MODELLING OF THE FORMATION OF LONGITUDINAL FACIAL CRACKS IN THE CONTINUOUS CASTING OF STEEL SLABS

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ABSTRACT

Longitudinal facial cracks are a serious quality problem in the continuous casting of steel slabs. Although numerous investigations have been conducted to eliminate this kind of surface defect, and significant improvement has been achieved, there is still a problem especially operating at high casting speeds. Thus, to improve both productivity and quality, additional research is required.

The purpose of this study is to understand the mechanism of formation of longitudinal facial cracks in the continuous casting process of peritectic steels and to propose methods to eliminate the formation of these defects. To achieve this objective, process modeling approach was applied.

Firstly, the delta-to-gamma transformation was modeled numerically assuming carbon diffusion control. The moving boundary (delta/gamma interface) problem was solved by employing a one-dimensional finite-difference method. The result of this calculation shows considerably rapid transformation from delta to gamma due to the high diffusivity of carbon in this temperature range.

Secondly, a heat transfer model of continuous casting of steel was developed and was combined with the phase transformation model. Three heat flux conditions (i.e., low, medium, and high) were obtained from literature data and applied as the thermal boundary condition. Differences in the delta-to-gamma transformation rate were compared for the heat flux conditions investigated. The results of the coupled model indicated that the difference in the heat flux at the meniscus results in large variations of the transformation rate in the meniscus region.
The results of the coupled model were transformed to fictitious temperature by using an steel shrinkage model and adopted to calculate the stresses in the solid shell applying the commercial finite-element program, ABAQUS. Based on the results of the calculations, it was concluded that, in order to generate a longitudinal crack on the solid shell surface, not only the tensile stress caused by rapid transformation (i.e. rapid cooling) but also the presence of hot spots is required.

The threshold values for the retardation of both heat removal at the meniscus and shell growth required to generate longitudinal cracks were obtained; in the present work, the values were approximately 10% and 16%, respectively.

Based on the findings of this study, uniform heat removal in the meniscus region is of utmost important to eliminate the longitudinal cracks. If uniformity in the heat extraction is achieved, even under high heat flux condition, the tensile stress at the shell surface does not exceed the UTS of the shell surface and cracking will not occur.

However, as heat flux increases (i.e., cooling rate increases), the maximum temperature fluctuation permissible before cracking occurs decreases. Thus, ironically, the practical way to eliminate longitudinal cracks when casting at high speeds is to reduce the heat flux in the meniscus region in conjunction with the elimination of non-uniform heat extraction.
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List of Symbols

- $a_\alpha$ - lattice parameter of the alpha phase, Å
- $a_\alpha^0$ - lattice parameter of the alpha phase for pure iron, Å
- $a_\gamma$ - lattice parameter of the gamma phase, Å
- $a_\gamma^0$ - lattice parameter of the gamma phase for pure iron, Å
- $A$ - volume fraction of delta phase, dimensionless
- $C$ - carbon content, wt.%
- $C_0$ - initial carbon concentration in the steel, wt.%
- $C_c$ - molar density of carbon, mole cm$^{-3}$
- $C_{Mn}$ - molar density of manganese, mole cm$^{-3}$
- $C_\gamma$ - carbon concentration in the gamma phase, wt.%
- $C_\delta$ - carbon concentration in the delta phase, wt.%
- $C_\gamma^0$ - equilibrium carbon concentration of gamma phase at peritectic temperature, wt.%
- $C_\delta^0$ - equilibrium carbon concentration of delta phase at peritectic temperature, wt.%
- $C_\gamma|_{\delta-\gamma}$ - carbon concentration of the gamma phase at the delta-gamma interface, wt.%
- $C_\delta|_{\delta-\gamma}$ - carbon concentration of delta phase at the delta-gamma interface, wt.%
- $C_\gamma^T$ - equilibrium carbon concentration of gamma phase at temperature $T$, wt.%
- $C_\gamma^*$ - carbon concentration of gamma phase at the delta-gamma interface at former time step, wt.%
$C_8^*$ - carbon concentration of delta phase at the delta-gamma interface at former time step, wt.%

$C_{peri}$ - carbon equivalent, wt.%

$C_{Pa}$ - apparent specific heat capacity, J g$^{-1}$ K$^{-1}$

$C_p$ - heat capacity of steel, J g$^{-1}$ K$^{-1}$

$C_{p_{eff}}$ - effective heat capacity (which takes into account of the latent heat), J g$^{-1}$ K$^{-1}$

$CR$ - cooling rate, °C s$^{-1}$

d - primary dendrite arm spacing (size of the grain), µm

$D_\gamma$ - diffusion coefficient of carbon in the gamma phase, cm$^2$ s$^{-1}$

$D_{ce}$ - diffusion coefficient of carbon in iron, cm$^2$ s$^{-1}$

$D_{c,Mn}$ - diffusion coefficient of carbon in iron under the presence of manganese, cm$^2$ s$^{-1}$

$E$ - Young’s modulus, GPa

$\Delta H$ - latent heat of fusion for steel, J g$^{-1}$

$J_c$ - molar flux of carbon, mole cm$^2$ s$^{-1}$

$k$ - thermal conductivity of steel, W m$^{-1}$ K$^{-1}$

$k_i$ - coefficient of shifting peritectic point of alloying element $i$, dimensionless

$L$ - length of the steel at temperature $T$, mm

$L_0$ - length of the steel at solidus temperature, mm

$M_i$ - number of atoms per unit cell, atoms

$N_a$ - Avogadro’s number, g-atom mole$^{-1}$

$q(t)$ - heat flux, kW m$^{-2}$
Q - extracted heat within 1 second from meniscus per unit area for normal position, kJ m⁻²

Qₐ - activation energy for carbon diffusion, cal mole⁻¹

Qₙₐ - extracted heat within 1 second from meniscus per unit area for hot spot, kJ m⁻²

r - thermal resistance, °C m² W⁻¹

R - gas constant, cal K⁻¹ mole⁻¹

Rₚ - volume fraction of gamma phase, dimensionless

Rₚ⁰ - initial volume fraction of the gamma phase, dimensionless

S - solid shell thickness for normal position, mm

Sₙₚ - solid shell thickness for hot spot, mm

t - time, s

Δt - time step, s

tₜ - transit time, s

T - temperature, °C

T₀ - temperature corresponding to a strain-free state, °C

Tₐvg - average temperature of the solid shell, °C

Tᵢ - initial temperature of steel, °C

Tₗiq - liquidus temperature, °C

Tₗ - solidus temperature, °C

T₈urₖ - temperature of the shell surface, °C

T* - fictitious temperature for ABAQUS calculation (derived from Eq. 6.4), °C

u - withdraw speed, m min⁻¹
$v_i$ - displacement of node i in y direction, mm

$V$ - velocity of the delta-gamma interface toward the delta phase, mm s$^{-1}$

$V_c$ - casting velocity, m min$^{-1}$

$V_{\text{delta}}$ - velocity of the delta-gamma interface due to the carbon diffusion in the delta phase, mm/s

$V_\delta$ - specific volume of delta iron, cm$^3$ g$^{-1}$

$V_\gamma$ - specific volume of gamma iron, cm$^3$ g$^{-1}$

$W$ - slab width, mm

$W_c$ - carbon content of the phase, wt.%

$W_{Fe}$ - atomic weight of iron, g mole$^{-1}$

$x$ - distance in x-direction, mm

$X_c$ - carbon content of the phase, at.%

$X_i$ - concentration of alloying element i, wt.%

$X_s$ - slab thickness, mm

$y$ - distance in y-direction, mm

$z$ - distance in z-direction, mm

$\alpha$ - thermal expansion coefficient, K$^{-1}$

$\varepsilon$ - total strain, dimensionless

$\varepsilon^{th}$ - thermal strain, dimensionless

$\varepsilon^{tr}$ - strain due to transformation, dimensionless

$\rho$ - density of steel, g cm$^{-3}$
$\rho_\delta$ - density of delta iron, g cm\(^3\)

$\rho_\gamma$ - density of gamma iron, g cm\(^3\)

$\sigma_{\text{max}}$ - maximum tensile stress at solid shell surface, MPa

$\sigma_{\text{UTS}}^{\text{low}}$ - UTS of the low carbon steel, MPa

$\sigma_{\text{UTS}}^{0.4}$ - UTS of the 0.4% carbon steel, MPa

$\sigma_{\text{UTS}}$ - UTS of the steel, MPa
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Chapter 1. Introduction

The continuous casting of steel, schematically shown in Fig. 1.1, has been widely adopted by steel industry for the last few decades because of its inherent advantages of low cost, high yield, and ability to achieve a high quality cast product. The process is capable of extracting heat at a remarkably high rate with the combination of mould, sprays, and radiant cooling. The rapid cooling, however, results in steep temperature gradients in the solid shell that can change rapidly and generate thermal strains as the shell

Fig. 1.1 Schematic diagram of the mould region of continuous casting process.
expands or contracts. Under particularly severe conditions, a number of defects can be formed and, therefore, crack formation has long been recognized as a problem in the continuous casting of steel.

Longitudinal, midface cracking is a particularly serious problem in the casting process. Since it is exposed to air, the crack surface oxidizes and, therefore, does not reweld during hot rolling. If the cracks are minor, they can be removed by scarifying, but if they are long or deep cracks, the slab has to be scrapped. Therefore, in order to improve the quality and production yield, it is very important to eliminate this kind of surface defect.

Recently, hot charging, which saves energy and is an environmental-friendly technique, has been promoted in the steel making industry [1, 2]. However, the applicability of hot charging is limited by the occurrence of defects such as longitudinal facial cracks.

High speed casting, which also increases the productivity of continuous casting machines, is a key technology in hot charging because the slabs have to be kept at high temperatures. However, when operating with higher casting speeds, surface defects in slabs tend to occur as a result of rapid cooling and large temperature gradients. Therefore, although higher casting speeds are sought to increase the productivity of the casting machine, there are technical barriers that prevent the use of high casting speeds.

Nowadays, some minimills that have introduced the thin slab casting process are operating at very high casting speed (4.5~6.0 m/min) [3], and it seems that a breakthrough has been achieved. However, even though the thin slab casting process is commercially
successful, there are a number of restrictions on the steel grades that can be cast such as peritectic steels which cannot be cast because of the surface defects at the present time [3].

Thus, although numerous investigation have been done in this field, and quite a few improvements have been accomplished to decrease longitudinal cracks, there are a number of issues which still remain unresolved and require additional research.

In this study, the concept of process modelling has been applied to study the formation of longitudinal cracks during continuous casting at high casting speeds and to propose methods to eliminate the formation of these defects. The approach taken involved the development and application of mathematical models of phase transformation, heat transfer, steel shrinkage and stress generation.
Chapter 2. Literature Review

In this chapter, the literature on the longitudinal cracks in continuous casting of steel slabs is reviewed. First, crack morphology and the effect of casting conditions on longitudinal cracks are presented. Then, material related to the high temperature mechanical properties of steel is reviewed. The final section is dedicated to a discussion of published mechanism of formation of longitudinal cracks.

2.1 Crack Morphology

Longitudinal facial cracks in continuously cast steel slabs (a typical example is shown in Fig. 2.1 [4]) are a common surface defect especially when operating at a high casting speed. Being exposed to air, the crack surface oxidizes and, therefore, does not reweld during hot rolling. If the cracks are short or shallow, they can be removed by scarifying but if they are long and deep, the slab has to be scrapped. Thus, in order to improve the quality and production yield, it is very important to eliminate this kind of surface defect.

In many cases, longitudinal cracks are accompanied by depressions [5, 6]. The cracks are inter-dendritic [5, 6] and, in some conditions, local segregation [5, 6] or the mould flux [5] were found in the cracks. These facts indicate that longitudinal cracks begin to form in the mould at the solidification front where a liquid film separates individual dendrites.
Fig 2.1 Typical longitudinal face crack on a continuously cast slab [4].

Furthermore, Saeki et al. [7] cast steel slabs using a mould which had an artificial groove. They changed the meniscus position from Level I (40mm below the top of the groove) to Level III (20mm above from the top of the groove) during casting as shown in Fig. 2.2. They found that when the top of the groove was 20 mm below the meniscus (Level III) the longitudinal crack was not observed. This can be seen in Fig. 2.3. In this figure, the total length of cracks became almost zero at the casting length between 30 and 45m. The results of their experiment indicate that retardation of solidification that occurs within 20 mm (equal to 1 second under their experimental condition) from the meniscus
caused the longitudinal cracks. This experiment also supports the fact that longitudinal cracks form during the early stage of solidification.

![Fig. 2.2 Scheduled relative position of liquid meniscus to the artificial mould groove [7]. (TC: Thermocouple)](image)

![Fig. 2.3 Crack formation showing its dependence on the relative position of liquid meniscus to the groove [7].](image)
2.2 The Effect of Casting Conditions

2.2.1 The Effect of Carbon Content

The effect of casting conditions on longitudinal facial cracks has been reported by many researchers [8-28]. It has been long recognized that the carbon content has one of the strongest influences on the frequency and severity of cracking. As can be seen in Fig. 2.4 (a), steels with carbon contents around 0.11% are particularly sensitive to crack formation [8]. The effect of Mn/S ratio (Fig. 2.4 (b)) and sulfur content (Fig. 2.4 (c)) will be discussed in section 2.2.2.

