Modeling of Thermomechanical and Metallurgical Phenomena in Steel Strip during Hot Direct Rolling and Runout Table Cooling of Thin-Cast Slabs

by

CORNELIUS ANAEDU MUOJEKWU

B. Sc. (Metallurgical Eng.) The University of Ife, Nigeria 1987
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to the required standard

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Department of Metals & Materials Engineering

The University of British Columbia
Vancouver, Canada

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ABSTRACT

The present research was directed at adequate prediction of the temperature, deformation behavior (roll force, flow stress, strain and strain rate) and microstructural evolution (recovery, recrystallization, grain growth, austenite and ferrite grain sizes) during rolling in the Compact Strip Production (CSP) process, as well as the final mechanical properties of the hot rolled strips. This was accomplished with the aid of integrated process modeling, involving mathematical simulation, laboratory experiments and industrial campaigns. The study covered two conventional plain carbon steel grades, the A36 (AISI 1018, 0.17C-0.74Mn) and DQSK (AISI 1005, 0.038C-0.3Mn), and a range of plain carbon steel grades (0.06-0.09 C, 0.16-0.9 Mn) produced at HYLSA’s CSP mill at Monterrey, Mexico.

In the laboratory, compression tests (both single and double-hits) were carried out on the Gleeble 1500 thermomechanical simulator in order to elucidate the effect of coarse austenite grain size on the flow stress and recrystallization behavior of the plain carbon steels. It was found that coarse grain size not only decreased the flow stress at a given strain but also substantially reduced the tendency toward dynamic recrystallization. An increase in grain size from 244 to 1110 μm which is typical of the first stands of a conventional finishing mill and CSP hot-strip mill respectively, resulted in up to a 30 MPa decrease in the flow stress of both A36 and DQSK steel grades at similar operating conditions of temperature, strain and strain rate. In combination with flow stress curves for finer grain sizes, it was found that a distinctive boundary exists between flow curves with peaks and those without peaks, a very important finding that allowed for a novel quantitative delineation of the occurrence of peaks in flow curves for any given set of deformation conditions. It was found that for the range of grain sizes and strain measured, complete recrystallization ($F_x \geq 0.95$) occurred between 2 and 4 seconds at 1100 °C and a strain
rate of about 5 s\(^{-1}\). The recrystallization kinetics obtained for fine austenite grains were found to be inadequate when extrapolated to the coarse grain size of A36 steel.

In order to validate the model and laboratory results with mill measurements from an operating CSP plant, an industrial trial was carried out at HYLSA's CSP mill in Monterrey, Mexico. During the industrial campaign, intermediate temperature measurements were made, CSP slab and coil samples were acquired, and all measured and recorded mill data and practices were obtained. The prior as-cast austenite grain size from one of the slabs was estimated to be 990 \(\mu\)m. Analysis of a cobbled strip revealed that it takes the first two stands to break down the as-cast structure and subsequent stands to refine the resulting equiaxed microstructure through recrystallization as in the conventional rolling. This finding is believed to have ramifications for CSP rolling of crack-sensitive grades as well as the emerging technology of strip casting. The ferrite grain size and mechanical properties of CSP strips were found to be dependent on the coiling temperature, strip thickness and steel composition. The final ferrite grain size decreased as the coiling temperature and strip thickness were reduced. Low coiling temperature sometimes lead to a non-polygonal structure as was the case in a 3.15 mm thick strip coiled at 560 °C. The yield and ultimate tensile strengths (YS and UTS) decreased with increasing coiling temperature and strip thickness as well as with reduced carbon equivalent and residual content, while the percent elongation increased. A 25 percent decrease in yield strength, a 10 percent decrease in ultimate tensile strength and a 2 percent increase in percent elongation were found to be associated with the annealing effect of slow cooling of coil bundles for strips coiled at 715 ± 15 °C and left to cool in air.

Comprehensive mathematical modeling of the rolling process was carried out employing finite difference and finite element analysis. The CSP mill measurements were utilized to validate model predictions of temperature, roll force, grain size and mechanical properties. Good
agreement was obtained between prediction and measurement in most of the cases. An estimate of the heat extraction from the various mill sub-units was conducted from the validated calculations. It was found that heat loss by radiation accounted for 48-51 percent of the total heat loss, the work rolls accounted for 41-44 percent, the descaling unit accounted for 4-6 percent and the interstand sprays accounted for the remaining 3-4 percent. It was found that the uniform strain model consistently predicts lower temperatures than the target exit temperature for thin gauges due to a low estimate of deformation heat. Model results captured the details of heat transfer, deformation, recrystallization and austenite decomposition in the CSP mill. The effect of various mill parameters were elucidated, and the similarities and differences between conventional cold-charge rolling and CSP rolling were highlighted.
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</tr>
<tr>
<td>P</td>
<td>Rolling Pressure (Pa)</td>
</tr>
<tr>
<td>Pe</td>
<td>Peclet number, $\frac{uL}{\alpha}$</td>
</tr>
<tr>
<td>P</td>
<td>Mean rolling pressure (Pa)</td>
</tr>
<tr>
<td>Pr</td>
<td>Prandtl number, $\frac{\mu C_p}{k}$ or $\frac{v}{\alpha}$</td>
</tr>
<tr>
<td>q</td>
<td>Heat flux (J/m$^2$)</td>
</tr>
<tr>
<td>Q</td>
<td>Activation energy (J/mol.)</td>
</tr>
<tr>
<td>r</td>
<td>Radius in cylindrical co-ordinate system</td>
</tr>
<tr>
<td>r₀</td>
<td>Roll radius (m)</td>
</tr>
<tr>
<td>Rₙ</td>
<td>Vapor stem radius (m)</td>
</tr>
<tr>
<td>Re</td>
<td>Reynold's number, $\frac{uL}{v}$</td>
</tr>
<tr>
<td>T</td>
<td>Temperature (°C)</td>
</tr>
<tr>
<td>Tₐ</td>
<td>Ambient Temperature (°C)</td>
</tr>
<tr>
<td>Tᵢ</td>
<td>Initial Temperature (°C)</td>
</tr>
<tr>
<td>TS</td>
<td>Tensile Strength (MPa)</td>
</tr>
<tr>
<td>ΔTₛ</td>
<td>Wall superheat (°C), $T_s - T_{sat}$</td>
</tr>
</tbody>
</table>
\( \Delta T_{\text{sub}} \) Water subcooling (°C), \( T_{\text{sat}} - T_i \)

\( t \) Time (s)

\( t_{0.5} \) Time for 50 percent recrystallization (s)

\( U \) Bulk flow fluid velocity (m/s)

\( \text{UTS} \) Ultimate tensile strength (MPa)

\( X \) Recrystallized fraction

\( \dot{W} \) Water flux (l/m²s)

\( \text{YS} \) Yield Strength (MPa)

\( Z \) Zener-Hollomon parameter (s⁻¹)

\( Z_b \) Boundary Zener-Hollomon parameter (s⁻¹)

