shell thickness. Cliff and Dain\(^{[29]}\), for example, have demonstrated that a 10% increase in casting speed decreases the exit shell thickness by 5%.

The magnitude of temperature fluctuations in copper mould plates has been observed to increase with casting speed.\(^{[25]}\) Moreover, mould powders which flow unevenly between the shell and mould give rise to temperature fluctuations that increase markedly with higher casting speed. On the other hand, powders which provide a uniform slag film reduce the effect of casting speed on temperature fluctuations in the mould wall. This behavior is demonstrated in Figure 2.3.\(^{[64]}\) Powder A flows unevenly between the strand and mould, while C provided a uniform amount of powder to the mould gap.

### 2.3 Mould Taper and Its Application in High Speed Casting

#### 2.3.1 Air Gap

As the steel cools and shrinks, an air gap forms between the mould and shell. As previously mentioned, the air gap usually represents the largest barriers to heat extraction. Heat transfer from molten steel to the walls of the continuous casting mould is controlled, in large part, by conduction across the air gap. In the upper region of the mould the air gap is considerably less than a millimeter wide but, in many cases accounts for as much as 80-90%\(^{[19]}\) of the total resistance to heat flow.
The quantification of the strand-mould gap is thus a primary step towards defining mould taper. The gap is, however, a complex function of several variables and its width changes in both the longitudinal and transverse directions which renders it extremely difficult to characterize. The shrinkage of the billet is affected significantly by the grade of steel being cast, particularly the low carbon grades where the contraction accompanying the solid-state transformation from delta to austenite phase must be taken into consideration.

2.3.2 Mould Taper

To minimize the gap and improve heat transfer moulds are therefore tapered. The inward taper of the mould, which compensates for the shrinkage of the solidifying shell,
varies from no taper to single taper and double taper. Excessive taper caused the billet to bind in the mould; moderate taper can cause an increase in heat transfer at the mid-face. Although mould taper undoubtedly improves heat transfer and reduces the surface temperature of the billet at the mould exit, excessive taper increases the resistance to withdrawal and exacerbates mould wear. In the limiting case, if the taper is made too large, the billet can jam in the mould.

Conventional mould shows a maximum outward bulge near the meniscus.[2] The maximum bulge is located below the meniscus, so that the mould acquires a negative taper above it and a positive taper immediately below. The negative taper at the meniscus can cause defects, such as deep oscillation marks, non-uniform lubrication, and to certain extent the possibility for breakout.[65,66,60,67]

Since 1980, research[65,66,60,67] has revealed the strong influence of mould taper on the depth and uniformity of oscillation marks, and, as result, on off-corner squareness and off-corner internal cracks. More recent work also has shown that mould taper at the meniscus has a profound effect on the local and overall heat extraction from steel, with consequences for mould distortion, oil lubrication and billet surface quality.

In the first published study of billet mould taper, Dippenaar et al. [30] evaluated mould tapers by estimating billet shrinkage. The two-dimensional transverse heat transfer model originally presented by Brimacombe[31] was used to calculate the temperature field in the solidifying billet. The model assumes a constant thermal expansion coefficient. The researcher concluded that large gaps formed in the lower region of single-tapered moulds. Calculation based on axial profiles of measured mould heat extraction, shrinkage of the cooling solid shell and mould distortion have shown that double taper is desirable.
Chandra et al \cite{32,33} modified the model to include a thermal-expansion coefficient which was a function of temperature and carbon content. His work was particularly important when calculating the shrinkage of peritectic steels.

### 2.3.3 Mould Taper Design in High Speed Continuous Casting

In high speed casting, Danieli has developed a new technique\cite{34} The key points of the new technology are a new concept for cooling the mould and a new mould itself. This new mould is called adaptable, because it is able to “adapt” its original taper to the billet shrinkage. This effect is controlled by the pressure of the mould cooling water. High pressure, large pressure drop and higher water velocity in the cooling channel are used to improve the cooling conditions.

This innovative design allows a substantial increase in the casting speed. Moreover, the mould is able to cast a wide range of steel grades, without altering other process parameters. The first design of the adaptable mould (DANAM phase 1, with a length of 780 mm and normal thickness of copper tube) permitted the casting of 130x130 mm billets at 4.3 m/min, using open stream pouring. The billet quality results were equivalent to or better than for conventional continuous casting process. Measured thickness of the solid shell along the casting direction were similar to conventional casting at lower casting speed. It was reported that a thinner adaptable mould (DANAM phase2) of 1000mm length, for speed of 6 m/min, will commence operation in the second half of 1995. But details of the trial were sparse.

For the purpose of improving productivity and billet quality, joint research was conducted on high speed casting performance with a convex mould developed by Concast Standard AG\cite{35} Beginning in September 1992, a maximum casting speed of 3.5 m/min is
achieved. The billet quality was suspected to be superior which lead to the claim that the convex mould was a proven technology.

2.4 Solidification of Stainless Steel

2.4.1 Fundamentals of the Family of Stainless Steel

Stainless steels can be classified by their microstructure. The four main families of stainless steel according to this criterion are martensitic, ferritic, austenitic and duplex (austeno-ferritic). Ferritic stainless steels contain 15-30% Cr, sometimes even higher Cr contents, with alloying additions of Mo, Ti or Nb. They are fully ferritic up to the melting point. The transformable or martensitic steels contain 12-17%Cr and 0.1-1.5%C. Austenitic stainless steels are by far the most widely used stainless steels, comprising 70-80% of stainless steel production. They contain 16-25%Cr, 7-20%Ni and current developments have decreased the carbon content to well below 0.03%. Duplex stainless steels consist of virtually equal amounts of austenite and ferrite. Many austenitic stainless steels contain up to 20% ferrite, but the modern duplex steels contain 50% ferrite.

In the manufacture of stainless steel, continuous casting has become, after years of development,[69,70] the leading casting technology worldwide.[71] Further developments concentrate on improving strand surface quality to attain the optimum ‘non-conditioning’ ratio before hot rolling. This is achieved with relative ease for ferrite grades and martensitic grades, with their even shell growth and smooth surface appearance. Austenitic grades, on the other hand, tend to have non-uniform shell growth with accentuated surface roughness - mainly in the form of transverse depressions.[72,73] Furthermore problems of solidification cracking, transverse and longitudinal depressions
have been attributed to interaction between the mould and the strand accompanying primary austenitic solidification.

2.4.2 Modes of Solidification of Stainless Steel

1) Austenitic Stainless Steel

Austenitic stainless steels can solidify by several mechanisms or modes,
\[ \text{mode A: liq. } \rightarrow \text{L}+\delta \rightarrow \delta \]
\[ \text{mode B: liq. } \rightarrow \text{L}+\delta \rightarrow \text{L}+\delta +\gamma \rightarrow \delta +\gamma \]
\[ \text{mode C: liq. } \rightarrow \text{L}+\gamma \rightarrow \text{L}+\gamma +\delta \rightarrow \gamma +\delta \]
\[ \text{mode D: liq. } \rightarrow \text{L}+\gamma \rightarrow \gamma \]

where \( \delta \) and \( \gamma \) represent ferrite and austenite respectively.

These modes are illustrated schematically on a vertical section through the Fe-Ni-Cr phase diagram in Figure 2.4.

Figure 2.4 Section through Fe-Ni-Cr phase diagram at 19%Cr showing Solidification modes\(^{[68]}\)
Chapter 2 LITERATURE REVIEW

The solidification sequence and subsequent transformation characteristics will determine both the level of segregation and the distribution of residual ferrite. The ferrite present may be dendritic or interdendritic depending on the solidification mode. Segregation will be more deleterious in austenitically solidifying steels (modes C and D above), since segregation to grain boundaries will not be redistributed by solid state transformation as with modes A and B.

As we know, complex austenitic stainless steel compositions can be reduced to simple Fe-Ni-Cr ternary alloys by the use of Ni and Cr equivalent compositions. The large number of addition elements like C, N, Mn, Si, Mo, Cu, Nb (or Cb), V, Ta, Al, etc., can be divided into:

- δ-former elements, on account of their body-centered-cubic crystallographic structure like Cr. Such is the case for Mo, Nb, V and Ta. Silicon is considered an δ-former on the basis of the presence of a γ-loop in the Fe-Si binary diagram comparable to that of the Fe-Cr binary alloy.

- γ-former elements like nickel, namely C, N, Mn, Co, etc.

One set of equations which has proved successful for determining solidification sequences is that recommended by Jernkontoret:

\[
Cr_{eq} = \%Cr + 1.37(\%Mo) + 1.5(\%Si) + 2(\%Nb) + 3(\%Ti) \quad (2.4.1)
\]

\[
Ni_{eq} = \%Ni + 22(\%C) + 14.2(N) + 0.31(\%Mn) + \%Cu \quad (2.4.2)
\]

By using equivalent compositions, it is possible to apply the Fe-Ni-Cr phase diagram to the prediction of a solidification sequence. They found the following phase fields:
\[ \frac{\text{Cr}_{eq}}{\text{Ni}_{eq}} = 1.38 - 1.5 \] (mode C)

\[ \frac{\text{Cr}_{eq}}{\text{Ni}_{eq}} = 1.5 - 2.0 \] (mode B)

\[ \frac{\text{Cr}_{eq}}{\text{Ni}_{eq}} > 2.0 \] (mode A)

Modes A and B are primary ferritic, whereas modes C and D are primary austenitic.

Figure 2.5 Schematic linescans for solidification of austenitic stainless steel as mode B\textsuperscript{[68]}
The first type of linescan is shown in Figure 2.5. The characteristics are consistent with mode B solidification, i.e. primary ferrite with interdendritic austenite. In these casts, the dendrite centres were depleted in Ni and enriched in Cr, indicating primary ferrite precipitation. Within the austenite between the primary dendrites, the Ni content rose moving away from the dendrites and the Cr concentration fell. This indicates that, in this part of the structure, austenite began to be precipitated as a separate phase while the ferrite was still precipitating.

In addition to this general pattern, seen most clearly between the primary dendrite arms, there were also areas in which all elements increased rapidly in concentration. These represent the final stages of solidification, in which the composition balance in the last pools of liquid was such as to precipitate only austenite, and the levels of solute element rose rapidly due to segregation.

The second type of linescan is shown schematically in Figure 2.6 and is consistent with mode C solidification, i.e. primary austenite with interdendritic ferrite. The centres of the dendrites were depleted in all elements. The interdendritic spaces, however, contained ferrite precipitated in final stages of solidification, marked by elevated levels of ferrite formers, especially Cr, and reduced levels of austenite former.

It can be seen that there was an enrichment in the austenite formers around the ferrite and also in some cases a variation in concentration of austenite and ferrite formers within the ferrite. This implies that the austenite formers were being rejected from the ferrite as it is formed. If all the austenite had been fully solid at this stage, rejection of austenite formers in this way would have required a considerable amount of solid state diffusion, making the creation of such a large zone of enrichment difficult. Therefore, the
observation of such a zone indicates that the final stage of solidification consisted of austenite and ferrite precipitating simultaneously from the interdendritic liquid in a similar manner to that described in the first case above, rather than the final liquid solidifying as ferrite instead of austenite.

Figure 2.6  Schematic linescans for solidification of austenitic stainless steel as mode C

\[68\]
The third type of linescan is shown in Figure 2.7 and is consistent with mode D solidification, i.e. fully austenitic. All the elements showed similar distributions, being
relatively depleted in the centers of the dendrites and rising smoothly to a peak in the interdendritic spaces.

The level of segregation will be determined by the mode of solidification and also by the coarseness of the structure. In general, as a result of reduced solute solubility, steels solidifying initially as austenite will exhibit higher levels of segregation to the interdendritic spaces than steels solidifying initially as ferrite. Thus, the highest levels of interdendritic P segregation were found in some casts to be up to 0.08%, all of which solidified austenitically. A similar effect was not found in the high N series of steels where lower levels of P segregation (up to 0.02%) were observed in other casts, whatever the primary phase. This could be due to a strong interaction between N and Cr or Mo reduction segregation generally. Alternatively, it could derive from the much finer primary and secondary dendrite arm spacing resulting in faster homogenization during cooling. In general, dendrite arm spacing decreased with increasing alloy content.

The solidification mode controls the levels of segregation and the amount of distribution of residual ferrite. Thus, ferritically solidifying steels should exhibit less deleterious interdendritic segregation of residual elements by contain higher levels of ferrite than austenitically solidifying steels. In ferritically solidifying steels the residual ferrite will be dendritic, whereas in austenitically solidifying steels any ferrite will be interdendritic. This pattern of segregation and ferrite will be redistributed by solid state transformation on cooling below the solidus. It has been demonstrated that ferritically solidifying steel contains in excess of 80% ferrite at the solidus and by necessity this transforms to a room temperature structure containing less than 10% residual ferrite. This massive transformation results in isolation of any segregates from grain boundaries. The
transformation is absent in austenitically solidifying steels and, therefore, segregation and interdendritic ferrite, which by its nature is highly segregated, will remain on grain boundaries. Thus, it can be seen that ferritically solidifying steels are preferable for optimum castability and hot working characteristics.

2) Martensitic Stainless Steel

![Fe-Cr equilibrium diagram with carbon content](image)

**Figure 2.8 Variation of the Fe-Cr equilibrium diagram with carbon content**

Figure 2.8 shows the equilibrium diagram of Fe-Cr with different carbon content. Increasing the carbon level from 0.05 to 0.10% widens the gamma loop and two phase region. As we know that the addition of more than 12%Cr is necessary to obtain a martensitic stainless steel, the solidification mode of martensitic stainless steel from the equilibrium diagram of Fe-Cr is as follows:

\[
\text{liq.} \rightarrow \text{L} + \alpha \rightarrow \text{L} + \alpha + \gamma \rightarrow \alpha + \gamma
\]

which is relatively simple and cause less problems in the continuous casting process.
The diagrams in Figure 2.8 show that alpha ferrite can be present in certain compositions, and could be retained in the as-quenched structures.

2.4.3 Ferrite Distribution in the Austenitic Stainless Steel

As discussed above, the ferrite is frequently formed during the solidification of the stainless steel. The initial solidification can be either to ferrite or to austenite depending on the composition, and either peritectic, eutectic or solid state transformations can subsequently occur. In highly ferrite forming compositions, ferrite dendrites form and after solidification some austenite can precipitate as a Widmanstatten structure in the ferrite. With less ferrite forming compositions, ferrite dendrites form, and austenite may then solidify into the ferrite dendrites whilst after solidification austenite can precipitate from the ferrite. This type of solidification is in fact a peritectic reaction. With still less ferrite forming compositions, ferrite and austenite dendrites can occur together but on cooling the ferrite may be entirely dissolved or reabsorbed in the solid state by the austenite.