![Diagram showing the effect of chemical factors on mid-face longitudinal cracking](image-url)

Fig. 2.4 Effect of chemical factors on mid-face longitudinal cracking [8].
Saeki et al. [7] also indicated the influence of the carbon content. According to their research, steels containing approximately 0.14% carbon are the most susceptible to longitudinal facial cracks. Nakai et al. [9] mentioned that steels containing 0.10 to 0.15 percent carbon are sensitive to the occurrence of surface longitudinal cracks. Vereecke et al. [10] suggested that the maximum crack probability was observed around 0.10% carbon. Further, some minimills which are applying thin slab casting cannot cast steel grades in the range of 0.065 to 0.15 % carbon because of the surface cracks [3].

Matsumiya et al. [11] commented that the reason for the variation of this critical carbon content may be caused by the influence of the other elements, such as Mn (Mn will change the peritectic reaction range) and the operating conditions or casting machine characteristics but details of this issue are still unknown. However, at least two points of information are evident from the previous works as follows.

1. Although there are slight differences, the critical carbon range is approximately 0.10 to 0.15 % for conventional slab casting; this corresponds to the hypo-peritectic carbon range (0.09 to 0.16%) in the Fe-C phase diagram [12] as illustrated in Fig 2.5.
2. In the case of thin slab casting, the carbon range does not deviate so much but the range itself becomes wider on the lower carbon side.

The reason why this hypo-peritectic carbon grade steels are so susceptible to longitudinal cracks will be discussed in a later section.
2.2.2 The Effect of Chemical Composition of Other Solutes

As already shown in Fig. 2.4 (b) and (c), the sulfur content and Mn/S ratio have an influence on longitudinal facial cracks. The cracking problem becomes more serious with decreasing Mn/S ratio and increasing sulfur contents. Similarly Fogleman and Orie [13] reported that increasing sulfur content will result in an increase of the occurrence of longitudinal cracks. Vereecke et al. [10] examined the influence of Mn/S ratio and verified that Mn/S was the most relevant chemical factor. They said that, at fixed Mn/S
ratio, Mn itself had no influence on the frequency of longitudinal cracks up to Mn/S ratios of about 40.

A similar effect of the manganese and sulfur content on the susceptibility to internal midway cracks has been reported [14]. Therefore, the effect of these elements on longitudinal cracks seems to be associated with the mechanical properties of the steels rather than the peritectic reaction itself. However, these elements also have an effect on the peritectic reaction and several researchers have investigated the effects of alloying elements on the change in the peritectic range.

With regard to the peritectic point, Ishida et al. [15] and Yasumoto et al. [16] have obtained the coefficient of each alloying element on shifting the peritectic point of Fe-C system. Using these coefficients, the carbon equivalent can be expressed:

\[ C_{\text{per}} = C + \sum k_i \cdot X_i \] (2.1)

where \( C_{\text{per}} \) is the carbon equivalent and \( C \) is the carbon concentration of the steel in wt.%. \( X_i \) is the concentration of each alloying element in wt.% and \( k_i \) is the coefficient of element \( i \) which are given in Table 2.1. for each element.

<table>
<thead>
<tr>
<th>Alloying element, X</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>Method</th>
<th>Remarks</th>
</tr>
</thead>
<tbody>
<tr>
<td>k_i Ishida et al. [15]</td>
<td>0.015</td>
<td>0</td>
<td>2.9</td>
<td>Thermo-dynamics</td>
<td>Equilibrium</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>0.02</td>
<td>-0.01</td>
<td>0.67</td>
<td>Thermal analysis</td>
<td>Non-equilibrium</td>
</tr>
</tbody>
</table>
Their results are in good agreement for all the elements except sulfur. In this table, a positive value of \( k \) signifies that the element shifts the peritectic point toward the lower carbon side.

Schmidtmann and Pleugel [17] obtained the phase diagram of Fe-C which contains 1.6 wt.% Mn and reported that the peritectic point is approximately 0.13% carbon. Using the peritectic carbon concentration in Fig. 2.5 (i.e. 0.16 wt.%), the peritectic carbon concentration for the 1.6 wt.% Mn steel should be 0.128 to 0.136 wt.% according to the coefficient in Table 2.1. This range corresponds to Schmidtmann and Pleugel's result.

### 2.2.3 The Effect of Casting Speed

The effect of casting speed is also one of the biggest issues. In general, the frequency of the longitudinal cracks increases with casting speed [8,18]. Table 2.2 shows the influence of the casting speed on longitudinal cracks as determined by Irving and Perkins [8]. They investigated the surface quality of 1830×180mm slab of 0.12-0.14% carbon steel.

Table 2.2 The influence of casting speed on surface quality [8]

<table>
<thead>
<tr>
<th>Casting speed m/min.</th>
<th>Tonnes cast</th>
<th>Slab with good surface %</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.8</td>
<td>660</td>
<td>100</td>
</tr>
<tr>
<td>0.8-0.9</td>
<td>8190</td>
<td>75</td>
</tr>
<tr>
<td>0.9-1.0</td>
<td>52530</td>
<td>69</td>
</tr>
<tr>
<td>1.0-1.1</td>
<td>20130</td>
<td>69</td>
</tr>
<tr>
<td>1.1-1.2</td>
<td>12960</td>
<td>67</td>
</tr>
<tr>
<td>1.2-1.3</td>
<td>1980</td>
<td>65</td>
</tr>
</tbody>
</table>
Clearly, the percentage of slabs with good surface quality decreases with increasing casting speed. In fact, as already mentioned above, some minimills which are employing the thin slab casting process with high casting speed (=4.5~6.0m/min.) cannot cast steel grades in the range of 0.065 to 0.15% carbon because of an unacceptable high frequency of surface defects [3].

The increase in the longitudinal cracks as casting speed increases might be related to the decrease of mould flux consumption. Wada et al. [1] showed that increasing the casting speed results in a decrease of mould flux consumption. Similar findings have been reported by other researchers [19, 20].

The decrease in mould flux consumption is associated with a decrease in the average thickness of the mould flux between mould and strand. Although the heat transfer in the mould is very complicated and related to many parameters, in general, this decrease in the thickness of mould flux results in an increase in the heat flux in the mould. In fact, some researchers [21, 22, 23, 24] have shown that the heat flux at the meniscus in the mould increases as the casting speed increases. A typical example of this can be seen in Fig. 2.6.

The reason for the increase in longitudinal facial cracks with increasing heat flux at the meniscus will be discussed later.
2.2.4 The Effect of Other Casting Conditions

Slab size also seems to have an effect on the frequency of longitudinal cracking. Larsen and Moss [4] suggested that wider slabs, i.e. above approximately 1220 mm width, are more susceptible to crack formation than narrower slabs. Vereecke et al. [10] explained that the influence of slab width is due to the high linear shrinkage of the solidifying shell along the wide mould faces. Sardemann and Schrewe [25] indicated that not only the width itself but also the thickness-to-width ratio is a very important factor. According to their findings, slabs of widths > 1500 mm or with a thickness-to-width ratio >1:6 were originally considered not to be castable free of longitudinal cracks. Similar behavior can be seen in Fig. 2.4 [8]. This figure suggests that thin slabs are more sensitive.
to longitudinal cracks than thick slabs. However, Irving and Perkins have cautioned against reaching that conclusion because thinner slabs are normally cast at a higher speed.

Mould flux properties such as viscosity, crystallization temperature, etc. have also a strong influence on the surface quality of steel slabs. The viscosity of the mould flux is strongly related to the casting speed [2, 18]. Thus, it is very important to design the mould flux according to the casting speed [26]. On the other hand, mould flux which has high crystallization temperature is one of the most successful countermeasures for longitudinal cracking [25, 27]. The reduction of heat transfer as a result of high crystallization temperature [28] is believed to be responsible for this decrease in the frequency of longitudinal cracks.

Although the origin of longitudinal cracks is in the mould, the effect of spray cooling, especially top zone spray cooling, has a very strong effect on the severity of the longitudinal facial cracks [4, 8, 13, 29]. This behavior can be explained by the fact that rapid spray cooling generates large tensile strains in the slab surface. Therefore, the subsurface cracks, which did not penetrate to the surface in the mould, may do so in the top spray zone. On the other hand, too little upper spray water also increases the occurrence of longitudinal cracks [8] because of bulging of the slab.

2.3 High Temperature Mechanical Properties of Steel

The nature of the stress and strains which cause cracks in the solidifying shell in the continuous casting process has been discussed by many researchers. In general, the mechanical properties of steel at elevated temperature are affected by several variables,
such as, temperature, composition, microstructure, strain rate and thermal history. Lankford [30] investigated the strength and ductility of steel at high temperature. He showed that not only in the temperature range of 800 to 1200°C but also above 1450°C, steel becomes quite brittle. The latter temperature range is called “the zero ductility temperature range” which extends to approximately 30-70°C below the solidus temperature as reviewed, for instance, by Thomas et al [31].

Interestingly, the strength of steel steadily decreases with increasing temperature and is not affected in the same manner as ductility. The temperature dependence of the strength of different steel grade was reviewed by Brimacombe and Sorimachi [32]. Recently, Shin et al. [33] examined the mechanical properties of steels. They measured the ultimate tensile strength (UTS) of Fe-C-1wt.% Mn samples at different temperatures and various carbon contents in the range of 0.06 to 0.60 wt.% carbon. They found that the UTS decreases linearly with increasing temperature and the gradient of this relation is independent of carbon content as illustrated in Fig. 2.7.

Fig. 2.7 Change of the strength of steels with temperature at various carbon contents[33].
On the other hand, the ductility-to-fracture strain of steel just below the solidification temperature appears to be of the order of 0.2 to 0.3 pct. according to Vom Ende and Vogt [34]. The elastic modulus derived by Mizukami et al. [35] under continuous casting conditions is expressed as

$$E = 968 - 2.33 T + 1.90 \times 10^{-3} T^2 - 5.18 \times 10^{-7} T^3$$  \hspace{1cm} (2.2)

where $E$ is the Young's modulus in GPa and $T$ is the temperature in °C. By this equation, the elastic modulus about 50 °C below the solidification temperature appears to be of the order of ~5000MPa. Using this elastic modulus and Vom Ende and Vogt's data, the UTS of the steel around this temperature should be of the order of 10 MPa assuming elastic behavior of steel. This corresponds to the results of Shin et al. as can be seen in Fig. 2.7. Therefore, in this temperature range, steel seems to be extremely brittle and as soon as it reaches the UTS it fractures without showing plastic deformation.

This remarkable decline of strength and ductility at high temperature makes the problem of crack formation of the steel shell at the meniscus more serious.

2.4 The Mechanism of Formation of Longitudinal Facial Cracks

2.4.1 Peritectic Transformation

As already mentioned above, the carbon content has a strong influence on the frequency and severity of longitudinal cracks. In the Fe-C phase-diagram shown in Fig. 2.5, it is clear that around 0.10 to 0.15% carbon, steel undergoes a peritectic
transformation. Therefore, it is believed that the influence of carbon content on the occurrence of longitudinal cracks is related to this peritectic reaction.

Research on the peritectic reaction in Fe-C alloys has been conducted by several investigators. Chuang et al. [36] concluded from studies in a 0.39% carbon steel that the delta-to-gamma transformation is controlled by carbon diffusion in the gamma phase. Further, the peritectic reaction occurs at relatively low undercooling from the equilibrium condition. For instance, the peritectic reaction was completed in 2 to 4°C below the peritectic temperature and solidification was ended 2.5 to 10°C below the solidus temperature.

Fredriksson and Stejerndahl [37] also showed that the peritectic reaction is controlled by carbon diffusion in the gamma phase. They calculated the temperature range of peritectic reaction and concluded that the peritectic reaction was quite fast and was finished at a maximum of 6 or 10K below the peritectic temperature. Suzuki et al. [38] also observed this peritectic reaction and found the process by which the gamma phase wraps the delta phase and grows toward the delta phase.

On the other hand, Takahashi et al. [39] presented a different mechanism. They investigated the distribution of the carbon concentration in a quenched sample of Fe-0.29%C alloy and found that the low carbon spots which corresponded to the carbon content of the delta phase remained untransformed. However, this result has to be viewed with care for the following reasons:
1) The experimental method employed by Takahashi et al. was based on thermal analysis for conditions where the peritectic reaction could be undercooled due to the absence of gamma nuclei as noted earlier by Chuang et al. [36].

2) The second reason is that they compared the carbon concentration with the equilibrium carbon concentration of the delta phase at peritectic temperature. They regarded the 0.11% carbon region as delta phase, although the equilibrium carbon concentration of delta phase at this temperature is approximately 0.04%.

Recently, by employing a diffusion couple, Matsuura et al. [40, 41, 42, 43] clearly proved that the delta-to-gamma transformation is controlled by carbon diffusion in the gamma phase. They also reported a significantly rapid transformation. Interestingly, they observed that the rate of the peritectic reaction increases with decreasing temperature because of the increase in the carbon concentration gradient in the gamma phase.

In conclusion, under normal casting conditions such that the gamma phase can nucleate easily on the pre-existing delta phase, the peritectic reaction in Fe-C alloy seems to be controlled by carbon diffusion in the gamma phase. Furthermore, because of the high mobility of carbon in this temperature range, most of the researchers have observed significantly fast transformations which are close to equilibrium.