\( \alpha \) Thermal diffusivity (m²/s)

\( \delta_{c0} \) Macrolayer thickness (m)

\( \varepsilon \) Mean strain rate (1/s)

\( \varepsilon \) Strain in flow stress equations

\( \varepsilon \) Emissivity in radiation equations

\( \varepsilon_c \) Critical strain

\( \varepsilon_p \) Peak Strain

\( \varepsilon_{0.5} \) Strain for 50 percent recrystallization

\( \gamma \) Grain boundary interfacial energy (J/m)

\( \lambda \) Vapor-liquid interface wavelength (m)

\( \mu \) Coefficient of dynamic viscosity (Kg/m.s)

\( \rho \) Density (kg/m³)

\( \sigma \) Stress (MPa)
\( \sigma \) \hspace{1cm} \text{Surface tension in Equation (5.80), } \text{N/m}

\( \sigma_{ss} \) \hspace{1cm} \text{Steady-state stress (MPa)}

**Subscripts and Superscripts**

\( \alpha \) \hspace{1cm} \text{Ferrite}

\( \text{conv} \) \hspace{1cm} \text{Convection}

\( \text{FB} \) \hspace{1cm} \text{Film boiling}

\( \text{gb} \) \hspace{1cm} \text{Grain boundary}

\( \text{gg} \) \hspace{1cm} \text{Grain growth}

\( \gamma \) \hspace{1cm} \text{Austenite}

\( l \) \hspace{1cm} \text{Liquid}

\( l-s \) \hspace{1cm} \text{Liquid-solid contact}

\( \text{nr} \) \hspace{1cm} \text{no recrystallization}

\( r \) \hspace{1cm} \text{Roll}

\( \text{rad} \) \hspace{1cm} \text{Radiation}

\( \text{rex} \) \hspace{1cm} \text{Recrystallization}

\( s \) \hspace{1cm} \text{Solid (strip)}

\( \text{sat} \) \hspace{1cm} \text{Saturation}

\( \text{TB} \) \hspace{1cm} \text{Transition boiling}

\( \text{tp} \) \hspace{1cm} \text{Triple-point (solid-liquid-vapor in contact)}

\( v \) \hspace{1cm} \text{Vapor}

\( \infty \) \hspace{1cm} \text{Free-stream or ambient}
ACKNOWLEDGEMENT

I would like to dedicate this work to the late Christopher Ndupu Muojekwu, my uncle and godfather, whose encouragement and assistance formed the foundation of my academic pursuit, and to my CREATOR who has been making my life a continuing miracle.

My gratitude goes to my supervisors, Dr. I.V. Samarasekera and the late Dr. J.K. Brimacombe, for their invaluable counsel and for providing the research assistantship that supported this study. I would like to acknowledge the contributions of Dr. D. Jin to the finite-element analysis, V.H. Hernandez-Avila to the runout table model, Professor E.B. Hawbolt, Dr. W.P. Sun, Dr. M. Militzer and Rassoul Pandi to the microstructural evolution investigations. The assistance of B. Chau, P. Wenman and R. Cardeno in performing the Gleeble tests and quantitative metallography is gratefully acknowledged. My gratitude is also extended to SMS (in particular, Dr. Gunter Knepe) for providing some CSP mill data and to HYLSA (in particular, Ing. Miguel Angel Pedroza, Dr. L. A. Leduc-Lezama, Ing. Leonel Elizondo Trevino, Ing. Julio M. Munoz Baca, Luis Z. Cruz Pitta and Ricardo) for a rewarding plant trial at their CSP plant in Monterrey, Mexico. My appreciation extends to my family, my friends and colleagues (in particular, Shirley-Ann Harris, Sunil Kumar, Bernado Hernandez-Morales, Mark Adjei-Sarpong, Joel Kapusta and Unmesh Wankhede) for their priceless support and solidarity.
A TRIBUTE TO DR. J. KEITH BRIMACOMBE (1943-1997)

It is with a heavy and saddened heart that I am writing this tribute to the great KEITH BRIMACOMBE who had supervised and guided my graduate research (in conjunction with Dr. I.V. Samarasekera) since I joined UBC in 1990. His death on the 16th of December, 1997 and his consequent absence at the final oral examination of this work is still a shock to me.

I joined the famous Centre for Metallurgical Process Engineering at the recommendation of Dr. Patrick Anagbo (my supervisor at the University of Lagos, Nigeria) who had worked with Keith during his sabbatical leave. Today, I am proud to say that my greatest treasure and asset is the priceless training and unrivaled exposure I received during my graduate studies with Keith and Indira.

Keith genuinely believed in the empowerment of the individual with knowledge, a belief that has driven most of his work. Keith’s career was full of countless awards and fame, but above all, I am convinced that Keith’s greatest gift to the world is the myriad of individuals around the globe that Keith had himself empowered with exceptional knowledge. To Keith’s credit, these well-rounded people that he trained are now contributing meaningfully to society, both locally and globally, bearing wonderful testimonies and living manifestations of his legacy.

May Keith’s colossal soul rest in perfect peace. We all miss you, Keith!
Chapter 1

INTRODUCTION

1.1. Advances in Flat Steel Rolling

Hot strip mills are utilized to manufacture coils of prescribed gauge and properties (physical and mechanical, surface tolerances, etc) from a slab of known dimensions and chemical composition. Steel strip is one of the most versatile hot rolled products, with a wide variety of applications ranging from automobile bodies and transformer sheets, to refrigerator and stove bodies. The classification of strip applies to sheet ranging in thickness from 2 to 8 mm. Hot strip mills owe their existence to the fact that hot steel is more easily deformed than cold steel. It is not precisely known when man first used the principle of rolling. Although Leonard da Vinci (1452-1519) entered sketches of hand-driven rolling mills in his drawings, recorded scientific studies\(^1^2\) of the rolling process started toward the end of the 18th century. The subsequent development of rolling technology is closely linked to the progress of iron and steel making on the one hand, and to the knowledge of drive systems on the other.

The economic rolling of excellent quality products is the ultimate goal of modern rolling technology. To achieve this goal, it is imperative to predict the properties of rolled products, and based on this prediction, to design the product’s chemistry, and rolling schedule, and control the rolling process for quality assurance. During steel rolling, complex thermomechanical and metallurgical phenomena occur, and the interaction of these phenomena with one another further complicates the analysis/quantification of the rolling process. However, with the aid of recent advances in physical metallurgy, rolling technology, thermomechanical processing, and computer engineering, the steel industry is now geared towards a new age of computer-aided prediction and control in rolling mills as shown in Figure 1.1\(^3\). The aims of computer-integrated
manufacturing are to increase productivity, reduce manufacturing cost and improve product quality through on-line prediction and control.