The variation of the delta ferrite content over the thickness of continuously cast slabs is generally due to two main factors. First, the cooling conditions between the liquidus and solidus temperatures determine the dendrite arm spacing. The dendrite spacing determines the diffusion distance for the peritectic and solid state transformation. The second factor is the cooling rate of the solid phase below the peritectic temperature. The cooling rate determines the diffusion time, which controls the extent of the diffusion controlled transformation to gamma phase. Thus the final delta ferrite content is influenced by the combined effect of dendrite arm spacing and diffusion time, although they have opposite effects.
The characteristic ferrite fraction, as a function of steel composition, can be obtained with the following relationship:\[^{[68]}

\[ f = 5.26 \left( 0.74 - \frac{\text{Ni}'}{\text{Cr}'} \right) \]  

(2.4.3)

where: Cr' equivalent contents for Cr and Ni can be calculated by equation 2.4.1 and 2.4.2.

Figure 2.9 Qualitative relationship between steel composition, fraction of primary ferrite solidification structure, and depression tendency for austenitic stainless steel\(^{[68]}\)

Based on equation (2.4.3), a qualitative relationship between steel composition, fraction of primary ferrite (and austenite), and the resulting depression tendency is proposed in Figure 2.9, together with the four principal types of solidification sequence and microstructure observed in austenitic stainless steels. Depression formation has a
maximum tendency for Ni'/Cr' between 0.55 and 0.60. On the other hand, fully ferritic grades, or ones with increasing austenitic solidification, show much more uniform shell growth. A peculiar composition exists for Ni'/Cr' = 0.65, where primary ferrite and austenite may form simultaneously.[74-75]

2.4.4 Hot Workability and ductility of Austenitic Stainless Steel

During solidification, stainless steels show marked columnar solidification which can impair hot workability. In general, ferritic stainless steels have high ductility which does not vary greatly with temperature, and they are quite soft and mark easily during rolling, giving rise to surface defects. On the other hand austenitic stainless steels have lower hot ductility than the ferritic steels at temperatures up to about 1200°C and the ductility is very temperature dependent. The ductility of austenitic stainless steel increases slightly as temperature rises from ambient, but declines to a minimum at 600-800°C. From there it rises steeply to a maximum (about 1230°C for 304 and 316[42] much larger than at room temperature), but then drops precipitously to zero near the solidus temperature T_s. This behavior is found in Fe-C, Fe-Ni, Fe-Ni-Cr and Ni-base alloys. The hot workability depends on[77-79]:

I. level of ductility, poor hot ductility can impair hot workability. Coarse austenite grains in the cast structure are detrimental to intergranular hot work cracking.

II. the temperature at which the ductility rapidly decreases, i.e. at which liquidation occurs or grain boundary cohesion is lost. This limits the maximum temperature of hot working. Boron, which not only segregates to austenite grain boundaries but also forms a low melting point eutectic, markedly lowers the solidus temperature,
as also does Nb which forms a eutectic of NbC similar, but less pronounced effect.

III. inclusion, which can initiate void formation and lead to lower hot ductility and poor hot workability, especially if they occur close to the surface when major surface defects can be produced.

IV. segregation, particularly of impurity elements into the interstices, which leads to intergranular separation. A little ferrite can be useful in dissolving some of these impurities, as in the case of weld metal cracking, but too much ferrite is very detrimental. So detrimental are certain elements that it is common practice to use scavenging additions such as Ca, Mg or Ce (rare earths) to improve hot workability by combing with such impurities.

For stainless steels, the temperature of the minimum and the failure strains are raised as strain rate $\varepsilon$ is increased form creep values up to a structure dependent maximum at about $1\text{s}^{-1}$. The intermediate temperature ductility minimum results from grain boundary fissures arising from intergranular sliding. Because of the high stresses, these wedge cracks originate at triple junctions due to stress concentrations on non-sliding boundaries normal to the tensile axis. The ability to relax such stresses by accommodating lattice flow decreases as the strength of the lattice is increased, or the recoverability decreased, by solid solution or particle hardening.

The as-cast material generally has a low ductility, fairly independent of temperature, which is markedly improved as it is progressively homogenized and refined by working. The poor properties stem from the macrostructure of coarse grains, segregation of impurities, inclusions, $\delta$ phase and eutectic. These features are able to cause crack
initiation and assist propagation with similar mechanisms discussed above. The complex behavior arising from combinations of the above parameters have been reviewed. \[80,86,87\]

It is well known that in austenitic stainless steels, the presence of large amounts of ferrite, 15-35\%, can markedly impair hot workability. This is because cracking occurs at the ferrite-austenite interface, but the mechanisms by which this cracking occurs and indeed by which ferrite lowers the hot ductility are not well understood. More work is required on the hot deformation and fracture of duplex structures. It should also be appreciated that the cumulative effects of many impurity elements, some of which can be strong ferrite formers, may introduce critical amounts of ferrite into the structure. At highly austenite forming compositions, austenite dendrites may solidify and undergo no further change after solidification. Whilst an absence of ferrite is ideally required for good hot workability, as has already been mentioned a little ferrite is useful to dissolve the impurities which lead to hot intergranular cracking, and the additional interfacial area created can lower concentrations of impurities segregated to boundaries. However, careful control over the solidification process, which can vary appreciably within a given specification range, is necessary for good hot workability.

In mode C, $\gamma$ solidifies first with the impurities and $\delta$ phase appearing interdendritically along the irregular $\gamma$ grain boundaries. When the ratio of Cr'/Ni' is above 1.5, $\delta$ solidifies first; however, the initially interdendritic $\gamma$ grows to consume most of the $\delta$. The remaining $\delta$ from the high purity dendrite cores often lies in rows giving rise to smooth $\gamma$ boundaries. Myllykoski and Suutala \[88\] have shown that the ductility rises with increase in the above ratio in the $\gamma$ mode and then declines in the $\delta$ mode. The former's ductility is very sensitive to intergranular impurities especially those like Pb.\[88\] The
ductility rises with δ content because it enhances both irregularity, reducing sliding, and recrystallization nucleation; however as δ approaches a continuous layer it promotes severe interfacial cracking. The δ mode is only half as sensitive to impurities and ductility declines with increasing δ content (ratio increase ) which is rendered more damaging by the smooth boundaries. \[^{[88]}\] For both modes, $\varepsilon_f$ and solidus temperature are lowered by S. \[^{[80,88,89]}\] Increased Mo can be beneficial if it raises the ratio causing a switch to δ mode, whereas increased Ni can be bad if it promotes γ mode (a rise from 8 to 12% Ni broadens and deepens the ductility minimum. \[^{[80,89]}\]) In δ mode, rising N can be good for reducing the δ ferrite, but in γ- mode (310 grade ) is always damaging. Inclusions with low plasticity relative to the matrix give rise to fissures by cracking or by boundary decohesion, as verified in Fe-25Ni in which the ductility fell with rise in chromite inclusions. \[^{[36]}\] The as-cast ductility of 304 is raised by reduction of non metallic particles through deoxidizing with Al and treating with B, or better with Ce. The columnar region is superior to the interior equiaxed one, but the reverse could be induced by impurities, or δ phase distribution. \[^{[80,85]}\]

2.5 Quality Pursuit in Billet Casting of Stainless Steel

2.5.1 Shell Growth in the Casting of Stainless Steel

Austenitic grades, as we know, tend to non-uniform shell growth with accentuated surface roughness - mainly in the form of transverse depression. \[^{[91,92]}\] Investigation of shell growth in the mould by tracer addition, and the examination of the breakout shells \[^{[93]}\] show that every depression on the surface coincides with a reduction in shell growth. This is a consequence of the locally reduced heat flow at depression sites (air gap), which retards
shell growth. Furthermore, the surface temperature in depressions remains high, causing a
dangerous ‘hot spot’ with low mechanical resistance. Concurrent coarsening of the
microstructure may also reduce high temperature ductility.\textsuperscript{[94]} \textbf{The general hazard of crack} formation and breakout occurrence at hot spots has been analyzed for carbon steels in the
context of cooling conditions near slab corners.\textsuperscript{[95]} With austenitic stainless steels, an
additional detrimental effect can be observed: local enrichment in ferrite coinciding with
depression sites.\textsuperscript{[96]} Excessive ferrite lowers the intrinsic hot workability of austenitic
stainless steels.

In order to assure high strand surface quality in austenitic stainless steels, it is
important therefore to maintain even shell growth with minimum depression occurrence.
The generally accepted mechanism of depression formation is the local ‘overcooling’ of
the initial strand shell leading to intermittent contraction and rebending. This behavior
strongly depends on steel composition.\textsuperscript{[97,98]} \textbf{The occurrence of transverse depressions is}
not so typical for grade 316. Longitudinal depressions and cracks are sometimes observed
for this grade as the result of major operational irregularities (mould powder sticking,
broken pouring tube). Transverse depression are more a problem with the grade 304
steel, while with the grade 321 steel, the tendency is less pronounced, but a wide scatter
range is observed – mainly attributed to occasional malfunctions of the mould powder on
picking up titanium nitrides and oxides. Thus, mould performance also significantly affects
depression occurrence.

For the grade 430 steel – the main representative of ferritic steels – depressions are
hardly ever observed. Resulphurized austenitic grades 304 and 316 show very little
tendency to depression and can be explained by a drastically reduced shell strength as a result of low solidus temperature.\(^{[99]}\)

The distinction of shell growth behavior as a function of steel composition applies to high-temperature mechanical properties as a consequence of microsegregation for strand tramp element contents, i.e. about 0.01%S and 0.03%P, respectively. In grades with much higher sulphur content (or low Mn/S ratio), for instance, the non-uniformity of shell growth is reduced, as also observed for carbon steels\(^{[97]}\). Austenitic stainless steels of 304 type composition (with an equivalent Ni/Cr ratio of 0.55–0.60) are very sensitive to such behavior because of the \(\delta \rightarrow \gamma\) transformation just at the end of solidification. This phenomena can be explained by the mechanism proposed by Grill and Brimacombe for low carbon steel.\(^{[22]}\) As countermeasures, local heat flux and shell growth near the meniscus should be kept low by using mould powder as lubricant and by the optimization of other casting conditions (casting speed, liquid slag thickness and viscosity, mould level stability, liquid steel convection) in order to assure high slag film stability.

### 2.5.2 Crack Formation

A crack will form when a tensile force generates a strain which exceeds the strain-to-fracture of the steel.\(^{[37]}\) The crack will form perpendicular to the tensile force. Virtually all surface cracks form in the mould.\(^{[37,38]}\)

Thermal stresses in the mould are generally tensile at the surface and compressive at the solidification front. Thermal stresses are complex however, and depend on heat transfer and steel grade. Reheating at the mould exit may cause tensile stress at solidification front, forming midway cracks. If the shell bulges due to ferrostatic pressure, a transverse tensile force is created.
In billet casting, virtually all cracks form in the zone of low ductility,\textsuperscript{[37]} within about 70°C of the solidus temperature. This mechanism, also called "hot tearing", is caused by solute rich liquid between dendrites. The effect is worsened with increasing concentrations of sulfur and phosphorus. Since these cracks form near the solidus temperature, the depth of a subsurface crack is a reasonable estimate of the shell thickness when the crack formed.\textsuperscript{[37]}

Lindenberg\textsuperscript{[39]} have correlated the internal cracking susceptibility of ferritic stainless-steel slabs with their high temperature mechanical properties. He suggested that the frequency of internal cracks (due to bulging and cracks and depressions) increase when the temperature difference for zero ductility and zero strength ($\Delta T_{zds}$) increase. With martensitic grades, susceptibility to internal cracking increases with increasing carbon content. This is explained by the fact that the $\Delta T_{zds}$ increase significantly beyond the peritectic point. In austenitic stainless steels, susceptibility to hot cracking is related to the cast structure and the amount of delta ferrite present.

The main surface cracks observed in continuously cast steels include longitudinal, transverse, and star cracks. Stainless steel slabs of AISI 310 grade are more susceptible to transverse surface cracking than AISI 304 grade because of lower ductility of the former.\textsuperscript{[9]} Slabs of AISI 631, a martensitic grade, are also prone to transverse cracks.\textsuperscript{[9]} In both cases, the poor ductility is attributed to segregation of sulfur at the grain boundaries.\textsuperscript{[9]} A reduction in the sulfur level minimizes the generation of surface cracks. In addition, high aluminum content martensitic stainless steel aggravates its lower-ductility problem because of the precipitation of AlN at the grain boundaries.
Inadequate mould taper (i.e. shallow single-taper mould) gives rise to excessive mould-shell gaps in the lower portion of the mould.\cite{30} Ferrostatic pressure may cause the shell to bulge slightly in these cases, until the shell mid-face impinges on the mould. The hinging action about the corner creates a tensile strain at the solidification front forming the off-corner internal crack.\cite{40} This effect is more likely to occur if shell thickness has been reduced in the off-corner region due to deep oscillation marks which have reduced heat transfer. Cracks which are within 8 mm of the surface likely formed in the mould; cracks deeper than 8 mm may have been formed due to bulging below the mould.\cite{37}
Chapter 3  SCOPE AND OBJECTIVES OF THE PRESENT WORK

As has been discussed in the previous chapter the key to increase casting speed lies in the mould. It is clear that in the practice of high speed billet casting, the role of heat transfer of the mould can never be overstated. Furthermore the interaction between mould and strand under different operating parameters and its effect on the billet quality also should be investigated carefully.

It is also obvious that, owing to different shrinkage characteristics of different grades of steel, there can not be a universal mould taper through which all grades of steel can be successfully cast, especially in the casting of stainless steel, where little work has been conducted.

With a view to understand the variables that affect mould heat transfer and taper design, plant trials were conducted in which an operating mould was instrumented with thermocouples and load cells. It was expected the analysis of the data collected from trials would help to fulfill the following objectives:

1. To study the heat transfer for different steel grades and operating conditions during high speed casting of stainless steel billets.
2. To simulate the solidification and shell growth of the strand and its shrinkage as a function of its position in the mould.