However these investigations have been conducted in hyper-peritectic (i.e. 0.16<C<0.53wt.% ) grades where liquid phase is present during the transformation. Hence, delta-to-gamma transformation in hypo-peritectic (0.09<C<0.16wt.%) steel has to be examined in order to evaluate the transformation rate for the steels in the carbon range of interest.
2.4.2 Steel Shrinkage due to the Delta-to-Gamma Transformation

The structural change from B.C.C. to F.C.C. during the delta to gamma transformation is associated with a considerable shrinkage which might be essential in evaluating the susceptibility to longitudinal facial cracks during continuous casting. Therefore, in order to understand the mechanism of longitudinal crack generation, it is necessary to know the steel shell shrinkage because of the delta-to-gamma transformation.

There has been limited work on the direct measurement of steel shrinkage associated with the delta-to-gamma transformation because of the experimental difficulties. Therefore, most of the research that has been done in this field has involved numerical predictions.

In order to predict the thermal expansion/contraction in a two phase region of steel, it is necessary to know the temperature and composition dependence of the density of the two phases. Once these densities are derived, the density and contraction of steel in the two phase region can be predicted by applying the following equation [11]:

\[
\rho = \frac{1}{\left\{ \frac{A}{\rho_6} + \frac{(1-A)}{\rho_\gamma} \right\}}
\]

(2.3)

where \(A\) is the volume fraction of delta phase, and \(\rho_6\) and \(\rho_\gamma\) are the density of delta and gamma iron respectively. In order to derive the volume fraction of delta phase, \(A\), most of the researchers have used lever rule and phase diagram. This is considered to be reasonable because the mobility of carbon in this temperature range is extremely fast and the transformation is close to equilibrium, as already mentioned in the previous section.
On the other hand, \( \rho_\delta \) and \( \rho_\gamma \) are function of temperature and carbon concentration. For the temperature dependence of the specific volume of pure delta and gamma iron, Wray [44] modified Lucas's [45] data and derived the following equations:

\[
V_\delta = 0.1234 + 9.38 \times 10^{-6}(T - 20) \tag{2.4}
\]

\[
V_\gamma = 0.1225 + 9.45 \times 10^{-6}(T - 20) \tag{2.5}
\]

where \( T \) is the temperature in °C and \( V_\delta \) and \( V_\gamma \) are the specific volume of delta and gamma iron respectively in cm\(^3\)/g.

Instead of measuring specific volume or density, the lattice parameter has been measured to examine the influence of carbon concentration. Fasiska and Wagenblast [46], in an earlier work, measured the lattice parameter of the alpha phase for different carbon contents at room temperature and obtained:

\[
a_\alpha = a_\alpha^0 + 8.40 \times 10^{-3}X_c \tag{2.6}
\]

where \( a_\alpha \) is the lattice parameter of the alpha phase in Å and \( a_\alpha^0 \) is the lattice parameter of the alpha phase for pure iron in Å. \( X_c \) is the carbon content of the phase in atomic percentage. This equation was used for delta-iron by Chandra et al. [47] based on the fact that both have the same B.C.C. structure.

In the case of gamma iron, Chandra et al. [47] used the method of least squares to fit curves to the data derived by Ridley and Stuart [48], and predicted the lattice parameters of gamma phase for a given carbon content and temperature as follows:

\[
a_\gamma = a_\gamma^0 + \left(0.0317 - 11.65 \times 10^{-7}T - 0.05 \times 10^{-7}T^2\right)W_c \tag{2.7}
\]
where \( T \) is the temperature in °C, \( a_\gamma \) is the lattice parameter of the gamma phase in Å and \( a_\gamma^0 \) is the lattice parameter of the gamma phase for pure iron in Å. \( W_c \) is the carbon content of the phase in weight percent.

Using these equations, Chandra et al. calculated the linear expansion coefficient of steel for different carbon contents. They found that a steel with a carbon content of 0.15%, which undergoes peritectic phase transformation, causes shrinkage due to both thermal contraction and phase transformation and the latter is much greater in magnitude (i.e., the shrinkage due to the phase transformation is about 3~4 times greater than that of the thermal contraction as will be discussed in chapter 6). Moreover, in this hypo-peritectic steel grade, total contraction in the early stage of cooling is dominated by the effect of phase transformation. On the other hand, the linear expansion coefficient for those grades that do not undergo phase transformation over the temperature range of interest is nearly constant with temperature.

### 2.4.3 Heat Transfer in the Mould

The first comprehensive investigation of the relationship between heat-flow rates in the mould and the carbon content of the steel being cast was conducted by Brimacombe and Weinberg [49]. They observed that the solid shell becomes thinner and more nonuniform in the trial with 0.1% carbon steel than that of in the high carbon steel trials. Employing bench-scale experiments, Singh and Blazek [50] also found that the mould heat flux reaches a minimum for around 0.1% carbon steel as shown in Fig. 2.8.
In this experiment, they also observed the shells and found a lot of wrinkles in 0.1% carbon steel. They concluded that the reason for a minimum in heat transfer is related to the uneven shell thickness and explained that this uneven shell-growth arises from the high shrinkage rate of this carbon range but they did not explain the detailed mechanism, such as the delta-to-gamma transformation.

Sugitani et al. [51] also reported that, in carbon steel, the unevenness of the solidified shell has its maximum value at 0.11-0.13% carbon and decreases rapidly when the carbon content deviates slightly away from this range.

The precise role that this shrinkage may play in altering the formation on the solid shell and thereby influencing heat flow in the mould, cannot be visualized easily because of the complex interaction of heat transfer and deformation of the solid shell at high,
temperatures. Nevertheless, Grill and Brimacombe [52] constructed a simplified mechanism that sheds some light on these phenomena and explains the observed behavior as illustrated in Fig. 2.9.

![Schematic representation of steel shell deformation in the mould as a result of the delta-to-gamma transformation [52].](image)

1) At a given instant in time, the newly solidified shell below the meniscus is considered to be in contact with the mould. During this period, the surface of the shell cools rapidly, about 100°C/s, and transforms from delta to the gamma phase.

2) The surface transforms and shrinks, while the inner region of the shell, being above the peritectic temperature (1493°C), remains unchanged. This results in an inward bending of the shell. If the surface of the shell has sufficient strength to withstand the
small ferrostatic pressure at this level, a gap forms between the surface and the mold wall.

3) As the shell descends through the mould, the surface of the steel in the newly formed gap region begins to reheat owing to the decrease in heat transfer across the gap. The reheating, in turn, leads to a lowering of the strength of the steel, such that increasing ferrostatic pressure can push the shell in the gap region back toward the mould wall. The resulting deformation may form wrinkles or indentations on the surface of the shell.

4) The solid shell in the gap region will be thinner than that of surrounding regions due to the reduction in heat flow across the gap.

5) The sequence described above repeats itself continuously, leaving wrinkles along the surface and fluctuations in the thickness of the solid shell, as reported by Singh and Blazek [50].

2.4.4 Uneven Shell Growth and Critical Heat Flux

Shrinkage in the solidified steel shell, especially that associated with the delta-to-gamma transformation, will result in gap formation between the mould and the strand. This gap will affect the heat transfer in the mould and results in the formation of wrinkles along the surface of the slab. Therefore it is important to relate this phenomenon to longitudinal crack formation.

As already mentioned, Saeki et al. [7] conducted an experiment using a mould with an artificial groove. They found that the slabs which had cracks exhibited a retardation in
the shell growth. They mentioned that the critical retardation ratio of shell growth for the longitudinal cracks was around 10%. They also suggested that the cooling rate at the crack region is lower than in the ordinary region (without cracks) because secondary arm spacing of dendrites in the crack region is larger than in the ordinary region. Therefore, it is essential to reduce the unevenness of the shell in order to eliminate the longitudinal facial cracks.

Sugitani et al. suggested that, to prevent the uneven shell growth, soft cooling in the mould is effective. They reported [53] that Al-killed steel containing about 0.12% carbon, whose shell is formed quite unevenly in a water-cooled copper plate mould, has been observed to solidify evenly under some low rate of heat removal. This critical rate was about 0.93 MW/m². This result corresponds to the fact that higher casting speed, which leads to the higher heat flux at meniscus, results in an increase in the longitudinal facial cracks.

Hiraki et al. [23] noted that there is a critical heat flux to prevent longitudinal cracks. They suggested that the critical heat flux depends on the carbon contents and that the medium (0.10-0.12%C) carbon steels are more sensitive to the heat flux as shown in Fig. 2.10.

Recently, Sugitani et al. [9, 54] changed the roughness of the mould inner surface in order to find the optimum conditions for reducing local heat flux density in the mould. As a result, they found that very small longitudinal grooves on the mould inner surface is an effective way to achieve stable casting and heat transfer. As already mentioned, a
mould flux with a high crystallization temperature will also play the same role on this aspect.

Fig. 2.10 The effect of the heat flux in the mould on the longitudinal cracking [23]

Nakato et al. [55] measured the temperature in the copper mould during casting. They found that the copper plate temperature increases, and the width-wise distribution of it in the vicinity of the meniscus, becomes uneven with increasing casting speed. Although the temperature measurements were carried out during steady state casting, the temperature distribution is not constant but changes occasionally with time.

In summary, the higher the heat flux, the more uneven the steel shell becomes. There seems to be a critical heat flux to prevent the steel shell from this uneven growth. Beyond this critical heat flux, the steel wrinkles and retardation of the shell growth, which
may cause longitudinal cracks, will increase. This phenomenon is strongly related to the carbon concentration.
Chapter 3. Scope and Objectives

From the previous work it appears that the effect of the casting conditions such as casting speed, mould flux properties, carbon concentration in the steel, etc. on longitudinal cracks is reasonably well understood. Importantly, the literature indicates that the meniscus is the most critical region for the generation of longitudinal cracks. However, due to the difficulty associated with high-temperature measurements, there is limited information in the literature that addresses the kinetics of the delta-to-gamma transformation; and therefore, the mechanism of the formation of longitudinal cracks still remains unclear.

The overall goal of the present study was to apply the process modelling approach to generate a basis for understanding the formation of longitudinal facial cracks in continuous casting of steel slabs. To achieve this objective, mathematical models of the delta-to-gamma transformation, heat transfer in the mould and steel shrinkage were developed. The results of the models were used to calculate the stress distribution in the solid shell applying the commercial finite-element program, ABAQUS. The primary objectives of the present work were:

(1) To explain the generation of longitudinal cracks in steel slabs on the basis of variability in heat transfer in the meniscus region;

(2) To clarify the effect of mould heat transfer in the meniscus region;

(3) To propose a method for the elimination of longitudinal facial cracks

To accomplish this, the following methodology was followed:
(i) a delta-to-gamma transformation model was developed assuming carbon diffusion control;

(ii) heat transfer in the continuous casting mould was modeled and coupled with the phase transformation model;

(iii) the results of the models described above, together with a steel shrinkage model, were input to a finite-element model of elastic stress generation in the solid shell.
Chapter 4. Mathematical Modelling of the Delta to Gamma Transformation in an Fe-C Alloy

From the literature review (Chapter 2), many investigations on the peritectic transformation have been conducted in Fe-C alloys. The transformation appears to be controlled by carbon diffusion in the gamma phase. However, these studies were conducted in hyper-peritectic carbon grades. In this study, a mathematical model was applied to evaluate the transformation rate of hypo-peritectic grade steels. The details of the model are described in this chapter.

4.1 Model Assumptions and Mathematical Formulation

4.1.1 Assumptions

A schematic diagram of the concept of this model is shown in Fig.4.1. In the figure, the peritectic reaction occurs at the bottom of the primary dendrite arms. At this point delta iron and the remaining liquid iron react to become gamma iron. Therefore, at this moment, delta iron is surrounded by the gamma iron. After the peritectic reaction, the carbon both in the gamma and delta phase diffuses toward the delta/gamma interface and the gamma phase grows toward the inside of the grain.

The diffusion is considered to be one-dimensional. The following assumptions were adopted in the formulation of mathematical model:
Fig 4.1 Schematic diagram of the concept for the hypo-peritectic transformation.
(i) The nucleation of the gamma phase is instantaneous and undercooling from the peritectic temperature is negligible. This is consistent with the observations [36,37] of very small undercoolings from the equilibrium temperature for the peritectic reaction because of the high mobility of carbon atoms in this temperature range. Therefore, equilibrium conditions are assumed at the peritectic temperature.

(ii) The domain of calculation is each grain the size of which corresponds to the primary dendrite arm spacing. A planar form has been used to describe this dendrite arm spacing in order to simplify the model.

(iii) There is no mass transfer across the grain boundary because symmetry is assumed.

(iv) The growth of the gamma phase is controlled by carbon diffusion.

(v) The carbon concentration inside the delta phase is uniform because of an extremely high carbon diffusivity.

(vi) The equilibrium carbon concentrations in delta ferrite and austenite are attained instantaneously at the delta/gamma interface (local equilibrium).