In the conventional cold charge rolling (CCR) arrangement, continuously cast cold steel slabs (~250mm thick) are conditioned (scarfing, grinding, cutting etc) at room temperature and charged into a reheat furnace which raises the slab temperature to about 1250 °C prior to rolling. After 2-4 hours in the furnace, the hot slab is descaled and fed to the roughing mill for first-stage reduction (~ 30-40 mm thick). The resulting strip is descaled again before its entrance to the finishing mill where it is reduced to its final gauge; the coil is then cooled on the run-out table and coiled.

Considerable strides have been made in producing defect-free slabs from continuous casting, permitting direct linking of continuous casting and hot rolling. This practice reduces production time and results in energy savings, improved performance, lower inventory costs and shorter delivery times. The first stage of this development involved hot charging conventionally cast defect-free slabs into a reheating furnace followed by hot rolling. In the second stage, direct rolling techniques emerged in which hot slabs were delivered to hot-strip mills atrollable temperatures from the continuous casting machine with only a quick temperature homogenization in between. It was shown from a survey of steel plants\textsuperscript{4} in 1985 that about 55 pct. of all slabs for hot rolling (strips and plates) in Japan were hot charged (300-500 °C average temperature) from continuous casting resulting in slab reheating energy consumption of 960 MJ/t, while in Europe where less than 30 pct. of similar slabs were hot charged, the slab reheating energy consumption ranged from 1069 to 2770 MJ/t.

Despite its obvious advantages, direct linking of continuous casting to hot rolling has been progressing slowly due to the difficulty of producing completely defect-free slabs, securing high slab temperature and effective production scheduling that will enable quality to shift from physical
inspection and slab conditioning to quality assurance. Furthermore, the distance between the continuous caster and the hot rolling mill was too far in the old integrated steel plants. To overcome these problems, major modifications have been made to both the continuous casting and slab delivery techniques\textsuperscript{5,6}. High slab temperature has been attained through high speed casting, reduction of the secondary cooling zone to ensure that the end point of solidification (crater end point) is as close as possible to the exit of the caster, rapid slab delivery from remote continuous casters, and slab insulation between the caster and the rolling mill. Surface defects have been reduced by efficient mold level control, optimized mold powder lubrication and soft-cooling with air-mist systems. The occurrence of internal cracks has been minimized by multipoint straightening, roll pitch reduction and the compression casting technique. Even with these modifications, direct linking is still limited to the lower quality end of steel production since the higher the quality of steel desired, the more stringent the tolerances.

According to Yamamoto\textsuperscript{7}, there are three distinct levels of direct linking of continuous casting and hot rolling. When hot slabs are charged into the reheating furnace below the austenitic temperature (typically in the 400 - 700 °C range), with or without conditioning, it is referred to as hot charge and rolling (HCR). The time elapsed between slab delivery from the caster and rolling is about 10 to 20 hours in this case. If hot slabs are charged into the reheating furnace in the austenitic temperature range (typically in the 700 - 1000 °C range), with or without conditioning, in the same order as they are continuously cast, it is referred to as direct hot charge and rolling (DHCR). For the DHCR technique, the time elapsed between slab delivery from the caster and rolling is about 1 to 2 hours. When hot slabs are rolled directly bypassing the reheating furnace, it is called hot direct rolling (HDR). Temperature homogenization and/or edge reheating in a holding furnace is often employed to ensure that the appropriate rolling temperature is attained. The time elapsed between slab delivery from the caster and rolling is less than one hour.
Hot strip production from thin slab casting is the innovative trendsetting technology being adopted in the next generation of hot-strip mills and is one of the most important developments in steel production in our time. It eliminates the roughing mill and employs hot direct rolling, requiring only a homogenization furnace between the caster and finishing train.

1.2 Rolling of Near-Net-Shape Cast Thin Slabs

With over 5 million tons already being produced and about 11.7 million tons currently being planned in North America alone, strip production from continuously cast thin slabs (50 to 150 mm) has become a revolution in flat steel production. Apart from the obvious incentives of reduced costs and lucrative small production units, this revolution was spurred on by the significant progress in the continuous casting of defect-free slabs, hot direct rolling practice and effective production scheduling that empowered mills to shift from physical inspection and slab conditioning to quality assurance. The Compact Strip Production (CSP) technology developed by Schloemann-Siemag AG (SMS, Germany) is the most widely implemented thin slab technology today.

In the CSP process, shown schematically in Figure 1.2, 50 mm thick as-cast slabs are rolled directly to a final gauge of 2-16 mm in a five or six stand tandem mill. Prior to rolling, the slabs, cast through a funnel-shaped continuous casting mold, are fed directly to an inline roller hearth furnace where the slab residence time is 12-20 minutes to attain a temperature in the range of 1100 to 1150 °C. The furnace not only soaks and equalizes the slab temperature but uncouples the casting speed (2.5 to 6 m/min.) from the mill entry speed (15 to 36 m/min.); it also creates a buffer during roll change and rolling mill delays. Depending on the finishing temperature, mill speed and the required coiling temperature, the hot strip is cooled on the runout table by arrays of top and bottom cooling headers in conjunction with air convection and radiation until it is wound in the downcoiler. In CSP rolling, the run-out table length is about 80 metres; the length
of the laminar cooling section is about 40 m in most cases but can vary with the required product mix and target coiling temperature which also determines the number of active headers.

Comparatively, more has been reported on the thin slab casting process than the subsequent rolling operation. Although the new CSP type hot-strip mill may at first appear similar to a conventional hot-strip mill, there are significant differences which preclude direct application of existing mathematical models for microstructure prediction to this operation. Table 1.1 lists some of the important differences between C-Mn strip production by the CCR and CSP processes.\(^9\)\(^{-15}\) The amount of reduction in each CSP mill stand can be up to 2.5 times the equivalent in conventional rolling. The heavy reduction can facilitate austenite grain refinement but will have detrimental effects on the shape of the strip. The slower entry speed to the CSP mill (about a third of the CCR value) reduces the operating strain rate and extends the interstand time for recrystallization and grain growth.

A very important aspect of CSP rolling is the difference in austenite grain evolution from mill entry to the exit. A typical conventional finishing mill operates in the grain size range of about 200 to 20 \(\mu\)m from mill entry to exit, while the CSP mill operates in the grain size range of 1000 to 20 \(\mu\)m. The entry microstructure in the conventional mill originated from reheated austenite (\(\geq\) 500 \(\mu\)m grain size) which has undergone a series of deformation, recrystallization and grain growth in the roughing mill. In the CSP mill, the entry microstructure is virgin austenite from dendritic solidification during continuous casting or in the case of very low carbon content (< 0.5 \%C), austenite that directly transformed from the initial \(\delta\)-ferrite produced by dendritic solidification during continuous casting. This difference in austenite grain evolution has implications for the rolling process in terms of the deformation resistance and microstructural evolution, especially at the first pass. The grain refinement of coarse, as-cast austenite has not been studied comprehensively and is bound to be different from that of reheated or recrystallized
austenite. The deformation resistance of steels and the tendency towards dynamic recrystallization during deformation, decrease with increasing grain size. Most microstructural parameters such as the peak strain, time for 50% recrystallization and recrystallized grain size are often expressed as a function of the initial austenite grain size before deformation. Even the heat transfer between the strip and roll is affected through heat generation (by deformation and friction) which in turn depends on the deformation resistance. Large austenite grain size as well as ferrite films, precipitates and non-metallic inclusions have been identified as the major microstructural features affecting the hot ductility of steel in various temperature regimes. Cracking problems during the unbending stages of continuous casting and during rough rolling of hot charged slabs have been attributed in part, to coarse austenite grains. Therefore, the finding that the admissible reduction, particularly in the first pass, can be limited by edge cracking during CSP rolling is not surprising.