3. To evaluate the interaction between mould and strand during of the casting of stainless steel.

4. To correlate the surface quality of the billets with the operating parameters in the casting of stainless steel.

To realize those objectives, a mathematical model INVERSE developed by the Billet Casting Research Group in UBC is employed to calculate the heat flux profiles. The existed shrinkage model is incorporated with the thermal-mechanical parameters of the stainless steel for shell growth calculation and shrinkage calculation. Mould distortion is also calculated using a mathematical model so as to evaluate the interaction between the strand and the mould. Surface quality of the samples collected in the plant trial were evaluated in lab. Factors influencing the surface quality in the casting of stainless steel were analyzed.
Chapter 4 INDUSTRIAL PLANT TRIAL

An instrumented mould trial was conducted at Atlas Steel, in Welland Ontario March 10-11, 1998. A retrofitted mould with 42 thermocouples was installed and mould temperature data were gathered for a total of ten heats. Thermal signals were obtained for six of ten heats. Two of the heats were high carbon (0.7-0.8 pct. C) grades, three were martensitic stainless steels (0.11 pct. C and 12.7 pct. Cr) and one heat of austenitic stainless steel (0.076 pct. C, 16.7 pct. Cr and 7.52 pct. Ni). LVDT's and strain gauges were also installed with the objective of obtaining mechanical signals.

4.1 Caster Details and Nominal Operating Practice

The operating parameters of the continuous casting machines at Atlas Steel and details of the mould system are provided in Table 4.1. The trial was conducted for a section size of 114 mm square cast at a nominal speed of 2.3 m/min. The oscillator stroke length was typically 6 mm while the frequency varied with the casting speed was set at 3.7 Hz for a speed of 2.3 m/min. The mould was Cu-Cr-Zr with a wall thickness of 11 mm and had a upper taper of 1.2%/m over the first 350 mm and a second taper of 0.5%/m over the lower half of the tube. The total mould length was 780 mm. The mould cooling water velocity was estimated to be 9.3 m/s.
Mould flux lubrication was employed during casting with a submerged refractory funnel.

4.2 Pre-trial Preparation

4.2.1 Mould and Mould Jacket

A mould tube was fitted with 42 intrinsic/copper thermocouple; 13 on each of the faces as shown in Figure 4.1. The thermocouples were embedded in the mould wall to a depth of approximately 4–6 mm from the hot face. In addition four commercially available extrinsic copper-constantan thermocouples were installed on the centerline of each face at the exit of the water channels to measure the outlet temperature of the cooling water on each of the four faces. Thermocouples were also placed in the housing to measure the bulk inlet water temperature.

4.2.2 Data Acquisition and Computer

The data was collected using Labtech Notebook (version 7.1.1), a commercially available software. For data acquisition, a Metrabyte Universal Expansion Interface, Model EXP-16 was employed together with Metrabyte multiplexer DAS-8 board designs for an IBM-PC. It has a resolution of 0.012 mV of approximately 0.5 °C in the case of thermocouples. Data was collected in 60 channels at a rate of 10 Hz for the entire heat for ten heats. A second computer was set up to collect the mechanical signals at data acquisition rates of 100 Hz for 30s intervals during each heat.

4.3 Measurement of Mould Temperature

With the technique developed by Brimacombe, Samarasekera and co-workers, the UBC continuous casting group has for the past two decades successfully instrumented billet moulds with thermocouples to measure wall temperature. Single-wire, type T
intrinsic copper-constantan thermocouples, recommended to measure temperature from -200°C to 350°C, were employed to measure the mould wall temperature. The thermocouples were prepared by forming a bead on the constantan wire with TIG welder. The bead was filed flat to produce an end of approximately 0.3–0.4 mm thick, then heat-shrinking tubing (1.6 mm in diameter) was applied to the wire to insulate.

To install the thermocouples into the mould, threaded holes were prepared through the water baffle and half-way into the copper mould wall. The bead side of the wires were inserted through the water baffle into the holes drilled in the mould wall and held in place by copper plugs screwed into the threaded holes. Silicone sealant was employed to prevent any water leak through the baffle. The constantan wires were then attached to insulated copper wires in the cooling water plenum. Groups of wires were bunched together and connected to a data acquisition system through a pipe fitting in the side of mould jacket. Thermocouples configuration was then tested for electrical continuity and assembly was pressure tested with water.

4.4 Measurement of Mould Displacement

Mould displacement was measured using LVDT’s at a rate of 100 Hz for 30 seconds every 600 to 1000 seconds during a heat or when a change in practice was imposed.

To record mould displacement, linear variable differential transformers (also called linear displacement transducers) were used. The LVDT is an electromechanical devices that produces an electrical output proportional to the displacement of a separate movable core. The LVDT’s are attached to the housing in such a way that they register the movement of the mould. Since it is not necessary to know the absolute displacement of the mould, the LVDT’s do not have to be calibrated on the oscillator. It is only necessary to
ensure that the full range movement of the LVDT, when in use, is within the linear output range of the circuitry.

4.5 Other Miscellaneous Measurements

4.5.1 Casting Speed

The signal from the withdraw roll tachometer which is proportional to the casting speed, was used to record the casting speed. The signal from the tachometer is typically 0–40 or 0–200 volts D.C and was stepped down (0–20 millivolts) and filtered for any electrical noise (8–45Hz) before being sent to the data acquisition system. The calibration of the signal was done on site by measuring a steel billet into the mould and measuring the output voltage relative to the casting speed gauge and adjusting a variable resistor so as to produce a convenient ratio.

4.5.2 Metal Level

The metal level signal was obtained from plant as 4–20 mA signal and converted by the UBC system into a 0–20 mV output.

4.5.3 Mould Water Velocity

During the trial the mould water flow rate remained relatively constant at 1600l/min which translates into a mould water velocity of 9.3 m/s.

4.5.4 Test Data

Strand No.3 was the test strand and data was collected for a total of ten heats.

4.6 Powder Used in The Trial

Mould powder is used in all the ten heats during plant trial. The composition and properties of the mould powders are presented in Table 4.3.
4.7 Chemical Compositions

The chemical compositions of the heats monitored at the plant trial are tabulated in Table 4.4.

4.8 Laboratory Work

Billets were collected for each heat monitored in the plant trial. The billet samples collected during the plant trials were subjected to a rigorous inspection procedure. For this study the samples were inspected for:

1. Surface quality:
   - Oscillation marks
   - Depressions transverse & longitudinal
   - Bleeds, laps and sticking marks

2. Dimensional quality
   - Rhomboidity or off-squareness (difference between two diagonals)

Oscillation marks were of particular interest to this study. Effort has been made to correlate the depth of oscillation mark with steel grade and casting speed. The depth of oscillation marks were measured using a profilometer table featuring three LVDT’s.
Table 4.1 Operating parameters for the caster in the Plant Trial

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Machine Type</td>
<td>Curved Mould</td>
</tr>
<tr>
<td>Mould Length</td>
<td>780 mm</td>
</tr>
<tr>
<td>Billet Size</td>
<td>114 mm sq.</td>
</tr>
<tr>
<td>Nominal Casting Speed</td>
<td>2.3 m/min</td>
</tr>
<tr>
<td>Reoxidization Protection</td>
<td>Yes</td>
</tr>
<tr>
<td>Nominal Meniscus Level</td>
<td>135 mm</td>
</tr>
<tr>
<td>Oscillation Type</td>
<td>Sinsoidal</td>
</tr>
<tr>
<td>Stroke Length</td>
<td>6 mm</td>
</tr>
<tr>
<td>Oscillation Frequency</td>
<td>3.7 Hz</td>
</tr>
<tr>
<td>Negative Strip Time</td>
<td>0.085—0.101 s</td>
</tr>
<tr>
<td>Mould Lead</td>
<td>1.76 mm</td>
</tr>
</tbody>
</table>
Table 4.2 Details of casting mould in the plant trial

<table>
<thead>
<tr>
<th>Material</th>
<th>Cu - Cr -Zr</th>
</tr>
</thead>
<tbody>
<tr>
<td>Thickness</td>
<td>11 mm</td>
</tr>
<tr>
<td>Corner Radius</td>
<td>3.175 mm</td>
</tr>
<tr>
<td>Construction</td>
<td>Tube</td>
</tr>
<tr>
<td>Taper</td>
<td>0-350 mm 1.2%/m</td>
</tr>
<tr>
<td></td>
<td>350-780 mm 0.5%/m</td>
</tr>
<tr>
<td>Mould Length</td>
<td>780 mm</td>
</tr>
<tr>
<td>Baffle Gap</td>
<td>5.08 mm</td>
</tr>
<tr>
<td><strong>Cooling Water:</strong></td>
<td></td>
</tr>
<tr>
<td>Flow</td>
<td>1600 l/min</td>
</tr>
<tr>
<td>Velocity</td>
<td>9.31 m/s</td>
</tr>
<tr>
<td>Constraint Type</td>
<td>Four Side</td>
</tr>
</tbody>
</table>
Table 4.3 Mould powder composition and properties in the Plant Trial

<table>
<thead>
<tr>
<th>Composition</th>
<th>CT3</th>
<th>BL29</th>
<th>STH</th>
</tr>
</thead>
<tbody>
<tr>
<td>(Wt %)</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>CaO</td>
<td>28.9</td>
<td>24.8</td>
<td>20.0</td>
</tr>
<tr>
<td>SiO₂</td>
<td>35.9</td>
<td>24.8</td>
<td>33.0</td>
</tr>
<tr>
<td>CaO/SiO₂</td>
<td>0.86</td>
<td>1.00</td>
<td>0.65</td>
</tr>
<tr>
<td>MgO</td>
<td>-</td>
<td>8.0</td>
<td>0.65</td>
</tr>
<tr>
<td>F</td>
<td>8.0</td>
<td>8.2</td>
<td>5.9</td>
</tr>
<tr>
<td>ZrO₂</td>
<td>-</td>
<td>2.7</td>
<td>-</td>
</tr>
<tr>
<td>Al₂O₃</td>
<td>7.9</td>
<td>3.6</td>
<td>4.9</td>
</tr>
<tr>
<td>Na₂O</td>
<td>6.6</td>
<td>10.5</td>
<td>9.3</td>
</tr>
<tr>
<td>Fe₂O₃</td>
<td>4.5</td>
<td>1.2</td>
<td>1.0</td>
</tr>
<tr>
<td>C total</td>
<td>2.7</td>
<td>3.0</td>
<td>16.7</td>
</tr>
<tr>
<td>Tₙ (°C)</td>
<td>983</td>
<td>925</td>
<td>960</td>
</tr>
<tr>
<td>Viscosity (poise)</td>
<td>4.4</td>
<td>1.3</td>
<td>4.8</td>
</tr>
</tbody>
</table>

Tₙ — melting temperature
### Table 4.4 Summary of heat chemistry in the Plant Trial

<table>
<thead>
<tr>
<th>Element [Wt %]</th>
<th>Heat Number</th>
<th>Heat Number</th>
<th>Heat Number</th>
<th>Heat Number</th>
<th>Heat Number</th>
<th>Heat Number</th>
<th>Heat Number</th>
<th>Heat Number</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>6327H (A)</td>
<td>6328H (A)</td>
<td>8110J (A)</td>
<td>8111J (A)</td>
<td>8113J (M)</td>
<td>6331H (A)</td>
<td>8115J (M)</td>
<td>6334H (C)</td>
</tr>
<tr>
<td>C %</td>
<td>0.011</td>
<td>0.054</td>
<td>0.008</td>
<td>0.016</td>
<td>0.11</td>
<td>0.076</td>
<td>0.13</td>
<td>0.79</td>
</tr>
<tr>
<td>Mn %</td>
<td>1.72</td>
<td>1.69</td>
<td>1.41</td>
<td>0.92</td>
<td>0.36</td>
<td>0.82</td>
<td>0.46</td>
<td>0.25</td>
</tr>
<tr>
<td>P %</td>
<td>0.029</td>
<td>0.029</td>
<td>0.025</td>
<td>0.026</td>
<td>0.019</td>
<td>0.025</td>
<td>0.018</td>
<td>0.011</td>
</tr>
<tr>
<td>S %</td>
<td>0.025</td>
<td>0.335</td>
<td>0.003</td>
<td>0.023</td>
<td>0.307</td>
<td>0.001</td>
<td>0.357</td>
<td>0.034</td>
</tr>
<tr>
<td>Si %</td>
<td>0.44</td>
<td>0.72</td>
<td>0.22</td>
<td>0.62</td>
<td>0.48</td>
<td>0.21</td>
<td>0.45</td>
<td>0.18</td>
</tr>
<tr>
<td>Cr %</td>
<td>16.37</td>
<td>17.23</td>
<td>18.04</td>
<td>17.58</td>
<td>12.43</td>
<td>16.67</td>
<td>12.72</td>
<td>0.62</td>
</tr>
<tr>
<td>W %</td>
<td>0.048</td>
<td>0.025</td>
<td>0.047</td>
<td>0.053</td>
<td>0.041</td>
<td>0.042</td>
<td>0.046</td>
<td>0.012</td>
</tr>
<tr>
<td>V %</td>
<td>0.073</td>
<td>0.073</td>
<td>0.079</td>
<td>0.089</td>
<td>0.054</td>
<td>0.074</td>
<td>0.090</td>
<td>0.045</td>
</tr>
<tr>
<td>Mo %</td>
<td>2.01</td>
<td>0.28</td>
<td>0.43</td>
<td>0.43</td>
<td>0.095</td>
<td>0.465</td>
<td>0.101</td>
<td>0.052</td>
</tr>
<tr>
<td>Ni %</td>
<td>10.07</td>
<td>9.11</td>
<td>9.32</td>
<td>11.44</td>
<td>0.33</td>
<td>7.52</td>
<td>0.29</td>
<td>0.22</td>
</tr>
<tr>
<td>Cu %</td>
<td>0.37</td>
<td>0.24</td>
<td>0.30</td>
<td>0.4</td>
<td>0.10</td>
<td>0.37</td>
<td>0.10</td>
<td>0.10</td>
</tr>
<tr>
<td>Al %</td>
<td>0.003</td>
<td>0.003</td>
<td>0.004</td>
<td>0.004</td>
<td>0.001</td>
<td>0.008</td>
<td>0.001</td>
<td>0.001</td>
</tr>
<tr>
<td>Sn %</td>
<td>0.013</td>
<td>0.003</td>
<td>0.009</td>
<td>0.007</td>
<td>0.007</td>
<td>0.007</td>
<td>0.008</td>
<td>0.006</td>
</tr>
<tr>
<td>Pb %</td>
<td>0.001</td>
<td>0.001</td>
<td>0.002</td>
<td>0.001</td>
<td>0.001</td>
<td>0.001</td>
<td>0.001</td>
<td>0.001</td>
</tr>
<tr>
<td>Co %</td>
<td>0.13</td>
<td>0.024</td>
<td>0.095</td>
<td>-</td>
<td>0.025</td>
<td>0.10</td>
<td>0.023</td>
<td>-</td>
</tr>
<tr>
<td>Ti %</td>
<td>0.006</td>
<td>0.006</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0.011</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Nb %</td>
<td>0.010</td>
<td>0.010</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0.002</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>N₂ %</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0.026</td>
<td>-</td>
<td>0.034</td>
</tr>
<tr>
<td>Mn/S</td>
<td>68.8</td>
<td>5.045</td>
<td>470</td>
<td>40</td>
<td>1.173</td>
<td>820.0</td>
<td>1.289</td>
<td>7.353</td>
</tr>
<tr>
<td>Mn/Si</td>
<td>3.909</td>
<td>2.347</td>
<td>6.409</td>
<td>1.484</td>
<td>0.75</td>
<td>3.905</td>
<td>1.022</td>
<td>1.389</td>
</tr>
<tr>
<td>Ni/Cu</td>
<td>27.22</td>
<td>37.96</td>
<td>2.824</td>
<td>28.6</td>
<td>3.3</td>
<td>20.32</td>
<td>2.9</td>
<td>2.2</td>
</tr>
</tbody>
</table>