4.1.2 Mathematical Formulation

One-dimensional carbon diffusion in the gamma phase is expressed by Fick’s second law:

\[
\frac{\partial C_\gamma}{\partial t} = D_\gamma \frac{\partial^2 C_\gamma}{\partial x^2}
\]  

(4.1)

where \(C_\gamma\) is the carbon concentration in the gamma phase, \(x\) is the distance, \(t\) is time and \(D_\gamma\) is the diffusion coefficient of carbon in the gamma phase.
According to assumption (i), the lever rule can be used to determine the initial volume fraction of the gamma phase, $R_\gamma^0$, from the phase diagram, i.e.

$$R_\gamma^0 = \frac{C_0 - C_8^0}{C_\gamma^0 - C_\delta^0}$$

(4.2)

where $C_0$ is the initial carbon concentration, $C_\gamma^0$ and $C_\delta^0$ are the equilibrium carbon concentrations at the peritectic temperature of the gamma and delta phases respectively.

There are two boundaries for the gamma phase. One is the outside surface of the grain and the other is the delta/gamma interface. The first boundary condition of this model is expressed assuming simple symmetry with no mass transfer across the grain boundaries, i.e.

$$\frac{\partial C_\gamma}{\partial x}\bigg|_{\text{surface}} = 0$$

(4.3)

For the second boundary condition, it is assumed that the equilibrium carbon concentrations in the delta and gamma phases are attained instantaneously at the delta/gamma interface. Assuming this local equilibrium condition, the boundary condition is:

$$C_\gamma\big|_{\delta-\gamma} = C_\gamma^T$$

(4.4)

where $C_\gamma\big|_{\delta-\gamma}$ is the carbon concentration of the gamma phase at the delta/gamma interface and $C_\gamma^T$ is the equilibrium carbon concentration of the gamma phase at temperature $T$ which will be derived from the equilibrium phase diagram.
In the present work, the fluxes both from the delta and the gamma phases contribute to the movement of the delta/gamma interface. Fig. 4.2 shows the schematic diagram of this mass balance at the delta/gamma interface.

Using Fick's first law, the velocity of the moving delta/gamma interface during the growth of the gamma phase is expressed as

\[
V = -\frac{D_\gamma}{(C_\gamma|_{\delta-\gamma} - C_\delta|_{\delta-\gamma})} \frac{\partial C_\gamma}{\partial x}_{\text{interface}} + V_{\text{delta}} \tag{4.5}
\]

where \( V \) is the velocity of the delta/gamma interface toward the delta phase, and \( C_\delta|_{\delta-\gamma} \) and \( C_\gamma|_{\delta-\gamma} \) are the carbon concentrations at the delta gamma interface of delta phase side and gamma phase side respectively. \( V_{\text{delta}} \) is the velocity of delta/gamma interface due to the carbon diffusion in the delta phase and, according to assumption (v), is expressed as

\[
V_{\text{delta}} = \frac{C_\delta^* - C_\delta|_{\delta-\gamma}}{(C_\gamma|_{\delta-\gamma} - C_\delta|_{\delta-\gamma})} \cdot (1 - R_\gamma) \cdot \frac{d}{2} \frac{1}{\Delta t} \tag{4.6}
\]
where \( C_\gamma^* \) and \( C_\delta^* \) is the carbon concentration at the delta/gamma interface of gamma and delta phase side at the former time step, \( \Delta t \) is a time step in s, and \( d \) is the size of the grain which corresponds to the primary dendrite arm spacing.

4.2. Parameters

4.2.1 Carbon Diffusion Coefficient in Delta and Gamma Phases

The model requires accurate knowledge of the carbon diffusion coefficient. Some researchers [56,57] have shown that the diffusion coefficient is a function not only of temperature but also of carbon content. However, the carbon range of interest is small and the effect of carbon content itself is minor. Therefore, in this work, the diffusion coefficient is assumed to be a function of temperature only. The carbon diffusion coefficients which are derived from the literature are compared in Table 4.1. and Fig.4.3.

<table>
<thead>
<tr>
<th>bulk phase</th>
<th>( D_0 ) (cm(^2)/sec)</th>
<th>( Q_a ) (cal/mole)</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>( \delta )-iron</td>
<td>0.0127</td>
<td>-19450</td>
<td>58</td>
</tr>
<tr>
<td>( \delta )-iron</td>
<td>0.0127</td>
<td>-19400 (=-81.09 kJ/mol)</td>
<td>59</td>
</tr>
<tr>
<td>B.C.C.</td>
<td>0.020</td>
<td>-20100</td>
<td>60</td>
</tr>
<tr>
<td>( \gamma )-iron</td>
<td>0.0761</td>
<td>-32160</td>
<td>58</td>
</tr>
<tr>
<td>( \gamma )-iron</td>
<td>0.15</td>
<td>-34160 (=-142.8 kJ/mole)</td>
<td>59</td>
</tr>
</tbody>
</table>

As can be seen, the literature data is in good agreement. Therefore, the data from Reference 58 was used in the present work. From this figure, the carbon diffusion in the delta phase is about one order of magnitude greater than that in the gamma phase in the
temperature range of interest. However, the diffusion coefficient in gamma iron itself is also relatively large at high temperatures (around 1450°C) and about same order of that in B.C.C. iron around 900°C which is normally considered to be very rapid diffusion [61].

Fig 4.3 Carbon Diffusion coefficient in delta and gamma iron as a function of temperature.

4.2.2. Primary Dendrite Arm Spacing

In this model, the domain is a grain the size of which corresponds to the primary dendrite arm spacing. Therefore, it is very important to define the primary dendrite arm spacing. Several researchers have reported [62,63,64] that the primary dendrite arm spacing is a function of cooling rate as well as secondary dendrite arm spacing. These are compared in Fig.4.4. In this work, the regression of Mimura’s measurement [65]
is applied. Here, \( d \) is the primary dendrite arm spacing in \( \mu m \) and \( CR \) is the cooling rate in °C/s.

Some researchers have also reported that the carbon content affects the primary dendrite arm spacing. However, the effect of carbon content is not clear. For example, Suzuki et al. [38] reported that the primary dendrite arm spacing becomes larger with increasing of carbon content and indicated that Edvardsson [66] observed an opposite tendency.

![Graph showing the effect of cooling rate on primary dendrite arm spacing.](image)

Fig. 4.4 The effect of cooling rate on primary dendrite arm spacing.

In this work the primary dendrite arm spacing is assumed to be only a function of cooling rate, and Eq. (4.7) is used to estimate the primary dendrite size.

\[
d = 352.46 \cdot (CR)^{-0.3896}
\]
4.2.3. Description of Iron-Carbon Phase Diagram Curve

In order to define the boundary condition at the delta/gamma interface, the equilibrium carbon concentrations of delta and gamma phase at each temperature have to be quantified. Therefore, the equations which describe the iron-carbon phase diagram curve were characterized mathematically. There is a small difference in different published iron-carbon phase diagrams but, in this project, the phase-diagram was chosen, as shown in Fig. 2.5, where the peritectic carbon content is 0.16 wt. %.

To simplify the calculation, each equilibrium curve in the phase-diagram was assumed to be linear. The modified phase-diagram is shown in Fig.4.5 and the equations of the lines are listed in Table.4.2. It is expected that this assumption will work well in the region of interest.

![Phase diagram](image)

Fig.4.5 Definition of the curve in the phase-diagram.
Table 4.2. Equations for the each equilibrium curve in the phase-diagram shown in Fig.4.5 (where $T$ is the temperature in °C and $C$ is the carbon content in wt.%) 

<table>
<thead>
<tr>
<th>Curve</th>
<th>The equation to determine a carbon content</th>
<th>Equation No.</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>$C = \frac{-0.09}{45} (T - 1538)$</td>
<td>(4.8)</td>
</tr>
<tr>
<td>B</td>
<td>$C = \frac{-0.53}{45} (T - 1538)$</td>
<td>(4.9)</td>
</tr>
<tr>
<td>C</td>
<td>$C = \frac{0.09}{99} (T - 1394)$</td>
<td>(4.10)</td>
</tr>
<tr>
<td>D</td>
<td>$C = \frac{0.16}{99} (T - 1394)$</td>
<td>(4.11)</td>
</tr>
</tbody>
</table>

4.3 Numerical Method

The diffusion problem was solved with an implicit finite difference method. In order to incorporate the moving boundary, the technique used by Kamat et al. [61] was applied. However, it should be notified that, in the present model, the carbon diffuses
toward the delta/gamma interface while, in the Kamat et al’s model, the carbon diffuses away from the alpha/gamma interface.

A schematic diagram illustrating the nodal arrangement of the present model is shown in Fig. 4.6. At each time step, the mesh size was changed with a fixed number of nodes. The flow chart of the computer code written in Fortran77 is illustrated in Fig. 4.7.

start
Calculate the initial condition

Set the initial temperature at solidus temperature-$\Delta T$

Calculate the boundary condition, $C_s, C_\gamma$, using Eqs. (4.10), (4.11)

Calculate the carbon diffusion in delta phase

Calculate the carbon diffusion in gamma phase

Calculate the velocity of moving boundary, $V$, using mass balance Eq. (4.5)

Calculate the volume fraction of gamma phase, $R_\gamma$

Output the result

$R_\gamma = 1.0$ ?

No

Yes

End

Fig. 4.7 Algorithm of the computer program
The model calculates the volume fraction of gamma phase, $R_\gamma$ as a function of temperature for the carbon content and cooling rate specified as input. Convergence analysis revealed that a relationship between the time mesh and cooling rate exists. A finer time mesh was required to satisfy the same standard of error under higher cooling rates. In order to reduce the error to less than 1%, the relationship

$$\Delta t \times CR \leq 1.0$$

had to be satisfied. Here $\Delta t$ is the time mesh in s and CR is the cooling rate in °C/s. In the present work, all the results were obtained under the condition that $\Delta t \times CR$ equalled 0.5.

### 4.4 Results of the model

The results of the calculation depend on the initial carbon content and cooling rate. The predictions for 0.10% carbon steel cooled under several cooling rates are shown in Fig.4.8. The equilibrium condition which is derived from the lever rule is also shown in Fig. 4.8 as a reference.

As can be seen in this figure, because of the high carbon mobility at this high temperature, the transformation is significantly fast. Surprisingly, the higher cooling rate does not cause any substantial delay of the transformation. The two factors leading to this small cooling rate effect are summarized in Fig.4.9. Since greater cooling rate results in smaller primary dendrite arm spacing, the distance which carbon has to diffuse will decrease with increasing cooling rate. Furthermore, the average interface moving velocity increases when cooling rate increases.
Fig. 4.8 Evolution of volume fraction of gamma phase for several cooling rates.

Fig. 4.9 The average interface moving velocity and dendrite arm spacing as a function of cooling rate.
These characteristic features of peritectic transformation stem from the shape of the phase diagram. In the equilibrium phase diagram, the difference between equilibrium carbon concentrations of delta and gamma phase decreases considerably with decreasing temperature, while the decrease in the diffusion coefficient is small. Thus, as can be implied from Eq. (4.5), the transformation rate increases with decreasing temperature. On the other hand, with most of the other transformations in steel, such as eutectoid transformation, the difference between equilibrium carbon concentrations of two phases increases with decreasing temperature and the transformation is delayed with increasing cooling rate. A schematic diagram of this comparison is shown in Fig. 4.10.
Although the mechanism itself is different, this result corresponds to the result of the hyper-peritectic carbon steel which was found by Matsuura et al. [40-43]. In Matsuura’s experiment, the difference between the carbon concentration of the gamma phase at the delta/gamma interface and at the gamma/liquid interface increased dramatically with decreasing temperature. Interestingly, the present work also shows that a higher cooling rate results in a faster average interface moving rate. Therefore, the delta-to-gamma transformation does not undercool easily even though the cooling rate becomes very high.

4.5 Model Verification

4.5.1 Effect of the Diffusion Coefficient in Delta Phase

As mentioned earlier in the chapter, it was assumed that the carbon diffusion inside the delta phase is infinitely fast. This assumption is believed to be fulfilled because the carbon diffusion coefficient in the delta phase is much higher than that in the gamma phase. In order to confirm this assumption, a calculation was conducted using a finite carbon diffusion coefficient inside the delta phase. Instead of using Eq. (4.6), \( V_\delta \) was expressed as:

\[
V_{\text{delta}} = -\frac{D_\delta}{\left(C_\gamma|_{\delta-\gamma} - C_\delta|_{\delta-\gamma}\right) \frac{\partial C_\delta}{\partial x}} \left|_{\text{interface}} \right.
\]

where \( D_\delta \) is the carbon diffusion coefficient in delta iron which was obtained from Table 4.1 or Fig.4.3.
The computations for 0.10% carbon steel under several cooling rates are shown in Fig. 4.11. The results are very close to those in Fig. 4.8 which was obtained under the assumption that the carbon diffusion coefficient in the delta phase is infinitely large. Thus, the assumption is confirmed and all the calculations after this section were conducted assuming that the carbon diffusion coefficient in the delta phase is effectively infinite in value.

Fig. 4.11 Evolution of volume fraction of gamma phase for several cooling rates assuming a finite carbon diffusion coefficient in delta phase.
4.5.2 Effect of the Starting Temperature of Transformation

The first several runs of this model were calculated under the assumption that the undercooling from the peritectic temperature is negligible. Therefore, the starting temperature of the transformation was set at the peritectic temperature.