Furthermore, the report that equivalent properties can be obtained with leaner chemistries in the CSP type process could have ramifications for future production of hot rolled steel. For example, Nucor claims that its CSP AISI 1008 steel (0.10% C max.) can match any CCR AISI 1015 steel (0.18% C max.) in terms of formability, tensile and yield strengths. More recently, Nucor is substituting Nb-HSLA with V-Nb-HSLA containing lower carbon and higher nitrogen levels due to the difficulty of producing defect-free Nb-HSLA. Finer grain size, higher levels of residuals (mainly Ni and Cr), and AlN and MnS precipitation have been proposed to account for the relatively higher strength of CSP C-Mn strip. In the case of microalloyed grades, the precipitation phenomenon and its consequent effect on grain refinement and mechanical properties are known to be different for direct rolling when compared to CCR.

The lucrative rolling of high quality strip from thin slab casting can only result from a complete understanding, and predictable control of, the complex thermomechanical and
metallurgical phenomena involved. Hot direct rolling is a relatively new technology, exhibiting some operational and metallurgical differences from cold-charge reheat rolling. The optimization of hot direct rolling and its extension into the high quality end of flat steel production is ongoing. The CSP hot strip mill not only employs hot direct rolling but also eliminates the roughing process, requiring that as-cast slabs be converted into quality hot band coils with five or six stand tandem rolling. Furthermore, there is the prospect of producing thin hot band coils (≤ 1 mm thickness) on CSP mills to compete with conventional cold rolled products. This can only be achieved by precise control of all hot rolling parameters to attain the stringent tolerances prescribed for cold rolled products.

The fact that such important rolling issues as the homogenizing temperature, descaling, number of stands, permissible reduction per stand, rolling speed and runout table cooling are still being experimented upon by various CSP mills, points to the urgent need for a careful study of the process in order to establish optimum operating windows for CSP rolling. The temperature attained in the tunnel furnace, coupled with the residence time of the slab in the furnace impacts on a host of operational and physical phenomena: precipitate redissolution, grain growth, scale morphology and growth. Scale morphology and growth determine the optimal design of the descaling unit. The number of roll stands has evolved from four to five and now six stands, with one of the new mills now implementing a seven stand operation. The reduction per stand, particularly in the first and last stands has been changed considerably in an effort to reduce edge cracking and shape problems. Improper choice of these parameters could lead to undesirable temperature gradients and inhomogeneous deformation, and unacceptable variations in strip width, gauge, microstructure and shape. On the runout table, cooling behavior is influenced by gauge, strip shape and flatness through their effects on the behavior of water on the strip surface. Without precise control of runout table cooling, even the strips that are rolled within the target
tolerances (usually specified for gauge, width, finishing mill temperature and profile) may result in unacceptable final microstructure and mechanical properties.
Table 1.1 Comparison of CSP and CCR processes for the production of C-Mn steel strips.

<table>
<thead>
<tr>
<th>Prior to Rolling</th>
<th>Cold Charge Rolling (CCR)</th>
<th>Compact Strip Production (CSP)</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Steel Chemistry</strong></td>
<td>Steel from blast furnace-BOF route followed by ladle refining. The levels of trace elements such as Al,N,Cu,Sn,Pb,Cr,Ni and Mo are easily controlled.</td>
<td>Steel from EBT-EAF remelting of scrap/DRI followed by ladle refining. Higher levels of Al,N,Cu,Sn,Pb,Cr,Ni and Mo utilized.</td>
</tr>
<tr>
<td><strong>Continuous Casting</strong></td>
<td>Longer spray zone, slow solidification rate, centerline segregation and coarse as-cast grain size. AlN precipitation depends on thermal/unbending strain. AlN precipitation leads to grain boundary embrittlement and transverse cracks during casting.</td>
<td>Short spray zone, fast solidification rate, less segregation and finer as-cast grain size. Higher thermal strains enhance strain-induced AlN precipitation. Grain boundary embrittlement and transverse cracks more likely. Higher Cu,Sn and Pb can lead to hot shortness during casting.</td>
</tr>
<tr>
<td><strong>Slab Reheating</strong></td>
<td>Characterized by α-γ transformation, heating up to, and soaking at ~1250 °C for 2-3 hours. Extensive grain growth and dissolution of precipitates. Surface layer (including shallow surface defects) converted to scale and subsequently removed by descaling.</td>
<td>As-cast coarse austenite. Homogenized for 12-20 min. at 1100-1150 °C. Limited grain growth and dissolution of precipitates from casting. Thin adhesive scale. Surface defects remain. Cu, Cr and Ni affect scale structure and growth.</td>
</tr>
</tbody>
</table>

Rolling and Runout Table Cooling

<table>
<thead>
<tr>
<th>Entry Microstructure</th>
<th>Reheated austenite at ~1250 °C with finishing mill entry grain size ≤ 250 μm.</th>
<th>As-cast austenite at ~1100-1150 °C with ~0.6-1.4 mm grain size.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Entry Speed</td>
<td>~ 0.8 - 1.6 m/s into the finishing mill.</td>
<td>~ 0.25 - 0.6 m/s leading to relatively lower strain rates.</td>
</tr>
<tr>
<td>Reduction</td>
<td>High total reduction (250 to 2-10 mm) in 10-12 passes. Roughing ensures large reduction per pass while finishing controls the final γ-grain size and shape.</td>
<td>Lower total reduction (50 to 2-10 mm) in 5-6 passes. Desired microstructure and shape must be achieved as the high reductions are applied.</td>
</tr>
<tr>
<td>Recrystallization and grain growth</td>
<td>Final γ-grain size determined by recrystallization at short interstand times in the last passes of the finishing mill. Modest reduction per pass coupled with high strain rate encourages static recrystallization. Limited grain growth.</td>
<td>Final γ-grain size determined by recrystallization at relatively longer interstand times. High reduction per pass coupled with relatively low strain rate may encourage dynamic recrystallization. Initial coarse grain size retards recrystallization. Relatively more time for grain growth.</td>
</tr>
<tr>
<td>Precipitation</td>
<td>Al and N in solution depends on redissolution of AlN during reheating. AlN precipitation has not been an important consideration in C-Mn steel rolling.</td>
<td>AlN precipitation may be important; it could retard recrystallization, refine grain size and contribute to precipitation hardening. Cu,Sn and Pb can lead to checking during rolling.</td>
</tr>
<tr>
<td>Transformation</td>
<td>Affected by γ-grain size, cooling rate and retained strain from incomplete recrystallization in the last stands which operate at low strains.</td>
<td>Affected by γ-grain size, cooling rate and AlN precipitation. High reduction per pass ensures complete recrystallization at all interstand locations except the first where coarse grain size may lead to partial recrystallization.</td>
</tr>
<tr>
<td>Mechanical Properties</td>
<td>Determined by chemistry, ferrite grain size, ferrite fraction and free nitrogen.</td>
<td>Determined by chemistry, ferrite grain size, ferrite fraction and free nitrogen. Fine AlN precipitates during rolling and transformation enhances strength. Cr,Ni and Mo can increase yield strength to undesirable levels when enhanced formability is desired.</td>
</tr>
</tbody>
</table>
Figure 1.1 Schematic representation of a conventional hot strip mill showing the concept of integrated model as a tool for process control and optimization.
Figure 1.2  Schematic of Compact Strip Production (CSP) technology.
Chapter 2