A: austenitic stainless steel grade  
M: martensitic stainless steel grade  
C: plain carbon steel grade
Table 4.5 Casting parameters change during Steel Plant Trial

<table>
<thead>
<tr>
<th>Heat Number</th>
<th>Steel Grade</th>
<th>Powder</th>
<th>Operating Parameters Change</th>
<th>Billet Samples</th>
</tr>
</thead>
</table>
| 6327H       | 316L36 (austenitic) | CT3    | casting speed change: 2.3~3.1 m/min  
water flow change: 1300~1700 l/min  
Nozzle depth change                          | #1 2.3m/min  
#2 2.8m/min  
#3 3.1m/min |
| 6328H       | 303F16 (austenitic) | CT3    | casting speed change: 2.3~3.3 m/min                      | #1 2.3m/min  
#2 2.8m/min  
#3 3.3m/min |
| 6331H       | AL375-0 (austenitic) | CT3    | casting speed change: 1.9~2.57 m/min                        | #1 2.3m/min |
| 8110H       | AL387-3 (austenitic) | CT3    | mould powder change from CT3 to BL23                        | #1: CT3  
#2,3,4,5:BL23 |
| 8111H       | AL305-5 (austenitic) | CT3    | casting speed change: 1.9~2.7 m/min  
water flow change: 1300~1750 l/min              | #1 1.9m/min  
#2 2.3m/min  
#3 2.7m/min |
| 8113J       | T416F1 (martensitic) | CT3    | casting speed change: 1.9~2.3 m/min                        | #1 1.9m/min |
| 8115J       | 416F1 (martensitic) | CT3    | constant casting speed: 1.9 m/min  
constant water flow: 1600 l/min                  | #1 1.9m/min |
| 8119J       | 416F2 (martensitic) | CT3    | constant casting speed: 1.9 m/min  
constant water flow: 1600 l/min                  | #1 1.9m/min |
| 8117J       | HD3001 (plain carbon) | STH    | casting speed change: 2.15~2.3 m/min  
nozzle position change                           | #1 2.15m/min  
#2 2.3m/min  
#3 2.3m/min (off-center) |
| 6334H       | HD3001 (plain carbon) | STH    | constant casting speed: 2.15 m/min  
constant water flow: 1600 l/min                  |                |

Note: nominal water flow rate is 1600 l/min.
Figure 4.1 Thermocouple layout (all four faces) for the instrumented mould in the plant trial
Chapter 5 RESULTS OF PLANT TRIAL

In this chapter the raw data, including LVDT's and temperatures of the mould wall collected from plant trials are presented with some analysis.

5.1 Mould Temperature Data

5.1.1 Mould Temperature Response

Mould temperature were recorded at thirteen locations on each face and samples of the mould response at thermocouples located at 105 mm, 180 mm, 340 mm, 700 mm down the top of the mould for the austenitic stainless steel, for all four faces are shown in Figure 5.1 to 5.4. These temperature signals are typical unfiltered response of selected thermocouples during plant trial. As we see from the thermocouples signals, that during the casting process the local temperatures of the mould wall are not constant due to the changes in operating parameters such as casting speed, metal level and mould powder consumption.

5.1.2 Average Temperature and Standard Deviation

Average values and standard deviations of the temperature signals from thermocouples in different positions were calculated and analyzed. Calculating results for all the six heats are listed in Table 5.1-5.4. Samples of the temperature profile and
standard deviation are shown in Figure 5.5-5.8. From those calculations, we have observed the following phenomena:

I. Peak temperature in the mould always occurs about 35 to 50 mm below the meniscus. In the plant trial, the metal level was kept 135 mm below the top of the mould. The maximum temperatures measured always occurred at 170–185 mm below the top of the mould. This can be attributed to two dimensional heat transfer factor in the upper portion of the mould. Maximum standard deviation always occur around meniscus and the values range from 7 to 15°C.

II. At the meniscus, temperature signals from the outside curved wall have the maximum Standard Deviation. Also for all the heats in the plant trial, we find that around meniscus, the outside curved wall always has a larger average temperature than the inside curved wall. The difference ranges from 8–25 °C. It is likely that the non-uniform mould powder behavior is responsible for this. The mould powder was fed manually from the inside curved wall, causing non-uniform slag rims for the inside curved wall and outside curved wall thereby influencing the heat transfer of the mould wall.

III. The temperatures at similar positions in the four mould walls are different from each other because of uneven mould powder behavior and gap size.

IV. In all the heats during the plant trial, we find that the average temperatures from the thermocouples at 280 mm, 400 mm and 520 mm below the top of the inside curved mould wall are significantly lower than those from neighboring thermocouples. A typical example is shown in Figure 5.5. It can be seen from the figure that the largest temperature difference occurs between 520 mm to 640 mm,
which is 34 °C. There is no such phenomena for the other mould walls. These
temperature fluctuation give rise to the wavy heat flux profiles along the mid-face
of the inside mould wall which will be shown in the next section.

5.1.3 Effect of Casting Speed on Mould Temperature

The influence of changing casting speed on the mould temperature is also delineated.
When the casting speed was increased from 1.9 to 2.4 m/min, approximately 10 minutes
into the heat, mould temperature on all four faces experienced a moderate increase. For
example, the average temperatures 150 mm below the top of the mould increase from 193
to 198 °C for the inside curved wall, 156 to 165 °C for the left side straight wall, 150 to
156 °C for the outside curved wall and 195 to 222 °C for the right side straight wall. A
further increase in speed to 2.57 m/min does not result in a noticeable increase in mould
temperature on all four faces. The result is expected because increasing the casting speed
will reduces the shell thickness of the strand in the mould, which in turn reduce the
thermal resistance of the shell and the shrinkage of the strand.

5.2 Billet Quality Evaluation

5.2.1 Surface Quality

All the billets were shown blasted to remove the oxide layer and were evaluated. This
evaluation included cracks and surface defects such as bleeds and laps, oscillation mark
appearance, including irregular spacing, non-parallel, non-linearity, billet surface
roughness, depressions both longitudinal and transverse, inclusions and pinholes. Table 5.1
presents the surface evaluation summary. Generally two main problems were observed.
The first being the appearance of bleeds and laps in 19 of the 36 faces observed and
secondly high surface roughness in 29 of the 36 observed faces. Without exception, all surface quality problems were predominant in austenitic stainless steel.

Different mould powders were tested in heat 8110, which was Al375 austenitic stainless steel with an equivalent Ni'/Cr' ratio of 0.58. Surface quality of the samples was evaluated and the results were listed in Table 5.6.

<table>
<thead>
<tr>
<th>Sample #1</th>
<th>Sample #2</th>
<th>Sample #3</th>
<th>Sample #4</th>
</tr>
</thead>
<tbody>
<tr>
<td>I L O R</td>
<td>I L O R</td>
<td>I L O R</td>
<td>I L O R</td>
</tr>
<tr>
<td>Lap</td>
<td>0 0 0 0</td>
<td>1 0 0 0</td>
<td>0 0 0 0</td>
</tr>
<tr>
<td>Breakout</td>
<td>0 2 0 0</td>
<td>0 0 0 0</td>
<td>0 0 0 0</td>
</tr>
<tr>
<td>L-depression</td>
<td>2 1 1 2</td>
<td>1 2 1 2</td>
<td>0 1 0 0</td>
</tr>
<tr>
<td>T-depression</td>
<td>0 0 0 0</td>
<td>0 3 0 0</td>
<td>0 1 1 0</td>
</tr>
</tbody>
</table>

Note: I - inside; L - left side; O - outside; R - right side

In other samples, only four faces exhibit longitudinal depressions down the center or near the center of billet while only one transverse depression was observed. Figure 5.9 illustrates a longitudinal and transverse depression in billet No.2 from Heat 8110J with mould powder CT3, while Figure 5.10 demonstrate the surface appearance of the samples with mould powder BL23.

We also find serious longitudinal depression in martensitic stainless steel, which are shown in Figure 5.11. The forming of those depression may attribute to the non-optimum mould taper design, which will be discussed in Chapter 8.

5.2.2 Oscillation Marks

The oscillation mark depths for all billet samples were measured using the UBC profilometer. The measurements were made at three locations on each of the straight sides.
of the billet samples - one along the centerline and two along the off-corner positions approximately 14 mm from each corner. The average depths of the oscillation marks and the corresponding standard deviations are presented in Table 5.7.

From Table 5.7, it can be seen that average oscillation mark depths range from 0.32\textendash}0.47 mm for austenitic stainless steel billets and less than 0.30 mm for martensitic stainless steel, even with a higher superheat (68°C for heat 6331). The standard deviations of the oscillation mark depth are usually larger than 0.25 for austenitic stainless steel billets, while they are usually less 0.20 for martensitic stainless steel. The higher standard deviation of the oscillation mark depths for austenitic stainless steel reveals the relative irregularity of the oscillation mark, which may be attributed to the delta to gamma transformation during the initial solidification process similar to peritectic steel. When the casting speed increased from 1.9 to 2.3 and then to 2.7 m/min in heat 8111, average oscillation mark depths changed from 0.37 to 0.36 and then to 0.39 mm. In heats 6328 and 6327, depths of oscillation mark from samples for different casting speed were studied, no significant influence of casting speed on the depth of oscillation mark were observed. Average oscillation mark depths along the centerline of the left-hand surface and right-hand surface were investigated for austenitic stainless steel. Oscillation mark depths in four out of the total six inspected faces has shown the tendency of increase when the casting speed was increased. The increase in oscillation mark depths was accompanied by the increase of standard deviation.

Compared with high carbon steel, both the depths and the standard deviations of the oscillation marks for stainless steel are larger as can be observed in Table 5.7. Figure 12
Chapter 5 RESULTS OF PLANT TRIAL

illustrates the surface of a high carbon steel sample collected in the plant trial. It is obvious that the oscillation mark was uniform and shallow.

5.2.3 Crack Inspection

All the samples in the plant trial were cut and cross section were inspected for internal defects. Unlike plain carbon steel, no crack were observed in the subsurface of the samples with longitudinal & transverse depressions.

5.3 Miscellaneous

5.3.1 Mould Displacement

LVDT’s were employed to monitor mould displacement. The mould displacement signal from the LVDT for heat 6331 at one frequency are shown in Figure 5.14. The operating stroke of the caster ranged from 5.8 ~ 6.4 mm. The stroke was seen to change with the changing of casting speed.

5.3.2 Oscillation Frequency and Casting Speed

Many billet casters operate a control scheme where the oscillation frequency is a function of casting speed. In Atlas Steel plant trial, the oscillation frequency is linked to the casting speed as shown in Figure 5.15. When the casting speed increased from 31mm/min to about 45 mm/min, the oscillation frequency also increased from 3.08 to 4.14Hz.

5.3.3 Mould Velocity and Negative Strip Time

Mould velocity was calculated by taking the time derivative of the mould displacement signal. Negative strip time is illustrated schematically in Figure 5.16 using sensor data. Thus negative strip time varies during casting depending on the speed and oscillation frequency and is seen to decrease with increasing speed as shown in Figure

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5.15. With billet casting and oil lubrication the negative strip time is considered optimum when it lies in the range $0.12 \sim 0.15$ s.$^{[54]}$ The optimum negative strip time for billet casting with powder lubrication is not known. However the values observed in this trial are in the $0.086 \sim 0.1$ s range.
Figure 5.1 Mould thermal response on inside curved wall during casting of an austenitic stainless steel; casting speed was changed from 1.90 to 2.57 m/min.
Figure 5.2  Mould thermal response on left side straight wall during casting of an austenitic stainless steel; casting speed was changed from 1.90 to 2.57 m/min.
Figure 5.3 Mould thermal response on outside curved wall during casting of an austenitic stainless steel; casting speed was changed from 1.90 to 2.57 m/min.
Figure 5.4  Mould thermal response on right side straight wall during casting of an austenitic stainless steel; casting speed was changed from 1.90 to 2.57 m/min.
Figure 5.5 Average temperature profile and standard deviation of measured temperature on inside curved wall for a austenitic stainless steel, casting speed is 1.90 m/min, mould powder CT3
Chapter 5 RESULTS OF PLANT TRIAL

Figure 5.6 Average temperature profile and standard deviation of measured temperature on left side straight wall for an austenitic stainless steel, casting speed is 1.90 m/min, mould powder CT3
Figure 5.7 Average temperature profile and standard deviation of measured temperature on the outside curved wall for an austenitic stainless steel, casting speed is 1.90 m/min, mould powder CT3
Figure 5.8  Average temperature profile and standard deviation of measured temperature on the right side straight wall for an austenitic stainless steel, casting speed is 1.90 m/min, mould powder CT3
Figure 5.9 Photograph of longitudinal and transverse depressions on a billet sample from heat 8110j, austenitic stainless steel, mould powder CT3
Figure 5.10  Photograph of surface defects on billet samples from heat 8110J (a) transverse depression, (b) breakout; mould powder BL23
Figure 5.11  Photograph of a longitudinal depression on a billet sample from heat 8113j, martensitic stainless steel, mould powder CT3
Figure 5.12  Photograph of the surface of a billet sample from heat 6334J, high carbon steel, 0.79% C, mould powder STH
Figure 5.13 Mould displacement signal for heat 6331 at a frequency of 3.13 Hz
Figure 5.14 Relationship between oscillation frequency and casting speed during the plant trial

\[ f = 1.5583v + 0.1296 \]
Figure 5.15  Negative strip time value for different casting speed during the plant trial
Figure 5.16  Mould velocity change during oscillation for heat 6331 during Atlas Plant Trial
Table 5.1 Average temperature and standard deviation, inside curved wall