In the present work, in order to estimate the effect of transformation starting temperature, the starting temperature was set manually 10 to 40 °C below the peritectic temperature. An example of the result for 0.10% carbon steel cooled at 100 °C/s is shown in Fig. 4.12.

![Fig. 4.12 The effect of starting temperature of transformation on the transformation rate.](image)
The effect of the starting temperature on transformation rate was relatively small because the undercooling, which causes the increase in the driving force, accelerated the transformation rate. The change of the transformation finishing temperature was about half of the change of the starting temperature.

4.5.3. Effect of Other Elements on the Carbon Diffusion

Commercial steels contain not only carbon but also other elements such as manganese, silicon, sulfur, phosphorous etc. These elements may have an effect on the carbon diffusion in the steel. Especially manganese has been reported to have a very strong influence on the inhibition of carbon diffusion in the temperature range of 700 to 1000 °C [67-71]. Therefore, in this model the effect of manganese was also evaluated.

According to Kirkaldy [67], the diffusion of elements in the Fe-C-Mn ternary system is expressed by the equation,

\[ J_c = -D_{cc} \frac{\partial C_c}{\partial x} - D_{c,Mn} \frac{\partial C_{Mn}}{\partial x} \]  

(4.14)

where, \( J_c \) is the molar flux of carbon, \( D_{cc} \) is the diffusion coefficient of carbon in iron and \( C_c \) is the carbon concentration.

The second term on the right hand side of this equation represents the effect of the manganese and \( D_{c,Mn} \) is the diffusion coefficient of carbon in iron under the presence of a manganese concentration gradient. Kirkaldy measured the ratio of these two diffusion coefficients, \( D_{c,Mn}/D_{cc} \), at about 800 to 1000°C and mentioned that this ratio was independent of temperature and the value is negative (around -0.10). Therefore, under the
presence of manganese concentration gradient, carbon diffusion will be retarded when it diffuses toward the lower manganese concentration direction.

Inside the dendrite arms, manganese segregates [72] and, therefore, the concentration of manganese at the surface is higher than that of at the center because the equilibrium distribution coefficient of manganese at the liquid-delta interface is less than 1. Therefore, the manganese segregation inside the dendrite arm will work to prevent the carbon from diffusing toward the inside of the dendrite according to Eq. (4.14).

In the present work, this effect is also calculated but the effect of manganese was small and the result was almost equivalent to that of without manganese. This is because of the high mobility of the carbon atoms at this high temperature range. Carbon atoms can immediately move and compensate for the manganese distribution.

On the other hand, manganese also has the effect of changing the peritectic point. However, from the literature review, it seems to be a small effect for steels which contains relatively low manganese, 0.20 to 0.30 wt.%.

In summary, the presence of alloying additions such as manganese in an iron-carbon alloy do not affect the delta-to-gamma transformation rate significantly. These elements are considered to have an influence on the susceptibility of the cracks rather than the transformation itself.

4.5.4. Comparison with the Literature Results

It is very difficult to verify the results of this model because there is no reliable data from direct measurements of the delta-to-gamma transformation rate. For instance,
Nilles et al. [73] mentioned that they measured the shrinkage of 0.10% carbon steel using a Gleeble machine but they did not give measured shrinkage data or important experimental conditions such as cooling rate. At this high temperature, shrinkage measurements are very difficult. Steel is very weak, brittle and, because of the high mobility of atoms such as carbon, careful adjustment of the atmosphere is required to avoid decarburization at this high temperature.

Therefore, instead of verifying with the direct measurement data, the present model result was compared with experimental data for hyper-peritectic carbon steels derived by Matsuura et al [40-43].

Matsuura et al. measured the peritectic reaction rate using diffusion couples. They presented the relationship between time and the migration distance of the delta/gamma interface at various cooling rates. According to their result, the average moving velocity of the delta/gamma interface is 28 µm/s at a cooling rate of 10 K/s while it is about 9 µm/sec at a cooling rate of 1 K/s. These velocities are similar to those predicted here for hypo-peritectic steels. As can be seen in Fig. 4.9, the average moving velocities of the delta/gamma interface are about 15µm/s and 4 µm/s at cooling rates of 10 K/s and 1 K/s, respectively.

From this result, although slightly slower than the transformation rate of the hyper-peritectic carbon grade, the hypo-peritectic grade steels also transforms very rapidly.
4.6 Conclusions

In summary, the transformation from delta to gamma is extremely fast. As a consequence of the high carbon mobility at the temperature range of interest, the transformation follows closely the equilibrium transformation.

Increasing the cooling rate results in a slight decrease of transformation finish temperature. However, the effect of the cooling rate seems to be rather small. This tendency is similar to the result of hyper-peritectic grade steel which were obtained by Matsuura et al.

The relationship between cooling rate and transformation rate is considered to be a distinctive feature for the peritectic reaction. This feature is different from other transformations such as austenite to alpha-ferrite which will experience a large deviation from the equilibrium condition under higher cooling rates.

Since the volume fraction of the gamma phase is not sensitive to the cooling rate, the temperature itself dominates the transformation; therefore, a high cooling rate at meniscus in the continuous casting will result in a rapid transformation. This rapid transformation just below the solidification temperature will cause some problems which will be discussed later.
Chapter 5. Heat Transfer and Delta-to-Gamma Transformation in the Continuous Cast of Steel Slabs

To understand the phase transformations occurring in the solid steel shell, the delta-to-gamma transformation model derived in the previous chapter was applied for the cooling rates and temperature distribution encountered during the continuous casting process. A heat transfer model was, therefore, developed to calculate the temperature distribution in the solid steel shell. This chapter describes the results of the model simulations and discusses the findings on the phase transformation occurring in the solidifying shell.

5.1 Heat Transfer Model

5.1.1. Assumptions

In the continuous casting of steel, heat flows through the solid shell in the thickness (x), transverse (y) and axial (z) directions by conduction. Heat also flows in the axial direction as a result of the bulk motion of the descending strand. This latter mechanism turns out to be considerably more important than axial conduction for steel casting because of the relatively high casting speed and low thermal conductivity of steel. Consequently, heat conduction in the axial direction can be neglected. In the case of slab casting, because of the high aspect ratio of slabs, heat conduction in the transverse direction can also be neglected.
The effect of forced convection in the liquid pool was evaluated by Brimacombe [74]. His results indicated that the effective thermal conductivity may be five to ten times larger than the normal value to account for convective heat flow. However, he also mentioned that this effective thermal conductivity has a minor effect on the thickness of and temperatures in the solid shell. The present work focuses on the phenomena in the solid shell. Thus, forced convection in the liquid pool can be neglected.

Based on the information described above, the following assumptions were adopted in the formulation of the mathematical model.

(i) Heat flows in the casting direction (i.e. longitudinal direction) and in the direction perpendicular to the longitudinal midface plane (i.e. transverse direction) respectively are negligible.

(ii) The effect of forced convection due to the liquid metal turbulence is neglected.

(iii) Variations in heat flux due to mould oscillation and metal level fluctuations are ignored.

(iv) The latent heat can be taken into account by increasing the specific heat of the steel in the temperature range between the liquidus and the solidus.

(v) The difference between heat conduction in the delta phase and in the gamma phase are neglected.

5.1.2 Mathematical Formulation

One-dimensional steady-state heat flow in the solid shell can be described by

\[
\frac{\partial}{\partial x} \left( k \frac{\partial T}{\partial x} \right) = u \rho C_p \frac{\partial T}{\partial z}
\]  

(5.1)
where \( T \) is the temperature, \( k \) is the thermal conductivity, \( u \) is the withdrawal speed, \( \rho \) is the steel density, \( C_p \) is the heat capacity of the steel, \( x \) is the distance from the strand surface and \( z \) is the distance from the meniscus.

For steady-state conditions (constant casting velocity) \( u \) can be substituted by \( \partial z / \partial t \), and Eq. (5.1) becomes a one-dimensional heat transfer equation expressed as:

\[
\frac{\partial}{\partial x} \left( k \frac{\partial T}{\partial x} \right) = \rho \ C_p \ \frac{\partial T}{\partial t}
\]

5.1.3 Initial and Boundary Conditions

The initial temperature of the liquid steel at the meniscus, was assumed to be uniform. Therefore the initial condition is expressed as follows:

\[
t=0, \quad 0 \leq x \leq X_s/2, \quad T=T_i
\]

where \( X_s \) is the slab thickness and \( T_i \) is the initial temperature of the liquid steel in °C. In this model, the calculation was performed mainly for a 0.10% carbon steel and the liquidus temperature of this grade was computed to be 1529.5 °C by using Eq. (4.9). The initial temperature of the liquid steel for this grade was set at 1540 °C; therefore, the superheat is about 10°C. When different carbon grades were calculated, the initial condition was chosen such that the superheat was approximately 10°C.

There are two boundary conditions for this heat transfer model: one for the shell surface and the other for the centerplane of the slab. The first boundary condition is expressed as follows:
\[ t > 0, \ x = 0, \quad -k \frac{\partial T}{\partial x} = q(t) \]  
\[ (5.4) \]

where, \( q(t) \) is the heat flux in kW/m\(^2\) which changes with time in the mould.

The other boundary condition can be specified if the heat flow can be assumed to be symmetrical about the centerplanes. Under this condition, heat does not flow across the centerplanes and is expressed as:

\[ \text{at } x = X_s/2, \quad -k \frac{\partial T}{\partial x} = 0 \]  
\[ (5.5) \]

### 5.1.4 Evolution of Latent Heat

In this model, the latent heat effect is accounted for by increasing the heat capacity in the phase change temperature range. If a linear release of the latent heat across the temperature range is assumed, the apparent specific heat capacity, \( C_{pa} \), is expressed as:

\[
C_{pa} = \begin{cases} 
C_p & T \leq T_{sol} \\
C_p^{eff} & T_{sol} \leq T \leq T_{liq} \\
C_p & T_{liq} \leq T 
\end{cases}
\]
\[ (5.6) \]

where \( C_p \) is the heat capacity of the liquid and solid phases, \( T_{sol} \) and \( T_{liq} \) are the liquidus and solidus temperatures of the steel, respectively. \( C_p^{eff} \) is the effective heat capacity, which takes into account the latent heat, and is expressed as

\[
C_p^{eff} = C_p + \frac{\Delta H}{T_{liq} - T_{sol}}
\]
\[ (5.7) \]

Here, \( \Delta H \) is the latent heat in J/g.
5.1.5 Numerical Method

The model is based on the equation for one-dimensional heat transfer that is solved using an implicit finite difference method. The model was coded in Fortran77 and calculates the temperature distribution in the solid shell. In order to take the effect of the latent heat into account, the temperature at each node was checked at each time step to make sure that no node skipped the mushy zone and that all of the latent was locally extracted.

5.2. Parameters

5.2.1 Thermophysical Data

The thermophysical properties of steel adopted in these model calculations are presented in Table 5.1.

Table 5.1 Thermophysical properties of steel adopted in the heat transfer model.

<table>
<thead>
<tr>
<th>Symbol</th>
<th>Property</th>
<th>Value</th>
<th>Units</th>
<th>Ref.</th>
</tr>
</thead>
<tbody>
<tr>
<td>k</td>
<td>thermal conductivity</td>
<td>$15.9+11.51*10^{-3} \times (T+273)$</td>
<td>W/mK</td>
<td>75</td>
</tr>
<tr>
<td>Cp</td>
<td>heat capacity</td>
<td>0.68</td>
<td>J/g K</td>
<td>75</td>
</tr>
<tr>
<td>$\rho$</td>
<td>density</td>
<td>7.36</td>
<td>g/cm$^3$</td>
<td>76</td>
</tr>
<tr>
<td>$\Delta H$</td>
<td>latent heat</td>
<td>272.1</td>
<td>J/g</td>
<td>76</td>
</tr>
</tbody>
</table>
5.2.2 Axial Heat Flux Profile

To calculate the temperature distribution in the solid shell accurately, it is necessary to determine the axial heat flux profile (in Eq. (5.4)). In the case of slab casting, many researchers have estimated this heat flux profile by measuring the temperature distribution in the mould wall. The results are often presented as a function of time in the mould [77-84]. However, most of the research has focused on the heat flux profiles for the whole mould length and there are considerable variations among the heat flux curves in the meniscus region. On the other hand, in the present model, since the region of interest is very close to the meniscus, the heat flux profile in this local region needs to be determined fairly accurately.

In the present work, the heat flux data derived by Hiraki et al. [23] were used to calculate the temperature distribution in the solid shell. There are two reasons for using this data. Firstly, the heat fluxes were measured for a range of casting speeds up to 5m/min, which is equivalent to the thin slab casting condition. Secondly, these authors observed a critical heat flux, above which longitudinal cracks appeared. Since the present work focuses on the generation of longitudinal cracks, it is necessary to evaluate this critical heat flux using the present model.

In their investigation, the heat flux was derived at a location 45 mm below the meniscus. Thus, the data shown in Fig 2.6 were modified in order to estimate the heat flux profile at the meniscus. The transit time $t_T$ (s) employed for the modification is expressed as:

$$t_T = \frac{45}{V_c} \frac{60}{1000}$$  \hspace{1cm} (5.8)
where, \( V_c \) is the casting velocity in m/min.

From the data of Fig. 2.6 and Eq. (5.8), a relationship between transit time and heat flux can be derived as shown in Fig. 5.1 (open circles).