LITERATURE REVIEW

The flat rolling process has benefited from a wide variety of experimental and modeling techniques. Various reviews on the different aspects of conventional cold charge strip rolling - deformation, heat transfer, microstructure, and final mechanical properties have been published. Compared to CCR, there are only a handful of references on hot charging and direct rolling of steel slabs. In the case of strip production via the thin slab casting route, emphasis has been placed on the casting process. Very little attention has been paid to the rolling process even though it has been well documented, albeit qualitatively, that the microstructure, mechanical properties and rolling schedules of strip from thin slab casting are largely different from conventional strip rolling.

In this chapter, the review of the major components of flat steel rolling (thermal, deformation and microstructural evolution), the room temperature mechanical properties of as-rolled steel and the effect of steel composition on rolling behavior and final properties will be presented. The few literature available for hot rolling of thin-cast slabs will be specifically highlighted.

2.1 Thermal Aspects of Flat Steel Rolling

Mathematical modeling has been widely adopted to characterize the thermal phenomena during hot rolling. The accuracy of model predictions depends on the capability of the model to simulate the physical phenomena such as the roll-strip interface and boiling regimes during water cooling. Radiation to the environment, convection to descaling and backwash sprays, and heat conduction to the work rolls have been considered to be the principal modes of heat loss during steel strip rolling. On the other hand, heat is gained by the strip as a result of the mechanical energy due to deformation and friction being converted to heat as well as from any
exothermic transformation. The thermal history of the strip depends on the balance of heat losses and heat gains, and exerts a profound influence on the deformation behavior, metallurgical changes and final properties of the rolled material.

During hot charging, the charging and rolling start temperature has been found to exert a marked influence on the metallurgical changes and mechanical properties of Si-Mn and Nb-bearing steels. Efforts have been directed at the maintenance of rollable temperature before the entrance of rolling mills in the hot direct rolling process; and coil property variations have been traced to inadequate temperature control which often results in mixed grain size. In the case of strip rolling from thin-cast slabs, such unique features as high reduction per pass, slow slab entry speed and deformation of as-cast grain size, are bound to influence the heat balance in the CSP mill. Furthermore, the thick scale from the casting process which undergoes very little reheating changes the dynamics of descaling and amount of heat removed from the slab.

2.1.1 Heat Loss to the Rolls

One important physical phenomenon during hot rolling is the nature of the roll/strip interface. The contact between two surfaces in relative motion often exhibits a combination of conforming, nonconforming and clearance gap at various locations. It has been pointed out that the contact points between two surfaces offer paths of lower resistance for heat flow in comparison to adjacent regions where heat transfer occurs by conduction through air gaps, the dimensions of which depend on the surface topography of the roll and strip.

It is important to note that due to the rapid rebound in surface temperature upon exit from a roll stand, interstand temperature measurements cannot effectively detect the extent of roll chilling or assess the accuracy of the roll-gap heat transfer coefficient. This observation emphasizes the importance of conducting separate measurements to characterize the roll-bite heat transfer coefficient. When an average through-thickness temperature is employed in the roll
bite, the chilling effect of the work roll, which results in a higher strength and a different structure at the surface and subsurface region, and increased roll forces required for deformation is then neglected. Hollander\textsuperscript{21} estimated that heat conduction to the work rolls is approximately 38\% of the total heat lost from the steel strip during rough and finish rolling.

The roll-gap interface heat-transfer coefficient is influenced by many factors, such as the percent reduction, rolling temperature, roll speed, the roll and rolled material, and their roughness. Using measured temperatures in the surface and subsurface regions, several researchers have attempted to quantify the heat-transfer coefficient at the roll-gap\textsuperscript{19,20,31-34}. It has been shown that the heat-transfer coefficient varies with contact time in the roll-gap\textsuperscript{19}. In a recent work, it was shown that the mean heat-transfer coefficient at the roll-gap can be expressed as a linear function of mean roll pressure in the form\textsuperscript{20}:

\[
h = a + b\bar{p} \tag{2.1}
\]

This finding is in accordance with the results of Fenech et al.\textsuperscript{33} who developed a model to simulate heat transfer between two stationary surfaces and determined that the heat-transfer coefficient is a function of contact area ratio. The contact area ratio can be transformed to its pressure equivalent using the Pullen and Williamson approximation\textsuperscript{35}. Hlady et al.\textsuperscript{34} further showed that the heat-transfer coefficient can be expressed as a power function of not only the mean rolling pressure but also the flow stress ($\sigma_f$), effective interfacial thermal conductivity ($k'$) and a roughness parameter (c) in the form:

\[
h = c k' \left( \frac{\bar{p}}{\sigma_f} \right)^n \tag{2.2}
\]

The existence of an oxide scale and/or a lubricant film on the surface of the strip introduces an additional thermal resistance between the roll surface and the strip since their respective thermal conductivity is much less than steel. It has been shown\textsuperscript{36} that during rough rolling, the
thickness of interstand secondary scale is approximately 90 μm, while in finish rolling, which is conducted at lower temperatures and shorter times, the thickness of the secondary scale is likely to be of the order of 20 μm. Hence, it is common to install descalers between stands in the roughing mill but not in the finishing mill. Although the thermal resistance of the oxide scale is small, it is nonetheless comparable to the interfacial and chill zone resistances respectively. Therefore, it might be important in some cases to include oxide formation in the interfacial phenomena at the roll gap especially for rough rolling.

In the CSP process, the higher reduction per pass will lead to higher roll pressure and consequently, increased heat transfer coefficient. The thicker secondary scale resulting from slow slab entry speed will increase the thermal resistance of the oxide scale. The as-cast coarse austenite will reduce the deformation resistance, thereby increasing the heat transfer coefficient in accordance with equation 2.2.