<table>
<thead>
<tr>
<th>Position (mm)</th>
<th>Heat 6331 (Austenitic)</th>
<th>H8113 (Martensitic)</th>
<th>H8115 (Martensitic)</th>
<th>H8119 (Martensitic)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
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<td>CSp=2.40</td>
<td>CSp=2.57</td>
<td>CSp=1.90</td>
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<tr>
<td></td>
<td>T(°C)</td>
<td>Std</td>
<td>T(°C)</td>
<td>Std</td>
</tr>
<tr>
<td>105</td>
<td>71.32</td>
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</table>

Std: standard deviation; CSp: casting speed (m/min)

Table 5.2 Average temperature and standard deviation, left side straight wall

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<tr>
<th>Position (mm)</th>
<th>Heat 6331 (Austenitic)</th>
<th>H8113 (Martensitic)</th>
<th>H8115 (Martensitic)</th>
<th>H8119 (Martensitic)</th>
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<tbody>
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<td>T(°C)</td>
<td>Std</td>
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Std: standard deviation; CSp: casting speed (m/min)
Table 5.3 Average temperature and standard deviation, outside curved wall

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<th>Position (mm)</th>
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<th>H8113 (Martensitic)</th>
<th>H8115 (Martensitic)</th>
<th>H8119 (Martensitic)</th>
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<td>T(°C)</td>
<td>Std</td>
<td>T(°C)</td>
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Std: standard deviation; CSp: casting speed (m/min)

Table 5.4 Average temperature and standard deviation, right side straight wall

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<th>H8113 (Martensitic)</th>
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<th>H8119 (Martensitic)</th>
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<td>CSp=2.57</td>
<td>CSp=1.9</td>
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<td>T(°C)</td>
<td>Std</td>
<td>T(°C)</td>
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Std: standard deviation; CSp: casting speed (m/min)
Table 5.5  Surface quality evaluation of the samples in the Plant Trial, mould powder CT3

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<th>Grade</th>
<th>No. of Billet</th>
<th>Bleeds or laps</th>
<th>Irregular spacing</th>
<th>Non-Parallel</th>
<th>Non-Linear</th>
<th>Roughness</th>
<th>Depression</th>
<th>Inclusion</th>
<th>Pinholes</th>
<th>Off-Corner Depression</th>
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<td>2</td>
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M: martensitic stainless steel;  A: austenitic stainless steel;  C: plain carbon steel
Chapter 5  RESULTS OF PLANT TRIAL

Table 5.7  Oscillation mark depths of the samples in the Plant Trial, mould powder CT3

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<th>Grade</th>
<th>Csp.:m/ min</th>
<th>Billet No.</th>
<th>Side</th>
<th>ICW Avg Depth</th>
<th>Std</th>
<th>CL Avg Depth</th>
<th>Std</th>
<th>OCW Avg Depth</th>
<th>Std</th>
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<td>0.346</td>
<td>0.20</td>
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</table>

*Not Measured
Chapter 6  MODELLING OF THERMAL RESPONSE OF MOULD AND STRAND SOLIDIFICATION

A pre-existing mathematical model to inverse calculate the heat flux along the mid-face of the mould based on the measured temperature data was introduced in this chapter to investigate the heat flux profiles for different steel grades. Also a pre-existing shrinkage model was modified to investigate the solidification, shell growth, shrinkage and the interaction between the mould wall and the strand in continuous casting of stainless steel billets.

6.1 Mathematical Model of Heat Transfer in the Mould

6.1.1 Inverse Heat Conduction Problem

Mould heat flux can be determined from measured mould wall temperatures. A mathematical model called INVERSE developed by UBC is employed to calculate the heat flux profile along the mid-face down the top of the mould.

A mathematically more rigorous approach consists of solving the Inverse Heat Conduction Problem (IHCP) by analytical or numerical methods. Among the variety of proposed methods for solving the IHCP, the regularization method is a procedure that minimizes a modified least squares function of the differences between the temperatures
measured by thermocouples and temperatures predicted by a direct model of the problem.

The regularization procedure modifies the least squares function by adding factors to
reduce fluctuations in the unknown function to be estimated. These fluctuations or
instabilities are not of physical origin but are inherent in the solution of ill-posed problems
unless they are specially treated. Detailed description of this model is presented by
Pinheiro in his Ph.D thesis.\textsuperscript{56}

\includegraphics{figure6.1.png}

\textbf{Figure 6.1} Schematic diagram of the mid-face longitudinal section of the mould
wall
Chapter 6 MODELLING OF THERMAL RESPONSE OF MOULD AND STRAND SOLIDIFICATION

A schematic diagram of the mid-face longitudinal section of the mould wall is shown in Figure 6.1.

The following assumptions have been made in the heat flow model:

I. The mould is at steady state, which is valid if no nucleate boiling occurs in the cooling water gap.

II. Temperature variations due to mould oscillation and metal level fluctuations are ignored. This is permissible as an average value of the temperature distribution in the mould is used.

III. Transverse heat flow in a direction perpendicular to the plane of interest are negligibly small.

IV. Temperature dependence of the thermal diffusivity has not been considered as its effect on mould temperature is negligible.

V. The top and bottom surface of the mould are treated as adiabatic surface as they have temperatures that are close to the ambient temperature.

VI. The cooling water channel extends to the top and bottom ends of the mould and the cooling water is in plug flow.

The governing equation for heat conduction in the mould wall in two dimensions is

\[
\frac{\partial}{\partial x} \left( K_m \frac{\partial T}{\partial x} \right) + \frac{\partial}{\partial z} \left( K_m \frac{\partial T}{\partial z} \right) = \rho_m C_{pm} \frac{\partial T}{\partial t} \quad (6.1)
\]

while a sectional heat balance for the cooling water yields

\[
\rho_w V_w d_w C_{pw} \frac{\partial T}{\partial z} - h_w(z,t)(T(0,z,t) - T_w(z,t)) = 0 \quad (6.2)
\]

The boundary conditions and details of the model have been discussed in the original work.\(^{18}\) The mould continuum was discretized and the finite difference equations for the
system of nodes obtained by transforming the heat flow equations and relevant boundary conditions for each node into a finite difference form was solved by the alternating direction implicit method.

6.1.2 Characterization of Heat Transfer in the water Channel

In a procedure described by Rohsenow\textsuperscript{[50]} and discussed in detail by Samarasekera and Brimacombe\textsuperscript{[20]} forced convection boiling curve for water was constructed (Figure 6.2). There are three distinct regions of the forced convection boiling regions of the forced convection boiling curve: Forced Convection (FC), the Transitions Region (TR) and the

Figure 6.2 Forced convection - heat transfer curve\textsuperscript{[20]}

6.1.2 Characterization of Heat Transfer in the water Channel

In a procedure described by Rohsenow\textsuperscript{[50]} and discussed in detail by Samarasekera and Brimacombe\textsuperscript{[20]} forced convection boiling curve for water was constructed (Figure 6.2). There are three distinct regions of the forced convection boiling regions of the forced convection boiling curve: Forced Convection (FC), the Transitions Region (TR) and the
Fully Developed Boiling (FDB). In the forced convection regime of the above mentioned curve the heat flux can be calculated by the following expression:

\[ Q_{FC} = h_{fc} (T_w - T_{sat}) \]  

(6.3)

where

\[ \frac{h_{fc} D_H}{K_f} = 0.023 \left( \frac{\rho_f V_f D_H}{\mu_f} \right)^{0.3} \left( \frac{C_p \mu_f}{K_f} \right)^{0.4} \]  

(6.4)

and all values are evaluated at the average temperature of water.

The heat flux in the fully developed nucleate boiling region is given by

\[ Q_{FD} = \left( \frac{T_w - T_{sat}}{H_{fg}} \right) x 1 \times K \times 1 \times 1 \times 1 \times 1 \times 1 \]  

(6.5)

The heat flux in the transition region is given by:

\[ Q_{TR} = Q_{FC} \left( 1 + \frac{Q_{FD}}{Q_{FC}} \left( 1 - \frac{Q_{FD}}{Q_{FC}} \right)^{2} \right)^{0.5} \]  

(6.6)

The heat flux at the inception of boiling is given by the expression

\[ Q_{FN} = 5.281 \times 10^{-3} P^{1.156} (1.8(T_w - T_{sat}))^{0.05} \]  

(6.7)

With a decrease in the number of active nuclei for bubble formation due to degassing of a surface over a long period of time, larger superheats than predicted by equation 6.7 are required to initiate and sustain boiling. The boiling of the water in the water channel can thus cause a hysteresis in the boiling curve leading to thermal cycling - a point explained in detail in the original work. A transient heat flow model was developed to simulate the time dependence effects of boiling on the thermal field in the mould.

### 6.2 Mathematical Model of Billet Contraction
6.2.1 Mathematical Model

A mathematical model to describe the heat flow in a continuous cast strand and to compute the shrinkage of the billet as a function of its axial position in the mould was developed by the Billet Continuous Casting Group in UBC. The model is based on the equation for two dimensional, unsteady-state heat conduction in one quarter of a transverse slice of the strand, as follows:

\[
\frac{\partial}{\partial x} \left( K_m \frac{\partial T}{\partial x} \right) + \frac{\partial}{\partial y} \left( K_m \frac{\partial T}{\partial y} \right) = \rho_v C_{pm} \frac{\partial T}{\partial t}
\]  

(6.8)

The initial and surface boundary conditions, expressed mathematically, are as follows:

\[
t = 0, \quad 0 < x < \frac{X}{2}, \quad 0 < y < \frac{Y}{2}, \quad T(x,y) = T_p
\]  

(6.9)

\[
t > 0, \quad x = 0, \quad 0 \leq y \leq \frac{Y}{2}, \quad -k_s \frac{\partial T}{\partial x} = q_0
\]  

(6.10)

\[
t > 0, \quad y = 0, \quad 0 \leq x \leq \frac{X}{2}, \quad -k_s \frac{\partial T}{\partial y} = q_0
\]  

(6.11)

Symmetrical heat flow about the center planes is assumed and formulated as follows:

\[
t \geq 0, \quad x = \frac{X}{2}, \quad 0 \leq y \leq \frac{Y}{2}, \quad -k_s \frac{\partial T}{\partial x} = 0
\]  

(6.12)

\[
t \geq 0, \quad y = \frac{Y}{2}, \quad 0 \leq x \leq \frac{X}{2}, \quad -k_s \frac{\partial T}{\partial y} = 0
\]  

(6.13)

Equation 6.8 was solved, subjected to the above initial and boundary conditions, using an alternating direction implicit finite-difference method as developed by Peaceman and Rachford. Convective heat transfer in the pool was neglected, an assumption justified by Mizikar and Szekely et al., and the latent heat of solidification was taken into account by
considering equilibrium freezing. The specific heat of the steel was increased in the temperature range between liquidus and solidus temperature of the steel to account for the release of latent heat.

6.2.2 Shrinkage Calculation

The following assumptions were made in the calculation of billet contraction.\[57\]

I. The effect of ferro-static pressure is neglected;

II. The mechanical behavior of the solidified shell is neglected. Neither the strain imposed by the stress field nor creep of the solidified shell is included in the model.

The initial dimensions of the steel billet were taken as those of the distorted copper mould at the meniscus. The differential coefficient of linear thermal expansion of steel was calculated.

The calculation of billet contraction was carried out by the following procedure:

I. A transverse slice of unit thickness of the billet was allowed to move down the mould in discrete time step. At each time step, the temperature distribution in the slice was calculated.

II. The calculation of billet contraction was initiated at the instant the first row of nodes solidified. Prior to the initiation of solidification the billet dimensions were taken to be the dimensions of the distorted mould.

III. For the shrinkage calculation, at a given time step, the temperature change of each node relative to the previous time step was determined.

IV. The length of each node in the row of solidified nodes, due to the temperature change, was calculated relative to the previous time step.
V. The change in the length of the first solidified row was obtained by summing the linear expansion/contraction of all nodes in that row.

VI. The calculation of the linear expansion/contraction of the second row was started as soon as all nodes in that row had solidified and so on.

VII. As soon as a row solidified, its starting length at that instant was taken as the average length of the rows adjacent to it that had already solidified.

VIII. The average length of all solidified rows at any given time step was taken as the billet dimension at that time step.

IX. This procedure was repeated until the bottom of the mould was reached.

The procedure outlined above could be done for rows or columns but the interaction between rows and columns was not considered. The model does not account for the bulging of steel shell on account of ferro-static pressure but this approach enables quantification of mould taper necessary to prevent the outward bulging of the shell.

6.3 Temperature Dependent Thermal and Mechanical Coefficients

In this plant trial, special attention was given to the thermal-mechanical response of the mould during casting of martensitic and austenitic stainless steel. As we know, stainless steels offer a large variety of structures and therefore a large variety of physical, thermal, and mechanical properties. A large number of diagrams have been established for predicting the microstructure and properties of stainless steel from the composition of the austenitic filler metal. The Schaeffler diagram (Figure 6.3) is commonly used to obtain an indication of the structure and ferrite contents of the steel grades. According to the chemical composition of different heats during Atlas Steel Plant Trial, corresponding
thermal-mechanical parameters are introduced into the mathematical model to make the pre-existing code suitable for the calculation of shrinkage of the stainless steel strand.

6.3.1 Parameters for Austenitic Stainless Steel at Elevated Temperature

Heat 6331 is AISI375 grade, which is a austenitic stainless steel in Schaeffler diagram. (Figure 6.3) The equivalent Ni and Cr contents are 10.40 and 18.16 respectively. The microstructure of this grade is similar to AISI304 in the Schaeffler diagram. So the temperature dependent specific heat, heat conductivity and linear thermal expansion coefficients of AISI375 come from AISI304, which are shown in Table 6.1 to Table 6.3.