Unfortunately, since the data within 0.5 second from the meniscus are not available, the heat flux profile in this region had to be evaluated by extrapolation of the existing data. The heat flux in the meniscus region is a function of many casting conditions such as mould flux, mould design, etc. which interact in a complex manner. Therefore, there is no theoretically based equation to extrapolate the heat flux profile near the meniscus region. In the present work, empirical equations were employed to estimate the heat flux profile in the meniscus region. Several empirical equations have been presented by many researchers [77-84]; these equations are categorized in three types which are:

i) \( q(t_T) = a t_T^{-b} \) (power function) \hspace{1cm} (5.9)

ii) \( q(t_T) = a - b \sqrt{t_T} \) (square root function) \hspace{1cm} (5.10)

iii a) \( q(t_T) = a \cdot \exp(-bt_T) + c \) (exponential function) \hspace{1cm} (5.11)

iii b) \( q(t_T) = a \cdot \exp(-t_T) + b \cdot \exp(-t_T / n) + c \) (exponential function) \hspace{1cm} (5.12)

where a, b, c and n are empirical coefficients.

If Eq. (5.9) is adopted, the heat flux cannot be evaluated at the meniscus because the heat flux becomes infinity at time \( t_T = 0 \). Thus, Eq. (5.9) was not applied in this work. In the case of the exponential function, the difference between the curves derived from
Eqs. (5.11) and (5.12) was not significant and, therefore, in the present work, the simpler exponential equation, Eq. (5.11), was used.

Regression of Hiraki et al.'s data, using Eqs. (5.10) and (5.11), was conducted. The regression curves are plotted in Fig. 5.1 and are expressed as:

\[
q(t_T) = 4510 - 2160 \sqrt{t_T} \tag{5.13}
\]

\[
q(t_T) = 5960 \cdot \exp(-2.3 t_T) + 1490 \tag{5.14}
\]

Fig. 5.1 Heat flux in the meniscus region as a function of transit time derived from the data of Hiraki et al. [23].
The regression was also conducted using a third order polynomial equation in order to compare the result with Eqs. (5.13) and (5.14). The result of the regression is also plotted in Fig. 5.1 and the equation is expressed as:

\[ q(t_T) = -837t_T^3 + 4348t_T^2 - 7644t_T + 6214 \] (5.15)

As can be seen in Fig 5.1, the meniscus heat flux varies from 4500 to 7500 kW/m\(^2\) depending on the equation type. These values are higher than the values, 2000 to 2500 kW/m\(^2\), which have been reported as meniscus heat fluxes under lower casting speed conditions [85,86] and rather correspond to conditions of meniscus heat transfer observed in billet casting with oil lubrication [87]. However, the derived heat flux curves lie between two literature data which were reported by Davies et al. [81] and Wolf [83] under slab casting condition.

Recently, Grilles [82] measured a heat flux of approximately 5200 kW/m\(^2\) in the meniscus region for a casting speed of 2 m/min. Furthermore, while casting round billets, Dubendorff et al. [88] also observed a peak heat flux of approximately 5500 kW/m\(^2\) in the meniscus region for a casting speed of 3 m/min. These values also correspond to the order of the meniscus heat fluxes shown in Fig. 5.1.

Even though the curves derived from the regressions compare favorably with the range of other data reported in the literature, the variation of the meniscus heat fluxes is sufficiently large that it is difficult to select only one heat flux curve from the available information. Hence, in the present work, instead of selecting only one curve to represent the condition of Hiraki et al., three heat flux profiles were examined and compared.
From Fig. 5.1, Eq. (5.13) (square root function) did not adequately represent the behaviour of the heat flux data. A similar finding was reported by Davies et al [81]. In addition, Eq. (5.15) (polynomial function) results in considerable deviation for times longer than 2.5 seconds. On the other hand, the exponential type equation showed a good fit for the data derived by Hiraki et al. Thus, the heat flux profiles were re-defined by using exponential type equations. By setting the meniscus heat fluxes at 4500, 6200 and 7450 kW/m$^2$, which correspond to the values in Fig. 5.1, the heat flux profiles were manually fitted.

The results are shown in Fig. 5.2. Here, the curve for the high heat flux profile is given by Eq. (5.14). Medium and low heat flux profiles, for which the meniscus heat flux values correspond to 6200 and 4500 kW/m$^2$ respectively, can be expressed as:

\begin{align*}
q(t_T) &= 4700 \cdot \exp(-2.0t_T) + 1500 \quad (5.16) \\
q(t_T) &= 3000 \cdot \exp(-1.4t_T) + 1500 \quad (5.17)
\end{align*}

These three curves show similar heat flux values between 0.5 and 2.5 seconds but the value between the meniscus and 0.5 seconds differs considerably. Thus, the ensuing calculations were conducted by applying the high heat flux (Eq. (5.14)), medium heat flux (Eq. (5.16)) and low heat flux (Eq. (5.17)); and the effect of the heat fluxes was compared.
5.3 Model Validation

To validate the present model, the solid shell thickness was calculated under low, medium, and high heat flux conditions and compared with data from the literature. The present work focused on the solidification behaviour in the meniscus region (i.e., short transit time from the meniscus). Thus, the computed shell profiles were compared with data reported by Samarasekera and Brimacombe [89]. According to their results, the shell thickness, $S$, increases linearly with time, $t$, as follows [90]:

$$ S = 0.64t \quad (5.18) $$
This relationship holds for short transit time, i.e., within 6 s from the meniscus. A comparison between the results of the present work and those of Samarasekera and Brimacombe [89, 90] is shown in Fig. 5.3. As can be seen in the figure, the computed results of the present model are in good agreement with the literature data.

![Graph showing computed shell thickness profiles](image)

**Fig. 5.3** Comparison of the computed shell thickness profiles as calculated in the present work and by Samarasekera and Brimacombe [89, 90].

### 5.4 Calculation of Delta-to-Gamma Transformation

The heat transfer model was coupled to the delta-to-gamma transformation model described in chapter 4. The cooling rate which is necessary to determine the primary dendrite arm spacing was derived by dividing the interval between solidus and liquidus temperatures by the time taken to solidify. Subsequently, instead of using a constant cooling rate as shown in Fig. 4.8 or 4.11, the program was modified to use the calculated
temperature of the solid shell at each time step, and the volume fraction of the gamma phase was calculated at each position and each time step.

The result for the 0.1% carbon steel under the medium heat flux condition is presented in Fig. 5.4.

The phase transformation begins at the meniscus and the surface of the shell completely transforms from the delta phase to gamma phase at a time of 0.21 seconds. At the point corresponding to 1 second from the meniscus, the solid shell thickness is about 1.08 mm and about 35% (i.e., 0.38 mm) of the solid shell is completely gamma phase.

Calculations were also performed following the same procedure with the low heat flux data given in the Eq. (5.17) to obtain the results shown in Fig. 5.5. As shown in Fig. 5.5, the shell surface completely transforms to gamma at 0.61 seconds from the meniscus, which is almost three times as long as the time observed in Fig. 5.4. At the location corresponding to 1 second from the meniscus, the thickness of the solid shell is 0.86 mm.
and only 19% (i.e., 0.16 mm) of the solid shell completely transforms to gamma. The results for the high heat flux profile given by Eq. (5.14) are shown in Fig. 5.6.

Fig. 5.5 The distribution of the gamma phase in the continuous casting shell when the heat flux profile defined by Eq. (5.17) was adopted.

Fig. 5.6 The distribution of the gamma phase in the continuous casting shell when the heat flux profile defined by Eq. (5.14) was adopted.

In the case of the high heat flux profile, the shell surface completely transforms to gamma at 0.15 seconds from the meniscus. At the location 1 second from the meniscus,
the thickness of solid shell is 1.2 mm and about 40% (i.e., 0.5 mm) of the solid shell completely transforms to the gamma phase. By comparing Figs. 5.4 to 5.6, the effect of heat flux differences on the delta-to-gamma transformation can be summarized as follows:

Firstly, in the region very close to (within 0.5 to 0.6 seconds) the meniscus, the delta-to gamma transformation will be accelerated by increasing the meniscus heat flux.

Secondly, at longer times (after 0.5 to 0.6 seconds from the meniscus), the shell thickness increases as the meniscus heat flux increases. On the other hand, the thickness of the delta+gamma region becomes almost independent of the heat flux condition (0.70 mm at transit time 1 second for all heat flux conditions). Therefore, the variation in the heat flux can also result in a difference in the thickness of the gamma phase at longer times (because the thickness of the gamma phase can be defined by subtracting the thickness of delta+gamma region from the total solid shell thickness).

In conclusion, the variation in the heat flux at the meniscus can result in a large difference in the distribution of the volume fraction of the gamma phase in the solid shell. The mechanism responsible for this phenomenon will be discussed in the next section.

5.5 Thermal Analysis

To elucidate the mechanism of heat transfer during the delta-to-gamma transformation, a thermal analysis was conducted. It was found that, in the meniscus region, most (approximately 90%) of the heat extracted by the mould is used to remove the latent heat for all heat flux conditions studied; therefore, the percentage of sensible to total heat extracted from the solid shell remains low, as can be seen in Fig. 5.7.
Although the percentage itself is small, these profiles show a characteristic behaviour depending on the heat flux condition. For the high heat flux condition, the percentage of sensible heat extracted from the solid shell is higher than that for the lower heat flux conditions in the region very close (within 0.5 to 0.6 seconds) to the meniscus. After 0.5 to 0.6 seconds from the meniscus, this tendency changes, and the percentage of sensible heat extracted for the low heat flux condition is higher than that for the high heat flux condition.

The computed temperature gradient in the solid shell, as a function of transit time, is shown in Fig. 5.8. In all cases, the temperature gradient decreases with transit time.
For transit times less than 0.5 to 0.6 seconds, the temperature gradient for the high heat flux condition is higher than that for the low heat flux condition. However, after 0.5 to 0.6 seconds, the gradient in the shell becomes almost independent of heat flux conditions.

![Graph showing temperature gradient of the solid shell as a function of transit time under low, medium, and high heat flux conditions.](image)

Fig. 5.8 The temperature gradient of the solid shell as a function of transit time under low, medium, and high heat flux conditions.

Based on the results shown in Figs. 5.7 and 5.8, a mechanism for heat transfer during the delta-to-gamma transformation has been proposed, and it is shown schematically in Fig. 5.9. In the figure, \( t_1 \) and \( t_2 \) correspond to approximately 0.2 to 0.3 and 0.8 to 0.9 seconds from the meniscus, respectively, in Fig. 5.7.
Fig. 5.9 Schematic diagram of the temperature gradient in the shell at time $t_1$ (very close to meniscus) and $t_2$ (more than 0.6 s from the meniscus), under high and low heat flux conditions. (The light shaded area corresponds to the heat removed between time $t=0$ and $t=t_1$; the dark shaded area corresponds to the heat removed between $t=t_1$ and $t=t_2$.)

For the high heat flux condition, to remove the large amount of latent heat, the temperature gradient in the shell has to become greater than that in the lower heat flux conditions (time $t_1$ in Fig. 5.9 (a) and (b)). To generate this temperature gradient, the shell surface has to be cooled rapidly (since the temperature of the solidification front, the other end of the shell, is always the solidus temperature). On the other hand, during the same period, the low heat flux condition does not require a large temperature gradient and, therefore, the sensible heat extracted from the solid shell is low, as shown in Fig. 5.7. This
difference causes the change in the delta-to-gamma transformation rate at the shell surface within 0.5 to 0.6 seconds.

However, the applied heat flux (see Fig. 5.2) becomes lower and saturates at 0.5 to 0.6 seconds from the meniscus. Consequently, the temperature gradient in the shell also decreases gradually and becomes almost independent of heat flux conditions (time $t_2$ in Fig. 5.9 (a) and (b)). It should be noted that, although the temperature gradient becomes almost the same at longer times, higher heat flux results in lower surface temperature at $t=t_2$ (compare $T_{\text{surf}}$ and $T_{\text{surf},t_2}$ in Fig. 5.9) because the shell thickness for the high heat flux condition is thicker. However, the thickness of the delta+gamma region is defined by the position of the solidus and transformation finish temperatures, and therefore, it is almost independent of heat flux at longer times (because the temperature gradient is not a function of heat flux) as shown in Figs. 5.4 to 5.6.

In conclusion, when the meniscus heat flux is higher, the delta-to-gamma transformation in the solid shell proceeds faster in the region very close to the meniscus ($t<t_1$) because of the rapid cooling of the shell surface. This explains the difference in the delta-to-gamma transformation rate in this region among the different heat flux conditions, as shown in Figs. 5.4 to 5.6. Subsequently, after the heat flux decreases and saturates ($t_1<t<t_2$), the temperature gradient in the solid shell becomes almost independent of heat flux condition. Therefore, the thickness of the delta+gamma region becomes also independent of the heat flux conditions at longer times (i.e., after 0.5 to 0.6 seconds). On the other hand, under the higher heat flux condition, the thickness of the total solid shell, and therefore, the thickness of gamma phase becomes thicker (because the thickness of the
gamma phase is obtained by subtracting the thickness of the delta+gamma region from the total solid shell thickness). This explains the difference in the gamma phase thickness at longer times (after 0.5 to 0.6 seconds) in Figs. 5.4 to 5.6.
Chapter 6. Steel Shell Shrinkage and Stress Generation
due to the Delta-to-Gamma Transformation

In this chapter, all the models described before were combined with a shrinkage model of the steel. Using these shrinkage data, finite-element calculations were conducted to evaluate the stress generation in the steel shell in the meniscus region. The results of the calculations will be described and discussed.