2.1.2 Heat Loss to the Water Cooling Systems

Steelmaking processes invariably use water cooling systems to control temperature, product quality and productivity. In steel rolling, water cooling systems are used to descale the steel strip, to cool the rolls, as interstand cooling and finally to cool the strip during austenite to ferrite transformation at the cooling banks of the runout table. Sprays, laminar jets and water curtains are the most commonly used cooling systems in steel rolling applications. Figure 2.1 illustrates the water cooling systems in the final stand of a finishing mill and the runout table. A comparison of the various cooling systems employed in steel rolling has been published by various authors. Figure 2.2 compares the heat transfer coefficient and cooling rate of water curtain, laminar jet and spray cooling. It is seen that water curtain is the most efficient cooling system while spray is the least efficient of the three systems. The difference in cooling rate between the
three cooling systems decreases with increasing strip thickness and decreasing specific water application\textsuperscript{38,39}.

Before entering the roughing or finishing mill, the transfer bar is usually descaled by a bank of high pressure water sprays. Sprays are preferred for descaling because minimal slab cooling combined with high pressure is good for efficient scale removal\textsuperscript{38}. During rolling, roll cooling by water sprays is employed to reduce the incidence of excessive firecracking, rapid roll wear and the consequent poor strip profile, due to the thermal cycling of the rolls. The thermal cycling results from the alternating contact between the rolls and the high-temperature strip and cooling water. Interstand cooling by water sprays is sometimes used to control the strip temperature between stands. Sprays are preferred in this case because the relatively low cooling efficiency allows for quick temperature homogenization before the next pass.

At the runout table, laminar jets or water curtains are applied to the strip surfaces (top and bottom) from selected active headers and banks, in order to achieve the desired coiling temperature, microstructure and mechanical properties. Laminar cooling is preferred in the runout table because it appears to be the best compromise between high specific cooling capacity and uniformity of strip cooling. Water curtain is used when very intense cooling in a short cooling zone length is desired. Spray cooling headers have been employed when soft cooling is required or when head-end deformation becomes a problem, particularly for thin strips (<2.0 mm) which are easily deformed under the impact of laminar jets or water curtains\textsuperscript{40}.

The heat transfer in water cooling systems is governed by the well known boiling curve as depicted in Figure 2.3 for a saturated liquid\textsuperscript{41}. The figure indicates the four different heat transfer regimes experienced by the surface of a hot metal in contact with water: film boiling, transition boiling, nucleate boiling and single-phase liquid cooling. In the film boiling regime, a thermally insulating layer of vapor quickly develops immediately after the water impingement at high
temperatures, resulting in slow cooling. Below the minimum heat flux point, the metal surface exhibits partial contact between water and vapor in the transition boiling regime, leading to fast cooling. In the nucleate boiling regime, the entire hot metal surface is in contact with water, and consequently accelerates the cooling process in conjunction with bubble formation. Pure liquid convection takes place in the single-phase liquid regime. It is noted that the heat transfer associated with each of the boiling modes is dependent on a myriad of factors (water temperature, surface condition, steady-state or transient, material, etc) and Figure 2.3 represents only a special case\textsuperscript{41}.

Dimensionless correlations and semi-empirical equations have traditionally been used to characterize the heat-transfer coefficient for water cooling systems and have been reviewed in various publications\textsuperscript{42-45}. Table 2.1 lists typical heat transfer correlations for each boiling regime for a spray cooling system\textsuperscript{46}. Further details of the fundamentals of heat transfer in water cooling systems, particularly the runout table, are currently being investigated at UBC\textsuperscript{47}.

Optimized descaling design has been an issue with the CSP mill, although it has been established that an efficient descaling unit should have just enough descaling power to effectively remove scale and operate with a minimum amount of slab/strip cooling during descaling\textsuperscript{30}. Rough surface, dimensional distortion and rolled-in scale have been attributed to inadequate descaling\textsuperscript{30,48}. Although the CSP slab is 65 to 100 pct. thicker than the typical conventional cold charge (CCR) transfer bar, the temperature loss from descaling has been estimated to be about 50 to 60 pct. higher when the descaling practice is the same\textsuperscript{30}. This slab overcooling results from the low CSP mill entry speeds. CSP mill operators claim that the use of very high pressures (up to 450 bar) utilizing small nozzles with reduced inter-nozzle distance and the distance between nozzle and strip, resulted in fewer multiple scale-affected zones\textsuperscript{8,49-51}. However, the optimum combination of pressure, water flow rate and spray coverage length remains controversial.
Beyond the descaling unit, growth of the scale ahead of F1 and F2 stands, which is expected to be thicker than in conventional hot strip mills because of the low CSP entry speeds, also results in scale-affected zones, rough surface and roll wear\textsuperscript{36}. Interstand cooling after the first two passes has been successfully utilized to reduce interstand scale growth. The interstand cooling ensures that a significant fraction of the strip surface between two stands is covered by water, thereby reducing the surface temperature and direct steel/air contact.

### 2.1.3 Heat Loss due to Radiation and Air Cooling

In regions where the strip is not in contact with water or rolls (interstand regions, between water banks in the runout table), the strip loses heat by radiation and by convection to the surroundings at ambient temperature. The Stefan Boltzmann law is used to quantify the radiation component while some convection correlation is used to quantify the second component. Hence, the following equation has been used\textsuperscript{43}:

\[ q = q_{\text{r}} + q_{\text{conv}} = \varepsilon\sigma(T_s^4 - T_a^4) + k\Delta T / x \cdot f(Pr, Re) \quad (2.3) \]

The second term of the above equation is based on an empirical relationship for turbulent flow over vertical plates. The emissivity is mostly assumed to be constant but it has been shown to be a function of temperature\textsuperscript{52} and strip thickness\textsuperscript{53}. It is noted that the presence of an oxide scale and/or a lubricant film on the surface of the strip results in a different emissivity value. Hence, the primary scale from casting that survives on the thin slab surface until descaling and the tangible secondary scale ahead of the first three stands in the CSP mill, will influence the emissivity at these locations.

### 2.1.4. Heat Gain from Deformation

There are two sources of heat generation at the roll bite - (i) mechanical energy from plastic deformation and (ii) friction between the rolls and strip. The deformation heat is distributed
throughout the strip being deformed while the frictional heat is limited to the surface layers of the strip and rolls. The heat due to plastic deformation (in Joules) can be written as:

\[ q_{\text{def}} = \eta \int \sigma \varepsilon dv \approx \eta \frac{m\sigma_f}{\rho} \ln(h_1 / h_2) \]  

(2.4)

where \( \eta \) is the fraction of plastic work that is converted into heat and \( J \) is the mechanical equivalent of heat. The approximation is often used where rigorous calculation of the strain and strain rate is not required\(^{53}\). The accuracy of the heat of deformation is dependent on the precision involved in the determination of the flow stress as a function of operating mill conditions. Temperature gradient and inhomogeneous deformation will change the value of deformation heat at various locations due to their influence on flow stress.