6.3.2 Parameters for Martensitic Stainless Steel at Elevated Temperature

Heat 8113J, 8115J and 8119J are AISI416 grade. According to their chemical composition, they fall into the field of M+F in Schaeffler diagram. As thermal-mechanical property data for AISI416 are not available, parameters either from AISI420 or AISI410 which are very similar to AISI416 in microstructure were introduced in shrinkage calculation. The temperature dependent specific heat, heat conductivity and linear thermal expansion coefficient are shown in Table 6.4 to Table 6.6 respectively.
Figure 6.3 Shaeffler Diagram\textsuperscript{[55]}
Table 6.1  Heat capacity for austenitic stainless steel\textsuperscript{[102]}

<table>
<thead>
<tr>
<th>Heat Capacity (kJ/kg K)</th>
<th>Temperature (K)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$C_p = 0.43895 + 1.98 \times 10^{-4} T$</td>
<td>$T \leq 773$</td>
</tr>
<tr>
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Table 6.2  Heat conductivity for austenitic stainless steel\textsuperscript{[102]}

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### Table 6.3 Instantaneous coefficient of thermal linear expansion for austenitic stainless steel [103]

<table>
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<th>Instantaneous Coefficient of Thermal Linear Expansion (10^{-6} K^{-1})</th>
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<td>21.69*</td>
</tr>
</tbody>
</table>

Note: * Extrapolated values
### Table 6.4 Heat capacity for heat martensitic stainless steel

<table>
<thead>
<tr>
<th>Heat Capacity (kJ/ Kg K)</th>
<th>Temperature (K)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$C_p = 0.44216 + 1.172 \times 10^{-4} T$</td>
<td>$T \leq 528$</td>
</tr>
<tr>
<td>$C_p = 0.12634 + 7.154 \times 10^{-4} T$</td>
<td>$528 \leq T \leq 1150$</td>
</tr>
<tr>
<td>$C_p = 1.91904 - 1.5875 \times 10^{-2} T$</td>
<td>$1150 \leq T \leq 1173$</td>
</tr>
<tr>
<td>$C_p = 0.569$</td>
<td>$1173 \leq T$</td>
</tr>
</tbody>
</table>

Note: quoted from AISI 420

### Table 6.5 Heat conductivity for heat martensitic stainless steel

<table>
<thead>
<tr>
<th>Heat Conductivity (W/ m K)</th>
<th>Temperature (K)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$k = 23.112 + 0.014035 T - 0.11137 \times 10^{-4} T^2$</td>
<td>$T \leq 1060$</td>
</tr>
<tr>
<td>$k = 13.731 + 0.011107 T$</td>
<td>$1060 \leq T \leq 1755$</td>
</tr>
<tr>
<td>$k = 160.382 - 0.0724 T$</td>
<td>$1755 \leq T \leq 1805$</td>
</tr>
<tr>
<td>$k = 8.61 + 0.01168 T$</td>
<td>$1805 \leq T$</td>
</tr>
</tbody>
</table>

Note: quoted from AISI 410
Table 6.6 Instantaneous coefficient of thermal linear expansion for martensitic stainless steel [1031]

<table>
<thead>
<tr>
<th>Temperature (K)</th>
<th>Instantaneous Coefficient of Thermal Linear Expansion ($10^{-6} \text{ K}^{-1}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>10</td>
<td>0.01</td>
</tr>
<tr>
<td>20</td>
<td>0.05</td>
</tr>
<tr>
<td>30</td>
<td>0.20</td>
</tr>
<tr>
<td>40</td>
<td>0.50</td>
</tr>
<tr>
<td>50</td>
<td>1.00</td>
</tr>
<tr>
<td>60</td>
<td>1.65</td>
</tr>
<tr>
<td>70</td>
<td>2.40</td>
</tr>
<tr>
<td>80</td>
<td>3.15</td>
</tr>
<tr>
<td>90</td>
<td>3.90</td>
</tr>
<tr>
<td>100</td>
<td>4.55</td>
</tr>
<tr>
<td>150</td>
<td>7.05</td>
</tr>
<tr>
<td>200</td>
<td>8.8</td>
</tr>
<tr>
<td>250</td>
<td>10.00</td>
</tr>
<tr>
<td>273</td>
<td>10.37</td>
</tr>
<tr>
<td>293</td>
<td>10.65</td>
</tr>
<tr>
<td>300</td>
<td>10.72</td>
</tr>
<tr>
<td>350</td>
<td>11.22</td>
</tr>
<tr>
<td>400</td>
<td>11.58</td>
</tr>
<tr>
<td>500</td>
<td>12.08</td>
</tr>
<tr>
<td>600</td>
<td>12.43</td>
</tr>
<tr>
<td>70</td>
<td>12.80</td>
</tr>
<tr>
<td>800</td>
<td>13.10</td>
</tr>
<tr>
<td>900</td>
<td>13.30</td>
</tr>
<tr>
<td>1000</td>
<td>13.55</td>
</tr>
<tr>
<td>1100</td>
<td>13.75</td>
</tr>
<tr>
<td>1100$^a$</td>
<td>18.13</td>
</tr>
<tr>
<td>1150</td>
<td>19.07</td>
</tr>
<tr>
<td>1200</td>
<td>20.03</td>
</tr>
<tr>
<td>1250</td>
<td>21.01</td>
</tr>
<tr>
<td>1300</td>
<td>21.99</td>
</tr>
<tr>
<td>1350</td>
<td>23.19$^*$</td>
</tr>
<tr>
<td>1400</td>
<td>24.30$^*$</td>
</tr>
<tr>
<td>1450</td>
<td>25.42$^*$</td>
</tr>
<tr>
<td>1500</td>
<td>26.55$^*$</td>
</tr>
</tbody>
</table>

Note: * Extrapolated values
Chapter 7  HEAT FLUX CALCULATION

Heat flux profiles for different steel grades and operating conditions were calculated employing INVERSE mathematical model. Model predictions of mould heat flux are analyzed in this chapter.

7.1 Heat Flux Profiles

Axial heat flux profiles were obtained by applying the mathematical model of heat flow in the mould. Samples of the results are presented in Figure 7.1 and Figure 7.2. It is evident that the peak heat fluxes, close to the meniscus, are in the $4500 \sim 5800 \text{ kW/m}^2$, which is significantly higher than previously measured by Pinheiro\textsuperscript{[56]} (Figure 7.3) during continuous casting of billets with mould flux lubrication. In previous work the maximum mould heat fluxes recorded were approximately $2500\sim3000 \text{ kW/m}^2$. This difference can be attributed to the casting speeds, which were significantly lower ($1.0\sim1.3 \text{ m/min}$) compared to the range in this trial ($1.9\sim2.5 \text{ m/min}$). It is also evident that the heat flux profiles on the four faces of the mould were different from one another, with the inside curved wall having the highest heat fluxes near the meniscus and the lowest values in the lower part of the mould. Comparing the four faces for a given heat, it is evident that the heat flux in the right straight wall and outside curved wall are consistently higher than the other two faces, while the average heat flux on the inside curved wall is consistently the lowest. The reason
for this is not quite clear, but the different melting behavior of the mould powder could be one of the factors. Taken Figure 7.1 as an example, the shape of the heat flux profile can be decomposed into four segments, as follow:

I. From the top of the mould down to 135 mm, the profile is characterized by very low values (< 700 kW/m\(^2\)) of heat flux;

II. From 135 to 190 mm, the meniscus region, the heat flux increase sharply from 700 to 5940 kW/m\(^2\). The maximum value, 5940 kw/m\(^2\), always occurs 30 to 50 mm always from the meniscus, which was 135 mm from the top of the mould in this plant trial.

III. Below the meniscus the heat flux decreases in magnitude, shrinkage of the solidifying shell gives rise to an increase in the gap between the shell and the mould which leads to an increase in the thermal resistance, thereby lowering the heat transfer to the mould. Since the initial shrinkage is larger due to the initial higher heat flux and also due to alpha to delta phase transformation, the heat flux reach its lowest value at 210 mm below the top of the mould. Ferro-static pressure acting on the thin shell pushes it back towards the mould wall, decreasing the air gap, thus increasing heat flux thereafter.

IV. From 210 mm further down the mould the progress of solidification leads to a progressive increase in the gap with a consequent decrease in heat flux. The rebounds in heat flux along the inside curved wall verified at 280, 400 and 520mm can be associated with slag rim which was formed and retained during the interval when the steel grade was changed and did not remelt in the following continuous casting process. The onset of slag rim had a larger thermal resistance for the heat
transfer, causing the temperature drop in the corresponding spots along the inside curved wall.

In this study, the maximum heat flux value was found to be below the meniscus. The same phenomena also happened in Pinheiro\cite{56} and Chandra's\cite{57} studies. This observation is probably caused by the inherently instability of the meniscus in billet casting when metal is controlled via casting speed.

7.1.1 Typical Heat Flux Profile for Austenitic Stainless Steel

Typical heat flux profile for austenitic stainless steel is presented in Figure 7.1. The casting speed is 2.57 m/min. The average heat fluxes and peak heat fluxes for each of the four faces are listed in Table 7.1, together with them are the total average heat fluxes for all the four faces. With different casting speeds, the average heat flux and peak heat flux range from 1700 to 2200 kW/m² and 3556 to 5940 kW/m² respectively for different faces.

<table>
<thead>
<tr>
<th>Table 7.1 Heat flux for austenite stainless steel under different casting speed</th>
</tr>
</thead>
<tbody>
<tr>
<td>heat flux (kw/m²)</td>
</tr>
<tr>
<td>Average (kw/m²)</td>
</tr>
<tr>
<td>peak (kw/m²)</td>
</tr>
<tr>
<td>Inside</td>
</tr>
<tr>
<td>Left side</td>
</tr>
<tr>
<td></td>
</tr>
<tr>
<td>Outside</td>
</tr>
<tr>
<td></td>
</tr>
<tr>
<td>Right side</td>
</tr>
<tr>
<td></td>
</tr>
<tr>
<td>Total Ave.</td>
</tr>
</tbody>
</table>

Note: CSp means casting speed

7.1.2 Typical Heat Flux Profiles of Martensitic Stainless Steel
In the plant trial, data from three heats were collected for martensitic stainless steel. Figure 7.2 is a typical heat flux profile for martensitic stainless steel. The compositions of the three heats are presented in the previous section and show slightly difference. During the plant trial, the casting speed for the three heats also show some difference. Average heat flux and peak heat flux are listed in Table 7.2. The average heat flux and peak heat flux range from 1545 to 1943 kw/m$^2$ and 2967 to 5866 kw/m$^2$ respectively.

Table 7.2 Heat flux for martensitic stainless steel for three heats, casting speed for heat 8113, 8115 and 8119 is 1.90 m/min

<table>
<thead>
<tr>
<th></th>
<th>Heat 8113</th>
<th></th>
<th>Heat 8115</th>
<th></th>
<th>Heat 8119</th>
</tr>
</thead>
<tbody>
<tr>
<td>heat flux (kw/m$^2$)</td>
<td>Average</td>
<td>peak</td>
<td>Average</td>
<td>peak</td>
<td>Average</td>
</tr>
<tr>
<td>Inside</td>
<td>1610.38</td>
<td>5463.2</td>
<td>1581.24</td>
<td>5258.4</td>
<td>1545.46</td>
</tr>
<tr>
<td>Left side</td>
<td>1645.58</td>
<td>4097.6</td>
<td>1711.73</td>
<td>4412.9</td>
<td>1560.33</td>
</tr>
<tr>
<td>Outside</td>
<td>1775.96</td>
<td>5435.1</td>
<td>1791.73</td>
<td>4814.9</td>
<td>1776.72</td>
</tr>
<tr>
<td>Right side</td>
<td>1813.89</td>
<td>3613.8</td>
<td>1943.54</td>
<td>4620.3</td>
<td>1737.09</td>
</tr>
<tr>
<td>Total Ave.</td>
<td>1711.89</td>
<td></td>
<td>1757.05</td>
<td></td>
<td>1654.91</td>
</tr>
</tbody>
</table>

7.1.3 Alternating Method for Heat Flux Calculation

The total heat extracted from the mould can be calculated with the following formulas:

\[ Q = W \rho C_p (T_{out} - T_{in}) \]  

\[ Q \text{ --- total heat flux, kW} \]

\[ W \text{ --- cooling water flow rate, L/s} \]
Chapter 7  HEAT FLUX CALCULATION

\[ q = \frac{Q}{S} \]  \hspace{1cm} (7.2)

\( S \) --- contact area between cooling water and mould, m\(^2\)

average heat flux from the mould is:

Thermocouples were installed in the water channel during the plant trial. For heat 6331 and 8115, the differences between in-water and out water temperature were 4.5°C and 3.7°C receptively and the cooling water flow rates were 1619 L/min. Corresponding average heat fluxes were calculated to be 1709 and 1405 kW/m\(^2\) for heat 6331 and 8115. Those values are 12% and 21% lower than the model predictions respectively. The difference could be attributed to both measurement and the INVERSE model.

7.2 Influence of Process Variables on Mould Heat Flux

7.2.1 Casting Speed

The influence of casting speed on mould heat transfer can be observed in Table 7.1, which show mould heat flux data for the austenitic stainless grade at speeds of 1.9, 2.4 and 2.57 m/min respectively. Consistent with the temperature data shown in the previous section, it is evident that the maximum peak heat fluxes, which generally occurs on the inside curved wall, increases from 5176 to 5404 kW/m\(^2\), when the casting speed increase from 1.9 to 2.4 m/min. A further increase in speed to 2.5 m/min gives rise to a maximum mould heat flux of 5940 kW/m\(^2\). The average mould heat flux profiles for the three speeds are shown in Figure 7.4, which also shows that the mould heat transfer rates are consistently higher with increased casting speed, particularly in the lower part of the
mould. When casting speed increases, the residual time of the strand in the mould becomes less, leading to a shallower temperature gradient in the shell. The shell contraction caused by the temperature variation thus decreased, leading to a smaller gap size between mould and strand, giving rise to a higher heat flux at higher casting speed.