6.1 Steel Shrinkage

6.1.1 Numerical Model for Steel Shrinkage

In chapter 5, the volume fraction of the gamma phase at each position in the solid shell was computed; thereafter, the density of the steel in the two phase region could be calculated using Eq. (2.3).

To estimate the density of the two phase region, the density of each phase needs to be calculated. As mentioned in chapter 2, Chandra et al. [47] used Eqs. (2.4) to (2.7) to calculate the mean coefficient of expansion. In the present model, to compute the shrinkage of the steels, the same equations were employed. Eqs. (2.4) and (2.5) were used to compute the specific volume of delta and gamma phase at each temperature, respectively; these were then converted to densities. In order to take the effect of carbon concentration into account, Eqs. (2.6) and (2.7) were used. In these equations, the lattice
parameters \( a_i^0 \) (i=\( \alpha \), \( \gamma \)) are obtained by converting the specific volume using the following equations.

\[
a_i^0 = \left( \frac{V_i \times 10^{24} \times M_i \times W_{Fe}}{N_a} \right)^{1/3}
\]

(6.1)

where \( W_{Fe} \) is the atomic weight of iron, 55.847 (g/mole), \( N_a \) is Avogadro's number, \( 6.022045 \times 10^{23} \), and \( M_i \) is the number of atoms per unit cell, i.e. \( M_\alpha = 2 \) (B.C.C.) and \( M_\gamma = 4 \) (F.C.C.).

Assuming isotropic behaviour, shrinkage in one-dimension was calculated from the density change with temperature. The program was written in Fortran77 program and the flow sheet of the computer program can be seen in Fig. 6.1.

Fig. 6.1 Algorithm of the computer program for calculating steel shrinkage
6.1.2 The Result of the Shrinkage Model

The calculated shrinkage for several steel grades, assuming equilibrium transformation, is shown in Fig. 6.2.

![Graph showing linear shrinkage of steel after solidification](image)

**Fig. 6.2** Linear shrinkage of the steel after solidification (equilibrium condition). Bold lines correspond to the zero ductility temperature (\(T_{sol} - 50°C\)) range.

From this figure, the following was obtained:

1. Thermal contraction in the gamma phase changes linearly with temperature. The slope of the curves is mainly a function of temperature and the effect of carbon
concentration on the slope is small. Differences in the shrinkage behaviour of different carbon grades are related to the delta-to-gamma transformation.

(2) The 0.10% carbon steel shows rapid shrinkage (i.e. delta-to-gamma transformation) that occurs within 50°C below the solidification temperature, which corresponds to the zero-ductility-temperature range. The total amount of shrinkage due to this transformation is approximately 0.3%. Since the ductility-to-fracture strain of steel just below the solidification temperature is approximately 0.2 to 0.3% [34], the shrinkage due to the phase transformation might be detrimental for this steel.

(3) The lower carbon steel (e.g. 0.05% carbon steel) also shows rapid shrinkage due to the phase transformation. However, for 0.05% carbon steel, the transformation occurs below 1450°C which is more than 50°C below the solidus temperature. Since this temperature range is below the zero-ductility-temperature range, the steel is stronger and the solid shell is thick enough to prevent the formation of longitudinal cracks. Therefore, longitudinal cracks are unlikely to form and may only result in wrinkling of the solid shell [50].

(4) In the case of 0.15% carbon steel, the delta-to-gamma transformation begins at the same temperature (1493°C) as in the 0.10% carbon steel. However, most of the transformation will be finished via the peritectic reaction at the solidification temperature. Therefore, only a small amount of delta phase will transform to gamma phase below the solidus temperature and, consequently, the total amount of shrinkage is not significant.
(5) In the case of higher carbon grades (e.g. 0.42% carbon steel), the entire delta-to-gamma transformation occurs with a co-existing liquid phase at the peritectic temperature. In this condition, molten steel will fill up and substitute the volume change due to shrinkage by flowing in between the dendrites. Hence, below the solidus temperature, steel shrinks only by thermal contraction; augmented shrinkage due to transformation does not occur. The differences between 0.10% carbon (i.e. peritectic carbon), 0.05% carbon (i.e., lower carbon) and 0.42% carbon (i.e. higher carbon) grades are schematically shown in Fig. 6.3.

(6) From comparison of the data for the 0.05 and 0.15% carbon grades, the shrinkage due to the delta-to-gamma transformation corresponds to a thermal shrinkage of more than 100°C which is about 4 times larger than the temperature range of the transformation.

Fig. 6.3 Schematic diagram of the difference of the phase transformation in the solid shell.
6.2 Stress Analysis

6.2.1 Concept and Assumptions

Based on this shrinkage model, the shrinkage at each point in the solid shell can be computed. The final stage of the present work was to evaluate the stress generation in the solid shell using the results from the shrinkage model. To this end, finite-element calculations were conducted to evaluate the stresses in the solid shell. In the present work, the commercial finite-element program ABAQUS was employed.

In order to estimate the generation of longitudinal cracks, the stresses in the transverse direction of the solid shell have to be evaluated. Thus, the calculation domain was chosen as illustrated (shaded plane) in Fig. 6.4. The planes were taken to be perpendicular to the axial direction and were selected at several positions in the solid shell.

![Fig. 6.4 Schematic representation of the domain of the stress calculation in the solid shell](image)

Fig. 6.4 Schematic representation of the domain of the stress calculation in the solid shell
The following assumptions were employed in the calculation.

(i) The temperature varies only through the thickness of the slab. In other words, the effect of the multi-dimensional heat transfer at the corner of the shell was ignored. The shrinkage was considered to occur only in the plane of the slice.

(ii) The shell was assumed to behave in an elastic manner because the temperature range of the calculation is within the zero-ductility-temperature range.

(iii) The ferrostatic pressure was ignored because the calculations were conducted in a region fairly close to the meniscus.

(iv) The effect of creep was ignored; however, it should be noted that the elastic modulus data normally includes the effect of creep [91].

The width of the calculated slab was set to 1000 mm in order to calculate the stresses under the same condition investigated by Hiraki et al. [23]. Because of the symmetric geometry, the calculation was performed for half of the width (500 mm). The casting speed was set at 2.0 m/min which is considered to be the critical condition for the medium carbon (i.e. peritectic grade) steels as shown in Fig. 2.6 [23].

6.2.2 Calculation Method and Parameters

Four-node bilinear elements were applied to calculate the stress distribution as shown in Fig. 6.5. In order to avoid numerical instability, the aspect ratio of the element was set to 10.
With regard to the mechanical boundary conditions in the direction of slab width (i.e., y direction), a symmetric boundary condition was applied at the centreline of the slab width. This boundary condition is expressed mathematically as follows:

\[ 0 < x < S, \quad y = 0, \quad v = 0 \quad \text{(6.2)} \]

where, \( S \) is the thickness of the slab in mm, \( v \) is the displacement in y direction.

The other boundary (the opposite side of the centreline) was set to move freely and the nodal displacements at that boundary were set equal to avoid the narrow face of the slab deforming beyond the mould region. This boundary condition is expressed as follows:

\[ 0 < x < S, \quad y = W/2, \quad v_1 = v_2 = \ldots = v_N \quad \text{(6.3)} \]

where, \( W \) is the slab width and \( v_i \) is the displacement of node \( i \) in y direction.
On the other hand, the displacement in the shell thickness direction (i.e., x direction) was set to move freely at all boundary nodes. No node in the wide face (x=0, 0 \leq y \leq W/2) was allowed to deform beyond the mould region. However, it should be noted that, no node in the wide face deformed beyond the mould region in the present computation.

Since the shrinkage of each position in the solid shell is already calculated in the present model, ABAQUS was used to compute only the stress distribution. The total elastic strain in the solidifying shell has thermal, \( \varepsilon^\text{th} \), and transformation, \( \varepsilon^\text{tr} \), components:

\[
\varepsilon = \varepsilon^\text{th} + \varepsilon^\text{tr}
\]  

(6.4)

The thermal strain is given by [92]:

\[
\varepsilon^\text{th} = \alpha(T_0 - T)
\]  

(6.5)

where \( \alpha \) is the thermal expansion coefficient, and \( T_0 \) is the temperature corresponding to a strain-free state (in this case, set to the solidus temperature).

Referring to Fig. 6.2, the total strain is a function of both carbon concentration and progress of the phase transformation. The finite-element code, ABAQUS, is not designed to compute transformation stresses and, therefore, the following procedure was implemented.

The shrinkage data, computed as explained in Section 6.1, already includes the effect of both thermal and transformation strains. To calculate the stress distribution generated by the total elastic strain using ABAQUS, a fictitious temperature, \( T^* \), was computed based on the shrinkage results of Section 6.1 by solving:

\[
\varepsilon = \alpha(T_0 - T^*)
\]  

(6.6)
for $T^*$. The thermal expansion coefficient, $\alpha$, was assumed to be constant and equal to $1.8 \times 10^{-5}$ °C$^{-1}$. The fictitious temperature at each node was then used to compute the strain and stresses assuming elastic behaviour.

The Poisson's ratio was assumed to be constant and equal to 0.3. In order to simplify the calculation, the elastic modulus was assumed to be constant in a given plane and was calculated by using Eq. (2.2). The average temperature of the solid shell, $T_{\text{avg}}$, which is given by:

$$T_{\text{avg}} = \frac{(T_{\text{sol}} + T_{\text{surf}})}{2} \quad (6.7)$$

was used in Eq. (2.2). Here $T_{\text{sol}}$ is the solidus temperature and $T_{\text{surf}}$ is the temperature of the shell surface.

Based on the data of Fig. 2.7, the UTS of the steel for the low carbon steel, $\sigma_{\text{UTS}}^{\text{low}}$, and 0.4% carbon steel, $\sigma_{\text{UTS}}^{0.4}$, can be expressed by:

$$\sigma_{\text{UTS}}^{\text{low}} = 170.8 - 0.113 \cdot T \quad (6.8)$$

$$\sigma_{\text{UTS}}^{0.4} = 147.7 - 0.100 \cdot T \quad (6.9)$$

where $T$ is the temperature in °C.

### 6.2.3 Results and Discussion

Since the surface shrinks the most, the maximum tensile stress was always observed at the surface of the solid shell. Therefore, the stress at the surface of the solid shell was compared to the UTS at the surface temperature.
The result of the calculation for the 0.1% carbon grade subjected to the heat flux obtained from Eq. (5.16) is shown in Fig. 6.6.

As the distance from the meniscus increases, the tensile stress at the surface of the solid shell increases rapidly. However, below 10 to 20 mm (i.e. 0.3 to 0.6 seconds) from the meniscus, the stress reaches a maximum value of approximately 7 MPa. An explanation for this observation can be obtained based on the results shown in Fig. 5.3. At first, the phase transformation occurs at the shell surface and the tensile stress increases rapidly. However, transformation just beneath the shell surface begins subsequently.
resulting in shrinkage in that region. Therefore, the tensile stress distributes to a wider range in the shell and the stress at the solid shell does not increase further.

In addition, the UTS of the solid shell surface also increases rapidly because the surface temperature decreases drastically. Interestingly, as the distance from the meniscus increases, the UTS behaves similarly to the maximum tensile stress at the shell surface; however, the maximum tensile stress at the shell surface is always below the UTS curve. Therefore, if uniform cooling is achieved, longitudinal cracks may not be generated with the medium heat flux profile.

6.2.4. Effect of the Heat Flux

As indicated in chapter 5, the heat flux profile at the meniscus will considerably change the phase transformation rate in the solid shell. Hence, stress calculations for a 0.1% carbon steel grade under low heat flux (i.e. Eq.(5.17)) and high heat flux (i.e. Eq.(5.14)) were also conducted. Comparison of the tensile stress and UTS at the shell surface for the low, medium and high heat flux conditions is shown in Fig. 6.7. Based on this figure, the maximum tensile stress at the shell surface becomes lower with decreasing heat flux in the meniscus region. The UTS behaves in a similar fashion. Normally, decreasing the meniscus heat flux is considered to be one of the most effective ways to prevent the generation of longitudinal cracks because of the lowering of stress generation. However, based on the present results, decreasing the meniscus heat flux also causes a decrease of UTS, since the solid shell temperature will increase with decreasing heat flux. In other words, even under the high heat flux condition, although the tensile
stresses at the shell surface become much higher, the stress level is still below the UTS. Therefore, increasing the heat flux itself does not result in a direct cause of longitudinal cracks.

![Graph showing comparison of UTS and computed maximum tensile stresses at shell surface under different heat flux conditions.](image_url)

Fig. 6.7 Comparison of UTS and computed maximum tensile stresses at shell surface under different heat flux conditions.

To estimate the longitudinal crack susceptibility, the ratio of maximum tensile stresses at the shell surface to UTS at the same distance from the meniscus were calculated under different heat flux conditions. The results are shown in Fig. 6.8. The figure shows that the ratio increases with distance from the meniscus and then reaches a maximum. The higher the heat flux, the sooner the ratio reaches a maximum value.
However, this particular ratio is almost independent from the heat flux condition and lies between 0.70 and 0.75, which is below the cracking condition.