The frictional heat will depend on the perceived roll-strip interface friction conditions (see Table 2.2). In the case of no-slip or Coulomb friction assumption, the frictional heat can be written as\(^{54}\):

\[ q_{\text{fric}} = C_0 \mu p dA \]  

(2.5)

In this case, the accuracy of the frictional heat is dependent on the precision involved in the determination of the friction coefficient (\( \mu \)), mean pressure (\( p \)) and the relative velocity (\( v \)). It has been suggested that the frictional coefficient, \( \mu \), increases with an increase in temperature as follows\(^{55}\):

\[ \mu = 4.86 \times 10^{-4} T - 0.0714 \]  

(2.6)

The roll pressure is also known to vary over the length of contact, displaying the well known friction hill\(^{53}\). The relative velocity between the strip and work roll also varies along the arc of contact. At the neutral point, the velocity of the roll and the strip are equal, while towards the entry side there is backward slip and forward slip towards the exit side. Furthermore, since the
roll temperature is much lower than that of the strip, the distribution ratio of the frictional heat is considered to be 60-70 percent to the rolls and 30-40 percent to the strip.

In the CSP mill, the higher reduction per pass is expected to increase the heat of deformation. The thicker secondary scale resulting from slow slab entry speed might influence the friction condition at the first few passes. The as-cast coarse austenite will reduce the deformation resistance, thereby affecting both the deformation and frictional heat.

2.2 Deformation Mechanics of Flat Steel Rolling

Deformation of steel in the roll bite provides the driving force for microstructural change and alters the shape and gauge of the rolled product. The driving force for structure-modifying metallurgical phenomena, such as dynamic or static recovery and recrystallization, is the dislocation density and associated strain energy imparted by the deformation, the distribution of which is characterized by the reduction per pass, the rolling speed, the temperature distribution, and the steel composition. These variables also strongly influence the roll forces which, together with the design of the roll stand (bending, deflection) and the associated control systems, have a tangible effect on gauge variations and the shape of the product.

Experimental techniques which vary from physical modeling using photoelastic or photoplastic materials, pressure pin method, and the caustics method have been employed to study deformation during rolling. These studies were basically directed towards the observation and measurement of flow patterns within the workpiece and have been able to provide information on strain distribution, pressure distribution, friction condition, and spread. Experimental techniques have not been able to provide information on the comprehensive description of deformation behavior of the workpiece.

Mathematically, a complete solution of the rolling problem must satisfy three physical conditions: (a) the conservation of energy, momentum and volume (continuity), (b) the yield
criterion, and (c) the stress-strain relations (constitutive equations). However, such a complete solution is difficult to obtain due to the complexity of adequately describing the boundary conditions, the material behavior and the interactions between temperature, structure and metal flow. Employing various simplifications, analytical solutions such as the homogeneous slab method, the inhomogeneous Orowan and Sims methods, the similarity theory, the extremum principle and the slip line field method have been utilized to obtain reasonable estimates of force and rolling torque, metal flow, pressure and stress distribution. However, for more complete results incorporating work hardening, with rate and thermal effects, recourse to numerical analysis is necessary. The finite-element method (FEM) has been found to offer obvious advantages over these analytical techniques due to the versatility of FEM in obtaining detailed solutions of the mechanics in a deforming body, namely, velocities, shapes, strains, stresses, temperatures and contact pressure distribution.

Numerical methods in rolling commonly utilize the energy method based on the principle of work equivalence which states that the work done by the external force is equal to the deformation energy of the metal on which the force is exerted.

\[
\int_\Omega \left( \sigma \cdot \varepsilon \right)^T d\Omega = \int_\Omega \left( \partial \mathbf{u} \right)^T \mathbf{F} d\Omega + \int_\Gamma \left( \partial \mathbf{u} \right)^T \mathbf{F}_b d\Gamma \tag{2.7}
\]

where \( \Omega \) is the flow domain, \( \Gamma \) is the boundary, \( \mathbf{u} \) is the velocity, \( \mathbf{F} \) and \( \mathbf{F}_b \) are the domain and boundary forces.

Several different formulations, based on the FEM have been developed for hot and cold rolling. An excellent overview of the application of the finite-element analysis to hot steel rolling is provided in two recent reviews. The principal difference in the methods stems from whether a Lagrangian or Eulerian frame of reference is employed. The former is based on material coordinates and employs the Prandtl-Reuss equations to relate the deviatoric stress to the strain...
increment (displacement method), whereas the latter is for a coordinate system fixed in space and relates the flow stress to the rate of deformation (flow-formulation method). The success of the different methods depends very much on the material behavior assumptions and other simplifications such as the characterization of frictional boundary condition which influences the redundant shear and homogeneity of deformation. Material behavior is often assumed to be some combination of elastic, plastic and viscous features; the Eulerian formulation neglects elastic effects. In the treatment of friction at the roll-strip interface, constant frictional stress, no-slip and coulomb friction conditions have been applied as listed in Table 2.2.

A comparative study of three FEM formulations (Eulerian, Lagrangian and Updated Lagrangian methods) and their applicability to flat steel rolling has recently been published. It was shown (Figure 2.4) that the through-thickness strain profiles predicted by the three models were almost identical over the central region of the strip and differ close to the surface because of the differences in the treatment of friction at the interface. Strain in-homogeneity was found to increase with increasing friction and strip thickness. Good agreement between model predicted and measured roll forces (less than 11 pct. deviation) were obtained for the Eulerian and Updated Lagrangian methods for a friction coefficient of 0.25. The Lagrangian model showed more disagreement (up to 20 pct. deviation) with measured values at an optimal interface friction factor of 0.85.

No comprehensive deformation modeling has been reported for strip rolling via the thin slab casting route. The higher reduction per pass is expected to influence the distribution of strain and strain rate across the strip thickness. On the other hand, the initial coarse as-cast austenite will lower the deformation resistance and could contribute to higher roll force in the succeeding stand through strain accumulation from incomplete recrystallization since coarse grain size retards recrystallization rate. It has been shown that the roll force per unit width in the second pass of a
roughing mill can increase by about 12 percent when the austenite grain size changes from 100 to 1400 \( \mu \text{m} \) due to work hardening from incomplete recrystallization in the previous pass\(^6\). The effect of coarse grain size on the homogeneity of deformation is not clear although it has been shown that the presence of coarse grain structure can introduce measurable strain inhomogeneity during the deformation of metals\(^5\).

It has been suggested that for good surface quality in direct rolling from thin slab casting, a new concept of rolling machine design with emphasis on higher torque transmission efficiency is required\(^6\). The higher torque transmission efficiency will allow for higher reduction per stand with a decreased work roll diameter, thereby lowering the rolling force and torque requirements. This will translate into a reduction in the specific power consumption per ton of rolled product. The reduction in work roll diameter can be achieved by a new design of flexible coupling, transmitting torque through the combined action of rolling and sliding friction. It was claimed that such a design will lead to a reduction of nonuniformity in work roll rotation, resulting from misalignment of the drive and work axes, thereby contributing to an improvement in strip surface quality due to a decrease of roll slippage on rolled products\(^6\).