7.2.2 Steel Grade

It is shown from Table 7.1 and 7.2 that mean heat flux was highest for the austenitic stainless steel heat ~1900 to 2000 kW/m²; the corresponding values were 15% lower for martensitic grades ~1650 to 1700 kW/m². Comparison of local heat fluxes of all the four faces between martensitic and austenitic stainless steel grades were shown in Table 7.3. Figure 7.5 compares the average heat flux profiles for the austenitic and martensitic grades at the same casting speed. Interestingly, at the meniscus, the heat transfer for martensitic grades steel is higher compared with austenitic stainless steel. This is not surprising as we know that if the ratio of equivalent Ni to equivalent Cr is 0.55–0.60 for austenitic stainless steel, the delta to gamma transformation occurs just at the end of solidification, causing a large gap at the meniscus region. For Heat 6331, this ratio was 0.57, thus the heat transfer rate was reduced. Another evidence is the oscillation marks are shallower for martensite grade (0.2 ~ 0.35 mm) compared to the austenitic grade (0.4 ~ 0.5 mm ). However below the meniscus, heat transfer for the martensitic grades are lower than for the austenitic steel billets. It will be shown in the next section that this difference can be attributed to differences in shrinkage of the billet and non-optimum mould taper design, resulting in an air gap which is larger for the martensitic steels.
Table 7.3 Comparison of the local heat flux between martensitic and austenitic stainless steel, from Heat 6331 and 8119 respectively, casting speed was 1.90 m/min, mould powder CT3

<table>
<thead>
<tr>
<th>Position (mm)</th>
<th>Inside (M)</th>
<th>Inside (A)</th>
<th>Left side (M)</th>
<th>Left side (A)</th>
<th>Outside (M)</th>
<th>Outside (A)</th>
<th>Right side (M)</th>
<th>Right side (A)</th>
</tr>
</thead>
<tbody>
<tr>
<td>104.0</td>
<td>691.2</td>
<td>610.3</td>
<td>706.7</td>
<td>671.9</td>
<td>647.9</td>
<td>437.9</td>
<td>688.9</td>
<td>568.1</td>
</tr>
<tr>
<td>121.3</td>
<td>691.2</td>
<td>610.3</td>
<td>706.7</td>
<td>671.9</td>
<td>647.9</td>
<td>437.9</td>
<td>688.9</td>
<td>568.1</td>
</tr>
<tr>
<td>138.7</td>
<td>598.8</td>
<td>1144.8</td>
<td>745.6</td>
<td>1572.2</td>
<td>861.1</td>
<td>2741.8</td>
<td>591.3</td>
<td>1400.8</td>
</tr>
<tr>
<td>156.0</td>
<td>1790.5</td>
<td>5176.6</td>
<td>1259.8</td>
<td>3266.3</td>
<td>1011.9</td>
<td>2407.2</td>
<td>1141.8</td>
<td>4824.4</td>
</tr>
<tr>
<td>173.3</td>
<td>1641.1</td>
<td>1745.5</td>
<td>1916.4</td>
<td>2680.2</td>
<td>4665.4</td>
<td>4000.4</td>
<td>4265.5</td>
<td>3286.1</td>
</tr>
<tr>
<td>190.7</td>
<td>5866.8</td>
<td>4806.2</td>
<td>5327.3</td>
<td>3556.6</td>
<td>2967.6</td>
<td>2044.8</td>
<td>3597.4</td>
<td>2665.4</td>
</tr>
<tr>
<td>277.3</td>
<td>1431</td>
<td>1536.4</td>
<td>1634.4</td>
<td>1937.2</td>
<td>2418.1</td>
<td>2627.6</td>
<td>2388.9</td>
<td>2726.6</td>
</tr>
<tr>
<td>346.7</td>
<td>2000.9</td>
<td>2375.8</td>
<td>1813.5</td>
<td>2194.7</td>
<td>2298.9</td>
<td>2688.2</td>
<td>2283.3</td>
<td>2864.2</td>
</tr>
<tr>
<td>398.7</td>
<td>1542.4</td>
<td>1725.3</td>
<td>1844.4</td>
<td>2301.3</td>
<td>2001.6</td>
<td>2271.5</td>
<td>1910.4</td>
<td>2412.4</td>
</tr>
<tr>
<td>468.0</td>
<td>2173.2</td>
<td>2533.8</td>
<td>1833.9</td>
<td>2283</td>
<td>1786.8</td>
<td>2157.1</td>
<td>1773.4</td>
<td>2269.3</td>
</tr>
<tr>
<td>520.0</td>
<td>1043.4</td>
<td>1149</td>
<td>1580.3</td>
<td>1745.9</td>
<td>1907.8</td>
<td>2152</td>
<td>1722.6</td>
<td>2201.3</td>
</tr>
<tr>
<td>641.3</td>
<td>1671.9</td>
<td>1909.6</td>
<td>1487.3</td>
<td>1690.7</td>
<td>1484.5</td>
<td>1623.1</td>
<td>1733.4</td>
<td>2194.9</td>
</tr>
<tr>
<td>693.3</td>
<td>1566.3</td>
<td>1835.6</td>
<td>1493.4</td>
<td>1773.5</td>
<td>1781.2</td>
<td>2061.8</td>
<td>1413.2</td>
<td>1820.4</td>
</tr>
</tbody>
</table>

A: austenitic stainless steel (heat 6331), M: martensitic stainless steel (heat 8119)
Figure 7.1 Heat flux profiles for the all four faces for austenitic stainless steel when casting speed is 2.57 m/min
Figure 7.2  Heat flux profiles for martensitic stainless steel, heat 8119, casting speed is 1.90 m/min
Figure 7.3  Heat flux profile for carbon steel, casting speed ~1.1m/min, mould powder lubrication (0.31≤%C≤0.33), reprinted from Pinheiro' Ph.D thesis[56]
Figure 7.4  The effect of casting speed on the heat flux profiles for austenitic stainless steel, Heat 6331
Figure 7.5  The effect of steel grades on the heat flux profiles in the casting of stainless steel, casting speed is 1.9m/min
Chapter 8 SHELL GROWTH, SHRINKAGE AND MOULD TAPER DESIGN

The billet mid-face temperature, shell thickness and shrinkage profile were all obtained from the mathematical model of billet solidification and shrinkage. The model was employed using the following logic:

1. The mould temperature profiles were taken from the plant trial as a 60~80 minutes average of mould temperature, to obtain a sense of average heat transfer.
2. Mould heat transfer was calculated using the INVERSE model described in Chapter 6.
3. Mould temperature distribution was obtained using the mould heat transfer model, and mould distortion was evaluated using this temperature distribution.
4. Chandra’s shrinkage calculation model\textsuperscript{571} was run using the mould heat flux profiles as boundary condition.
5. The initial billet dimension was taken to be the dimension of the distorted mould around the meniscus.

Six heats from Atlas Steel plant trial were investigated for their shell growth, billet shrinkage and mould-shell interaction.

8.1 Billet Solidification and Shell Thickness
8.1.1 Billet Corner and Mid-face Temperature

The predicted thermal history of a billet passing through the mould for martensitic stainless steel and austenitic stainless steel grades are shown in Figure 8.1 to Figure 8.3. At the beginning, owing to the two dimensional heat transfer, the temperature at the corner of the billet declines very quickly. Soon after the formation of the shell, an air gap is formed because of the rapid cooling around the corner of the strand, leading to the low heat transfer rate between the strand and mould. After reaching the lowest value at 80 ~ 100 mm below meniscus, the temperature begins to rebound. The surface temperature of the billet, at mould exit, for all grades is approximately 995 to 1102 °C, depending on the heat flux and casting speed for each heat. Taking heat 6331, the austenitic stainless steel, for example, which had a pouring temperature of 1530°C, the temperatures at mould exit along the corner and mid-face dropped to 1012 and 995 °C respectively at a casting speed of 1.90m/min, while the corresponding temperatures were 1102 and 1045°C when the casting speed was increased to 2.57 m/min. The centerline temperatures dropped to 1471 and 1484°C when the casting speeds were 1.90 and 2.57m/min respectively.

8.1.2 Shell Growth

Typical shell growth for both martensitic stainless steel and austenitic stainless steel is also presented in Figure 8.1 to 8.3. There are significant differences in the exit shell thickness between different heats because of the difference in the liquidus and solidus temperatures, freezing range, the superheat, heat flux and casting speed. The austenitic and martensitic steels have freezing ranges between 50 and 55 °C. However the liquidus and solidus temperatures of the austenitic stainless steel are significantly lower than the values for martensitic stainless steels; 1454 °C and 1399 °C versus 1510 °C and 1454 °C.
Owing to the high superheat (68°C) for the austenitic grades, shell growth of the strand at the meniscus region is delayed; thus the shell thickness for the austenitic grade at the exit of the mould is 6.36 ~ 8.30 mm as compared to 9.0 ~ 10.8 mm for the martensitic grade. Casting speed has a significant effect on the shell thickness as can be seen in Figures 8.1 and 8.2, which show the shell thickness growth for the austenitic grade at casting speeds of 1.9, 2.57 m/min. The shell thickness at the mould exit are 8.30, 6.36 mm respectively, which represents a reduction of 23% for the higher casting speed. Clearly with increasing casting speed there is a greater risk of breakouts due to the drastic reduction in shell thickness. The increase in heat flux with casting speed does not compensate for the influence of the reduction in strand residence time in the mould. Calculated shell thickness for all the six heats in the plant trial are listed in Table 8.1.

Table 8.1 Calculating results from mathematical model for different heats in steel plant trial, billet size: 114x114mm, mould length: 780mm, mould powder lubrication

<table>
<thead>
<tr>
<th>heat No.</th>
<th>steel grade</th>
<th>casting speed (m/min)</th>
<th>superheat (°C)</th>
<th>shell thickness (mm)</th>
<th>billet shrinkage (mm)</th>
<th>gap at mould exit (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>6331</td>
<td>AL375-0</td>
<td>1.90</td>
<td>68</td>
<td>8.30</td>
<td>2x0.47</td>
<td>0.171</td>
</tr>
<tr>
<td>6331</td>
<td>AL375-0</td>
<td>2.40</td>
<td>68</td>
<td>7.22</td>
<td>2x0.39</td>
<td>0.176</td>
</tr>
<tr>
<td>6331</td>
<td>AL375-0</td>
<td>2.57</td>
<td>68</td>
<td>6.36</td>
<td>2x0.39</td>
<td>0.177</td>
</tr>
<tr>
<td>8113</td>
<td>T416F1</td>
<td>1.90</td>
<td>11</td>
<td>10.68</td>
<td>2x0.59</td>
<td>0.260</td>
</tr>
<tr>
<td>8115</td>
<td>T416F1</td>
<td>1.90</td>
<td>13</td>
<td>10.85</td>
<td>2x0.61</td>
<td>0.295</td>
</tr>
<tr>
<td>8119</td>
<td>T416F2</td>
<td>1.90</td>
<td>12</td>
<td>9.02</td>
<td>2x0.48</td>
<td>0.171</td>
</tr>
</tbody>
</table>
8.1.3 Billet Shrinkage

Shrinkage calculation was also performed using a mathematical model of billet shrinkage developed by Chandra, etc.,[57] and modified for the calculation of stainless steel. Billet shrinkage for different steel grades and different casting speeds were shown in Figure 8.4 and Figure 8.5. For heat 6331, which is austenitic stainless steel, when the casting speed was increased from 1.90 to 2.57 m/min, the billet shrinkage decreased from 0.96 to 0.78mm. This is evident as increasing casting speed will reduce the resident time of the strand in the mould, leading to a thinner shell thickness, a higher temperature field in the shell thus less billet contraction. Figure 8.5 has shown the billet shrinkages of martensitic and austenitic stainless steel at a casting speed of 1.90 m/min. It can be observed that at the same casting speed of 1.90 m/min, the billet shrinkage at the bottom of the mould for martensitic stainless steel was 1.20 mm, which is higher than 0.96 mm for the austenitic stainless steel shown in Figure 8.4. This is consistent with the lower heat flux in the lower portion of the mould for martensitic stainless steel as shown in Figure 7.4. Billets shrinkage values for all the stainless steel heats are listed Table 8.1. The calculating results show that in the mould the strands of the martensitic stainless steel contract more than austenitic stainless steel grade although they have lower average heat fluxes.

8.2 Binding Evaluation between Strand and the Mould and Billet Quality

8.2.1 Binding Evaluation between Strand and the Mould

Billet dimension along the mould length was compared to mould dimension. During the plant trial, the mould taper is 1.2%/m from top to 350 mm down the top and 0.5%/m in the lower part of the mould. Mould thermal distortion and expansion are calculated
employing the temperature field in heat flux calculation. Billet dimensions along the mould length were compared to mould dimensions to investigate mould/strand gap, mould/strand binding. Some results of the calculation are shown in Figure 8.6 to Figure 9, which were for austenitic stainless steel and martensitic stainless steel under different operating conditions.

In above cases, we have observed the gap formation between strand and the mould surface along the mid-face of the mould in the lower portion of the mould. The size of the air gap were different for different heats. Steel grade, casting speed and superheat are among the most important factors that influence the size of the air gap. For austenitic stainless steel, the formation of air gap began at about 300 ~ 320 mm below the top of the mould with the mould taper in the upper portion (0 ~ 350 mm) of 1.2%/m. When the casting speed was raised from 1.9 to 2.57 m/min, the size of the air gap at the mould exit decreases from 0.177 mm to 0.171 mm, in accordance with heat flux increase shown in Figure 7.3. The strong influence of steel grade on the air gap size can also be observed in Table 8.1. Take heat 8115 which was martensitic stainless steel grade for example, when the casting speed was 1.90 m/min, the size of the air gap at the mould exit are 0.295 mm, 42% higher than 0.171 mm for heat 6331 which is austenitic stainless steel grade. This air gap had established a larger thermal resistance at the mould exit in the casting of martensitic stainless steel in view of heat transfer. Heat flux calculation has shown that at the mould exit, the heat flux is 1665 kW/m² for martensitic stainless steel, 22% lower than that for austenitic stainless steel which is 2031 kW/m².

Binding may be interpreted by simply comparing the billet and distorted mould dimensions at any axial position. If the billet is larger than the mould, the billet may bind or
jam in the mould with resulting generation of high axial forces on the billet shell such that cracks may form or a breakout may be initiated. As it can be shown in Figure 8.6 to 8.10, the taper is most seriously influenced by distortion near the meniscus. While lower in the mould, the distortion impacts the taper less severely. Also from those figures, binding phenomena can be observed in the upper part of the mould for austenitic stainless steel. The occurrence of lightly binding in the casting of austenitic stainless steel may be related to its high superheat (68°C), the shell growth was delayed, the shell became thinner and temperature of the shell was higher, causing less billet contraction. Lightly binding was also found in the upper portion of the mould in heat 8113 which was a martensitic stainless steel grade. (Figure 8.10)

8.2.2 Billet Quality

Surface quality evaluation for the samples from heat 6331, 6327 and 6328, which were all austenitic stainless steel was shown in Table 5.2. Surface defects related to interaction of the mould and strand around meniscus such as laps and bleeds were observed in the inspection of those austenitic stainless steel samples, while for martensitic stainless steel, no such surface defects were observed. We already know that in this plant trial the temperature fluctuations of the mould at meniscus is sever. The standard deviations of the local temperature singles at meniscus region range from 7 to 15 for all the four faces in the casting of austenitic stainless steel. But they are usually less than 5 in the lower portion of the mould. This means a serious metal level fluctuation in the casting process. The metal level fluctuations combined with uneven mould flux distribution influenced the generation of laps and bleeds, consistent with S.Kumar’s description.[58] Another important factor is the initial solidification process of austenitic stainless steel.
The ratio of equivalent Ni and Cr for those austenitic heats ranges from 0.56~0.64. The solidification process followed the mode:

\[
\text{liquid steel} \rightarrow \text{L+}\delta \rightarrow \text{L+}\delta+\gamma \rightarrow \delta+\gamma
\]

The delta to gamma transformation accompanying the solidification process will cause volume contraction similar to peritectic steel and the existence of residual delta ferrite can impair the hot ductility of the austenitic stainless steel. The delta to gamma phase transformation during initial solidification process causes volume contraction, thereby forming a large air gap between the mould and the shell. The air gap around meniscus and metal level fluctuations not only delays shell growth but also causes uneven shell growth. It is assumed that the non-uniform distribution of the mould powder, together with a thin solid shell in the austenitic stainless steel grades as a result of high superheat and larger air gap, causes tearing of the solid shell and the formation of bleeds and laps.