![Graph showing susceptibility to longitudinal cracks under different heat flux conditions](image)

Fig. 6.8 The susceptibility to longitudinal cracks under different heat flux conditions

Thus, from the point of view of stress generation associated with rapid transformation (i.e. rapid cooling), it is not possible to explain why a lower heat flux at the meniscus region reduces the longitudinal cracking problem. However, as shown in Fig. 6.7, the maximum tensile stresses for high and medium heat flux conditions exceed the UTS corresponding to the low heat flux condition. Therefore, it can be inferred that if there is a hot spot caused by a local reduction of heat flux in the meniscus region, the solid shell may experience cracking. To further study the formation of longitudinal cracks, the role of these hot spots shall be discussed.
6.2.5. Effect of Hot Spots

After the shell solidifies, it is cooled further. During this period, there is a temperature fluctuation at shell surface which may be caused by the irregularity of the heat transfer in the meniscus region due to the metal level fluctuation, non-uniform infiltration of mould flux, etc. The non-uniform infiltration behaviour of the mould flux may be related to many factors, such as variation of the melting rate of mould flux due to the turbulent fluid flow in the mould, non-uniformity of mould flux properties due to, for example, the inclusion absorption, etc. Although these factors are important for the uniform heat transfer, they are also difficult to control perfectly in a real process at present time. Therefore, the shell surface always experiences the temperature fluctuation when it is cooled.

To estimate the effect of this fluctuation, the temperature of the solid shell surface was artificially increased locally until the UTS became equal to the maximum tensile stress at the shell surface, which is the critical condition for longitudinal crack formation. The calculation was conducted by changing the percentage increase of the temperature fluctuation and the results under different heat flux conditions are shown in Fig. 6.9. This figure shows that if there is a point with approximately 30% lower cooling (in terms of temperature) than the rest of the shell, a longitudinal crack will be generated under the high heat flux condition. On the other hand, in the case of the low heat flux condition, a critical fluctuation increase of approximately 40% is needed. Based on this result, although the ratio of maximum tensile stresses at the shell surface to UTS itself did not change significantly (i.e. Fig. 6.8), the heat flux condition can affect the value of the
maximum temperature fluctuation permissible. When the heat flux increases, this allowance becomes smaller, which leads to an increase in the sensitivity to form longitudinal cracks.

In order to derive the critical condition for the generation of longitudinal cracks under the conditions of the present work, the ratio of maximum tensile stresses to UTS was employed. Taking the point which gives a maximum ratio, the effect of hot spots was calculated as follows. First, the heat fluxes shown in Fig. 5.2 (i.e., given by Eqs. (5.14), (5.16) and (5.17)) were integrated from time 0 to 1 second under each heat flux condition. As a result, the total quantity of heat extracted from the solid shell within 1 second from
the meniscus could be estimated; the calculated heat extracted per unit area was 3110, 3530 and 3820 kJ/m² for the low, medium and high heat flux profiles respectively. Then, the ratio of these values was calculated. For example, the ratio of the extracted heat under the low heat flux condition to that under the medium heat flux condition was 0.88. This ratio shows how much the hot spot delays the solidification and, therefore, affects the weakness of the solid shell. When this ratio is equal to 1, it means that there is no hot spot on the shell.

The relationship between this ratio and the ratio of maximum tensile stress and UTS can be seen in Fig. 6.10. When there is more than 10% retardation of the heat removal, (i.e., \( \frac{Q_{HS}}{Q} \) is less than 0.9) the ratio of maximum tensile stress to UTS exceeds 1 and causes cracking.

![Fig. 6.10. The relationship between retardation of heat removal (as given by the ratio \( \frac{Q_{HS}}{Q} \)) and the critical stress.](image)
The same calculation was conducted regarding the shell thickness. Instead of taking the ratio of extracted heat, the ratio of shell thicknesses shown in Figs. 5.4 to 5.6 was calculated. The results are shown in Fig. 6.11. From this figure, the critical retardation of the shell growth is approximately 84%.

![Graph showing the relationship between retardation of shell growth and the critical stress.]

Fig. 6.11 The relationship between retardation of shell growth and the critical stress.

Based on experimental results, Saeki et al. [7] observed that 10% retardation (corresponding to 90% in Fig. 6.11) of the shell growth is critical for longitudinal cracks under low casting speed conditions (i.e., 1.2 m/min). Furthermore, Tanaka et al. [93] examined the cross section of longitudinal cracks in SUS 304 stainless steel coils cast by a,
twin-roll machine; the 304 stainless steel also undergoes delta-to-gamma transformation. They observed longitudinal cracks when the retardation of the shell growth became more than 20%.

These results from the literature are in fairly good agreement with the results of the present work. Based on the comparison with literature data, it is interesting to note that the critical retardation for the formation of longitudinal cracks seems to increase when a process associated with higher heat flux is applied. The reason for this tendency may be related to the steel properties, such as the microstructure of the solid shell, given that the higher heat flux results in a finer grain size (i.e. dendrite size). It should be noted, however, that the strip casting process is quite different from the conventional continuous casting processes, including thin slab casting process, and therefore, to explain the difference in these threshold values, further study needs to be done.

In conclusion, as the distance from the meniscus increases, the maximum tensile stress at the shell surface increases but, in the present calculation, the stress is always below the UTS curve under any heat flux condition. Thus, shrinkage due to rapid transformation itself does not cause longitudinal cracks even under the high heat flux condition.

However, as the heat flux increases, larger fluctuations in the heat flux may be experienced for a given variation in thermal resistance (see Fig. 6.12). This fluctuation of the heat flux will generate hot spots and, if the retardation of shell growth or the heat removal from the shell deviates from the threshold values, the shell may experience longitudinal cracks.
Many factors can cause this non-uniformity of the thermal resistance in the meniscus region: uneven infiltration of mould flux, metal level fluctuation, etc. Whatever their origin, in order to eliminate longitudinal cracks, the elimination of hot spots (i.e., uniform heat removal) in the meniscus region is very important. In fact, despite the fact that the process is associated with extremely high heat fluxes, pilot plant strip casting runs, free of longitudinal facial cracks, have been reported [94] when a uniform heat extraction was achieved.
However, in continuous casting with mould flux, it is very difficult to eliminate all the factors which will cause the non-uniformity of the thermal resistance. Thus, from a practical point of view, one of the most desirable strategies consists in reducing the heat flux in the meniscus region in conjunction with an effort to minimize the non-uniformity of the heat extraction.

### 6.2.6 Effect of Carbon Concentration

In the case of higher carbon steels (0.16-0.53% C), the steel shrinks only by thermal contraction which does not generate tensile stresses at the solid shell surface sufficient to induce longitudinal cracks. In this work, a stress calculation was conducted for a 0.4% carbon steel under a medium heat flux condition (i.e. Eq. (5.16)). The result of this calculation is shown in Fig. 6.13.

![Graph showing tensile stress at the solid shell surface for a 0.4% carbon steel grade under medium heat flux condition. The UTS is also shown.](image-url)
The UTS for the 0.4% carbon steel, as calculated by Eq. (6.9), is plotted as well. As expected, the stress for the 0.4% carbon grade is far below the UTS of this steel and the ratio of maximum tensile stresses to UTS at the shell surface was always below 0.5, which is much smaller than the result for the 0.1% carbon grade.

In order to evaluate the effect of the delta-to-gamma phase transformation on stress generation at the solid shell surface, the maximum tensile stress at the shell surface for the 0.4% carbon steel was compared with that of 0.1% carbon steel. As shown in Fig. 6.14, the stress for the 0.4% carbon steel grade is much lower than the one for the 0.1% carbon steel grade and, therefore, the susceptibility to longitudinal cracking is much lower than that of the 0.1% carbon steel.

Fig. 6.14 Comparison of stress at the solid shell surface under medium heat flux condition for the 0.1% and 0.4% carbon steel grades.
Chapter 7. Summary and Conclusions

The present study was undertaken to evaluate the formation of longitudinal cracks in steel slabs on the basis of fluctuations in heat transfer in the meniscus region. Mathematical models of the delta-to-gamma transformation, heat transfer in the mould and steel shrinkage were developed. The results of the models were used to calculate the stress distribution in the solid shell employing the commercial finite-element program, ABAQUS.

Firstly, the delta-to-gamma transformation model was developed assuming carbon diffusion control. From the model, it has been found that:

1. The transformation from delta to gamma is extremely fast. As a consequence of the high carbon mobility in the temperature range of interest, the transformation follows the equilibrium condition closely.

2. Increasing the cooling rate results in a slight decrease of the transformation finish temperature. However, this effect seems to be rather small. This tendency is similar to the results for hyper-peritectic steel grades, obtained by Matsuura et al [40-43].

3. The weak relationship between cooling rate and transformation rate is considered to be a distinctive feature of the peritectic reaction. This feature is different from other transformations such as austenite to alpha-ferrite, where high cooling rates cause a significant deviation from the equilibrium condition.
(4) Since the rate of transformation is not sensitive to the cooling rate, the temperature itself dominates the transformation and, therefore, high cooling rates at the meniscus during continuous casting will result in a rapid transformation.

Secondly, the heat transfer in the continuous casting mould was modeled and coupled with the phase transformation model. The following results were obtained:

(1) In the region very close to (within 0.5 to 0.6 seconds) the meniscus, the delta-to-gamma phase transformation will proceed faster when the meniscus heat flux is higher because of the rapid cooling of the shell surface. Thus, the variation in the heat flux at the meniscus can result in a large difference in volume fraction of gamma phase at the surface of the solid shell in this region.

(2) At longer times (after 0.5 to 0.6 seconds from the meniscus), since the temperature gradient in the solid shell decreases and becomes independent of the heat flux condition, the thickness of the delta+gamma region becomes also independent of heat flux condition. On the other hand, as the meniscus heat flux increases, the thickness of the total solid shell increases and, therefore, the thickness of the gamma phase (= total solid shell thickness minus the thickness of delta+gamma region) increases by increasing the meniscus heat flux at longer times.

(3) As a result, the difference in the heat flux at the meniscus can result in a large variation of the volume fraction of gamma phase in the solid shell. This phenomenon is mainly caused by the temperature change in the solid shell in order to balance the latent heat with the heat extracted by the mould because the rate of the delta-to-gamma transformation is mostly dominated by only temperature as mentioned above.
Finally, the phase transformation model and heat transfer model, as well as steel shrinkage model, were coupled and finite-element calculations have been conducted using these model results to predict the stress distribution in the solid shell. This analysis revealed that:

(1) Shrinkage due to the rapid transformation itself does not cause longitudinal cracks even under the high heat flux condition. As the distance from the meniscus increases, the UTS and the maximum tensile stress at the shell surface behave similarly; however, the maximum tensile stress at the shell surface is always below the UTS curve under any heat flux condition. Therefore, if a uniform cooling is achieved, longitudinal cracks may not be generated even under the high heat flux condition.

(2) Although the ratio of maximum tensile stresses at the shell surface to UTS does not change significantly with heat flux, the heat flux can affect the value of the maximum temperature fluctuation permissible.

(3) When there is more than 10% retardation of the heat removal or the retardation of the shell growth becomes less than 84%, the ratio of maximum tensile stress to UTS exceeds 1 and causes cracking. The latter value is in good agreement with literature data.

(4) The stress for the 0.4% carbon steel grade, which does not undergo delta-to-gamma transformation after it solidifies, is much lower than that for the 0.1% carbon steel grade and, therefore, the susceptibility to longitudinal cracking is much lower than that of the 0.1% carbon steel.
It is apparent from the above findings that the formation of longitudinal cracks is related to the stress generation due to the delta-to-gamma transformation and the appearance of hot spots in the meniscus region. A schematic diagram of the proposed mechanism of the formation of longitudinal cracks is shown in Fig. 7.1.

(a)
The variation of thermal resistance occurs in the meniscus region due to, for example, the non-uniform infiltration of the mould flux, meniscus level fluctuation, etc.

(b)
The variation of thermal resistance generates hot spots, while the remaining shell shrinks and generates tensile stress. When this tensile stress exceeds the UTS of hot spots, the shell may experience longitudinal cracking. (At this point, retardation of the shell is beyond the threshold value.)

Fig. 7.1 Proposed mechanism of formation of longitudinal cracks.
In conclusion, to eliminate longitudinal cracks, uniform heat removal in the meniscus region is of utmost important. If the uniformity of the heat extraction is achieved, even under high heat flux condition, the tensile stress at the shell surface does not exceed the UTS of the shell surface and there is no cracking problem. However, as heat flux increases (i.e., cooling rate increases), the maximum temperature fluctuation permissible before cracking occurs decreases. Thus, ironically, the practical way to eliminate longitudinal cracks when casting at high speeds is to reduce the heat flux in the meniscus region. This is also important from the economical point of view because the mould life in the thin slab casting process, which is associated with the high heat flux in the meniscus region, has been reported to be relatively short compared to mould life in conventional continuous casting. However, in terms of longitudinal crack elimination, uniform infiltration and uniform properties of the mould flux in conjunction with the heat flux reduction are also important.
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