Mould powder properties strongly influence the surface quality of the stainless steel. Different mould powders were tested in heat 8110, which was Al375 steel grade with a equivalent Ni/Cr ratio of 0.58. The properties of the mould powder are listed in Table 4.2. When mould powder CT3 was used for the first 47 minutes, the casting operation was smooth with a relatively uniform oscillation marks. No depression was witnessed on the surface of the billet. Then mould powder CT3 was replaced with mould powder BL23 in the following 15 minutes. During this short period, 2 breakouts, 1 lap, 6 transverse depressions, 15 longitudinal depressions were observed, with a very rough surface. The operation was unstable. The viscosity of the mould powder BL23 is only 1.3 compared with 4.4 of CT3. The melting temperature is 925°C for BL23, and 983°C for mould powder CT3. From these observations we can see that in the casting of austenitic stainless
Chapter 8 SHELL GROWTH, SHRINKAGE AND MOULD TAPER DESIGN

steel, the frequency of the occurrence of transverse and longitudinal depression is much higher with mould powder of low viscosity and melting temperature. We know that at higher casting speed, the consumption rate of mould powder decreases. Mould powder with physical properties of high viscosity and high surface tension are effective for slag infiltration. Under high casting speed, the low infiltration rate of the mould fluxes combined with metal fluctuations may lead to a non-uniform lubricating film. Variable heat flux causes non-uniform shell growth of the strand. This trend is further enhanced by the delta to gamma phase transformation during the initial solidification of the austenitic stainless steel and the inappropriate mould taper design at the meniscus. The interaction between the mould and the strand in the meniscus region is responsible for quality problems such as transverse and longitudinal depression and even breakout. While with mould powder of high viscosity, we have found few quality problems because of the improvement of the mould flux infiltration rate. So in high speed casting of austenitic stainless steel billets, the optimum utilization of mould powder is very important for the control of surface quality.

Longitudinal depression was also observed in heat 6331. For austenitic stainless steel, the shell thickness growth may be delayed and tend to be non-uniform due to the high superheat and the large air gap in meniscus region. The inappropriate mould taper in the upper portion of the mould cause binding between the mould and the solid shell. The mould might squeeze the thin solid shell that buckles inward, generating longitudinal depression.

Longitudinal depression was also found in martensitic stainless steel which was shown in Figure 5.11. The metal level fluctuations and mould powder behavior may cause uneven
shell growth in the meniscus region. The inappropriate mould taper in the upper portion of the mould cause binding (Figure 8.10) between mould and the strand. The squeezing effect of the mould wall cause the inward bucking of the shell, leading to the forming of longitudinal depression.

Although we find air gap in most cases in the evaluation of the interaction between the strand and mould for martensitic stainless steel grade when the calculations were based on the idea of average temperature, binding might happen due to unsteady heat transfer in actual casting operations. It is well known that the continuous casting process is a dynamic process. The mould is in constant oscillation, so the metal level fluctuates. Combining with the variations of other operating parameters such as flow rate of molten steel, mould powder behavior and metal level, the temperature of the mould wall will respond correspondingly, which means the change of heat transfer in the mould. For heat 8119, which was a martensitic stainless steel, an air gap can be observed in the shrinkage calculation using average temperatures. The instantaneous thermal reaction of the mould wall was further investigated. The sum of the temperature signals of all the thermocouples for all the faces at the same moment was calculated. The temperatures at the moment of minimum sum value were chosen to investigate the interaction between the mould wall and the strand. Figure 8.10 demonstrate the heat flux profiles of maximum and minimum heat fluxes in the proceeding of heat 8119, which was a martensitic stainless steel grade. The average heat flux for the moment of maximum heat flux 1933 kW/m$^2$, 53% higher than the minimum value which was 1261 kW/m$^2$. This difference mainly came from the lower portion of the mould. The phenomena strongly suggest the influence of the upset of operating variables. The valley in the profiles in the meniscus region could be attributed to
the slag rim. Figure 8.11 show the interaction of the mould wall and the strand at that moment when the heat flux is minimum. It can be clearly observed that binding happened from 220 mm below the top of mould. In the lower portion of the mould, the mould wall severely squeezed the strand, which may leading to the formation of longitudinal depression. So stabilizing the operating variables was a very important condition in quality control.

8.3 Mould Taper Design

Discussion in the previous sections have clearly revealed the significant effect of the pre-existing mould taper on the heat flux along the mid-face of the mould. It is impossible to use a universal mould with one type of taper design to cast different steel grades under different casting conditions. Optimization of mould taper design should be conducted for billet casting.

8.3.1 Method of Mould Taper Design

The wall of the mould are tapered inwardly so as to enhance the rate of mould heat extraction by reducing the air gap between the mould and the billet. It is not sufficient, however, to design a mould taper based solely on the shrinkage profile of the billet, because during service, the mould wall distorts and moves away from the billet, opening up an air gap further. Thus the design of the taper must take into account of the formation of an air gap caused by both the distortion of the mould while in use and shrinkage of the billet. Mould tapers are expressed in %/m and are based on the expression given below:

\[ \text{Taper}\left(\%\right) = \left(\frac{M_i - M_n}{M_i}\right) \times 100\% \]  

\( (8.1) \)
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where \( M_t \) and \( M_b \) are the dimensions of the mould at the top and bottom respectively and \( M_l \) is the length of the mould in meters.

8.3.2 Suggested Mould Taper

Effective mould tapers have been calculated for both martensitic and austenitic stainless steel, taking into account the effect of speed variations. Based on the design philosophy discussed above, we assumed that an ideal mould taper should make the inner surface of the mould coincide with the billet surface during casting operation. Cold mould dimension was further inversely calculated so as to get an effective mould taper. The ideal mould taper should be an curved shape, steep in the meniscus region, becoming shallower when going down from top of the mould. This is on account of the large air gap formed by the rapid initial shrinkage of the billet and the distortion of the mould away from the billet. The results are shown in Figure 8.11 and Table 8.1. It is clear that the mould taper in the plant trial, which was 1.2%/m from 0~350 mm and 0.5%/m from 350~780 mm below the top of the mould, was not appropriate for stainless steel billet casting. For austenitic stainless steel billet casting, the predicted taper in the upper part of the mould, from meniscus to approximately a position 350 mm below the top of the mould, lies in the range of 1.52 to 1.26%/m. This taper is required to compensate for the effects of the outward expansion of the mould wall at the meniscus and billet shrinkage. In the second region, which extends from a position of 350 mm from top of the mould down the bottom of the mould, the required tapers lie in the range 1.10 ~ 1.11%/m. While for martensitic stainless steel, owing to its higher heat flux around meniscus, taper requirement for the upper part of the mould is 2.26%/m and 1.33 %/m in the lower part of the mould. The low
superheat (<15°C), high liquidus temperature (1510°C) and rapid contraction of the shell may contribute to the high taper requirement of martensitic stainless steel.

The effect of increasing casting speed on taper can be observed from Figure 8.11 and 8.13. For austenitic grade H6331, an increase in speed from 1.9 to 2.57 m/min reduce the upper taper from 1.52 to 1.265%/m and the lower taper from 1.11 to 1.10 %/m. It is seen that despite the increase in mould heat transfer with increasing casting speed, the reduction in residence time in the mould results in less shrinkage; thus less taper is required.

Mould taper is also seen to be extremely sensitive to superheat. In heat 6331, which was austenitic stainless steel, the high superheat (68°C), the lower liquidus temperature strongly delayed the shell growth below meniscus. This can be easily observed from Figure 8.10 and Figure 8.11. The taper requirement in this region was primarily to compensate for the effects of the outward expansion of the mould wall at the meniscus.
Figure 8.1  Temperature profiles and shell growth of the strand down the top of the mould for austenitic stainless steel, 114x114 mm, casting speed is 1.90 m/min
Figure 8.2 Temperature profiles and shell growth of the strand down the top of the mould for austenitic stainless steel, 114x114 mm, casting speed is 2.57m/min
Figure 8.3 Temperature profiles and shell growth of the strand down the top of the mould for martensitic stainless steel, 114x114 mm, casting speed is 1.90m/min
Figure 8.4 Billet shrinkage down the top of the mould for heat 6331, austenitic stainless steel, 114x114 mm, casting speeds are 1.90 m/min and 2.59 m/min
Figure 8.5 Billet shrinkage down the top of the mould for heat 6331 (austenitic), and heat 8115 (martensitic), 114 $\times$ 114 mm, casting speed is 1.90 m/min
Figure 8.6 Mould and billet dimension calculated from mathematical model for austenitic stainless steel, heat 6331, casting speed is 1.90 m/min
Figure 8.7 Mould and billet dimension calculated from mathematical model for austenitic stainless steel, heat 6331, casting speed is 2.57 m/min
Figure 8.8 Mould and billet dimension calculated from mathematical model for martensitic stainless steel, heat 8115, casting speed is 1.90 m/min
Figure 8.9  Mould and billet dimension calculated from mathematical model for martensitic stainless steel, heat 8113, casting speed is 1.90 m/min
Figure 15 Maximum and minimum heat flux profiles for heat 8119, martensitic stainless steel, casting speed is 1.9 m/min, mould powder CT3.
Figure 8.11  Mould and billet dimension calculated from mathematical model for martensitic stainless steel, heat 8119, casting speed is 1.90 m/min
Figure 8.12 Suggested mould taper for heat 6331, austenitic stainless steel, billet size 114x114 mm, casting speed is 1.90m/min
Figure 8.13  Suggested mould taper for heat6331, austenitic stainless steel, billet size 114x114mm, casting speed is 2.57m/min
Figure 8.14  Suggested mould taper for heat 8115, austenitic stainless steel, billet size 114x114 mm, casting speed is 1.90 m/min
Chapter 9  SUMMARY, CONCLUSION AND RECOMMENDATION

Instrumented mould trial was first conducted for stainless steel billet casting at Welland Ontario March 1998. Data on mould temperature have been collected at Atlas Steel. Operating billet mould was instrumented with arrays of thermocouples, and LVDT's. Billets samples were collected and subsequently analyzed for surface quality.

Different models were employed to analyze the collected data. Axial mould heat flux profiles were obtained using INVERSE model. The thermal-mechanical parameters for martensitic and austenitic stainless steel were incorporated into the shrinkage model. Using the calculated heat flux profiles as input, shell thickness and shrinkage of the strand in the mould were calculated and the interaction between and the mould were analyzed. Based on those calculations, the operating mould taper were evaluated and new mould tapers were designed for different operating conditions and steel grades.

Analysis of billet surface quality was carried out. Several mathematical model predictions could be verified by measurements made on billets.

Important conclusions from the work are listed below:
1. Mould heat fluxes were calculated for all heats for all four faces of the mould. It is evident that the peak heat fluxes close to the meniscus were in the 4500 ~ 5900 kW/m\(^2\) range, for casting speed of up to 2.57 m/min, which is significantly higher than previously measured (2500 ~ 3000 kW/m\(^2\)) in an early trial for casting at lower speeds (~1.0 m/min).

2. There were significant differences in the heat flux profiles for the four faces, particularly at the meniscus region. This could be related to differences in mould flux melting behavior and metal level fluctuation.

3. The average heat flux for the mould, computed from the profiles for each face, was highest for the austenitic stainless steel grades (~1900 to 2000 kW/m\(^2\)); the corresponding values were 15% lower for the martensitic grades.

4. Mould heat flux is strongly influenced by casting speeds when casting with mould flux lubrication. An increase in speed from 1.9 to 2.57 m/min, caused the peak mould heat flux to increase from 5176 to 5940 kW/m\(^2\). The heat fluxes are also consistently higher in the lower part of the mould.

5. Peak heat flux for martensitic stainless steel was higher than that for austenitic stainless steel due to the rapid contraction of the austenitic stainless right after the solidification below meniscus. While in the lower portion of the mould, the heat flux for martensitic stainless steel was consistently lower than that of austenitic stainless steel owing to the larger air gap formed for martensitic stainless steel.

6. Shell growth and billet shrinkage profiles were computed for each heat. The shell thickness at the bottom of the mould was 9.02 ~ 10.68 mm for the martensitic
grade and 6.36–8.30 mm for austenitic stainless steel. Although the freezing range for the austenitic stainless steel is similar to the martensitic grade (~50°C), it is cast at significantly higher superheats (68°C) than the martensitic steel grade, which retards shell growth in the mould.

7. Mold taper were calculated for each grade of the cast steel in the plant trial. It is clear that one type of mould taper is not suitable for casting both martensitic and austenitic stainless steel. The taper requirements for martensitic grade were steeper; 2.26%/m: 135~350mm; 1.33%/m 350~780mm, significantly higher than the vales currently in use (1.2%/m for the first 350 mm and 0.5 %/m in the lower half of the mold). The taper requirements for the austenitic stainless steel in the lower portion of the mould should be steeper than the current taper in service. In case of the high superheat in the casting of austenitic stainless steel, the taper around meniscus (~150mm below meniscus) should be shallower, primarily for the compensation of the outward distortion of the mould.

8. Casting speed was also found to have an influence on mould taper requirements. For the austenitic stainless steels, an increase in casting speed from 1.9 to 2.57m/min caused the upper taper to decrease from 1.52 to 1.26%/m. The influence became significant in the lower portion of the mould.

9. The use of non-optimum tapers at Atlas can be associated with a number of the defects observed in the billets. Longitudinal depressions were observed in some of the austenitic stainless billet samples. The lightly binding would cause squeezing of the shell by the mould wall in meniscus region and is doubt to be responsible for buckling and longitudinal depressions.
10. Mould powder behaviors strongly influence the surface quality in the casting of austenitic stainless steel. The frequency of occurrence of surface defects such as depression, lap and breakout is significantly higher with mould powder of low viscosity, especially under high casting speed.

RECOMMENDATIONS FOR FUTURE WORK

It is clear from this research that in continuous casting of stainless steel billet, mould taper together with mould powder controls, to a large extent, the productivity and surface quality of the billet product. Suggested mould taper should be implemented in future production or plant trial. The option of mould powder for high speed casting also need to be further studied.


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