The accuracy of any deformation model depends largely on how well it captures the deformation behavior of the material and rolls, the friction conditions and the interaction between deformation, temperature and structural changes. Therefore, the adequate characterization of the flow stress of the material being rolled is of uttermost importance.

### 2.2.1 Flow Stress Determination

Most of the theoretical knowledge of the distinct mechanisms that contribute toward the plastic behavior of metals at various temperatures were garnered from creep studies of pure metals\(^6\). These mechanisms involve dislocation motion (climb, slip, slide) and diffusion of
vacancies through the lattice or grain boundary. However, the present state of classical phenomenological knowledge does not allow for accurate prediction of the elastic-plastic response of real materials, within a sufficiently wide range of variation of the geometric and physical parameters \(^6^9\). Therefore, experimental determination of flow stress and true strain for a range of strain rates and temperatures is still the preferred route in most cases.

The high temperature mechanical behavior of steel can be studied with tension, torsion and compression tests. Tension tests are usually performed with Instrons and other universal testing machines but the application of the data generated from these tests is limited by the anomalous rise in localized strain at the onset of necking. Torsion machines have the major advantage of high strain (up to 4) and strain rates (up to 20/s) at the surface which allows for a better estimate of the combined effect of high strains and high cooling rate on microstructure since it is easy to achieve high cooling rates at the surface. The major disadvantage of torsion testing is the steep gradients of stress, strain and strain rate across the radius of the specimen, making it difficult to correlate bulk microstructure to testing conditions. Compression tests can be either axisymmetric or plane strain and the strain associated with the applied load can be evaluated from the change in specimen cross section (usually a change in height or diameter is measured). Compression tests allow for a better distribution of deformation within the bulk of the specimen and higher heating rate (resistance heating) than torsion tests but are limited by low strains and strain rates, and by barrelling in the case of axisymmetric tests.

The measured true stress-strain curve of steels in the austenite range reflects the competition between strain hardening and dynamic softening processes, and exhibits one of the shapes shown in Figure 2.5\(^7^0\). In all cases, plastic deformation commences with work hardening where the stress increases with strain due to the rising dislocation density caused by dislocations interacting with each other and with barriers which impede their motion through the crystal lattice, followed
by dynamic recovery which reduces the dislocation density. With further increase in strain, any of the following three changes will be observed:

(i) the flow curve attains a peak or multiple peak values at high temperatures and low strain rates, followed by a steady-state or minimum value as a result of dynamic recrystallization.

(ii) the flow curve attains a steady-state value due to the balance between hardening and recovery at a favourable combination of temperature and strain rate.

(iii) the flow curve continues to increase with strain due to the domination by hardening when recovery rate is slow.

To develop mathematical models of the hot rolling process, it is necessary to convert the experimental data into constitutive equations that can predict the flow stress for a given set of rolling conditions. The development of these equations is by no means an easy task since it must capture the changing nature of the three processes of work hardening, dynamic recovery and recrystallization; and the effect of other variables on them. The flow stress of a material is generally dependent on such parameters as composition (C), temperature (T), strain (ε), strain rate (ε), grain size (d) and previous history. Quantitatively, the flow stress can be expressed as follows:

\[ \sigma = A.f(C, T, \varepsilon, \dot{\varepsilon}, d) \]  \hspace{1cm} (2.8)

where A incorporates the previous history such as prior deformation.

Although there is a body of data in the literature as listed in Table 2.3\textsuperscript{70-71}, no universal correlations exist relating stress, strain, strain rate, temperature and grain size for different grades of steel. It is observed that none of the equations explicitly account for all the important variables that contribute to constitutive behavior of metals. The common approach has been to incorporate the missing variables via the fitting parameters. For example, by expressing the
parameters \((Q,n,A)\) in the hyperbolic sine equation as functions of strain, the complete stress-strain curve can be obtained for a given grain size and steel composition\(^72\). In general, the stress at a given strain increases with increasing strain rate and decreasing temperature. Dynamic recrystallization is enhanced by low strain rates and high temperature. There are only a few literature on the effects of composition\(^70,73\) and grain size\(^73-76\) on flow stress of austenite. The measured influence of various elements on the flow stress of steel at high temperatures is depicted in Figure 2.6\(^70\). The specific effect of carbon has been included in the constitutive equation in some cases\(^70\). Recently, it has been suggested that the apparent activation energy for deformation in C-Mn steels is dependent on the carbon, manganese and silicon contents as follows\(^73\):

\[
Q_{\text{def}} = 282.7 + 92.4[\%C] + 6.57[\%Mn] + [\%Si]
\] (2.9)

In the case of grain size, only a few authors have included it as a variable in the flow stress equation\(^73,76\) although it is well documented from creep studies\(^69\) that the normalized flow stress (ratio of stress to shear modulus, \(\sigma/G\)) can be expressed as polynomial function of normalized grain size (ratio of grain size to Burgers vector). The effect of grain size on flow stress has been associated with strain hardening-recovery theory of creep and utilizes the concept that, at high temperatures, fast grain boundary transport generates large dislocation meshes in and near grain boundaries before spreading into the grain interior\(^74\).

In the CSP mill, the constitutive equation is bound to be affected by the wide range of grain size from entry to exit as well as the initial solidification microstructure. Solidification morphology due its preferred dendritic orientation and the presence of segregation, can affect local stress distribution when compared to randomly orientated grains of reheated austenite\(^77\).
2.3 Microstructural Changes during Flat Steel Rolling

Microstructural models are geared towards the quantitative description of the structural changes during hot rolling and cooling in the runout table, such as recrystallization, grain growth, precipitation and austenite decomposition.

2.3.1 Recovery and Recrystallization

During deformation, strain hardening and dynamic softening take place. Strain hardening describes the phenomenon by which the strength of material increases during deformation due to dislocation pile-up or blocking of dislocations by obstacles. Dynamic softening can have two components: dynamic recovery and dynamic recrystallization. Dynamic recovery occurs as result of an accumulation and rearrangement of dislocations in such a way that dislocations of opposite signs annihilate each other or rearrange to form cells of relatively low dislocation density often surrounded by boundaries of high dislocation density. Dynamic recovery can also occur by cross slip and climb at high temperatures. As deformation continues, the cellular structure is transformed into subgrains.

Dynamic recrystallization results when new grains are nucleated due to the dislocation density exceeding a critical value, and often occurs when the dynamic recovery is slow thereby permitting the dislocation density to increase to an appreciable level. Experimentally, well developed substructures resulting from dynamic recovery have been observed in dynamically recrystallized microstructures and are the origin of dynamic recrystallization nuclei. The nucleation of dynamic recrystallization is commonly said to begin at a critical strain, $\varepsilon_c$, which corresponds to a critical dislocation density, and involves the bulging of preexisting grain boundaries at low strain rates and growth of the high angle boundary cells formed by dislocation accumulation at higher strain rates. The driving force for the growth of recrystallized nuclei is
the difference in dislocation density in front of, and behind the boundary. Growth of each new nucleus is terminated by concurrent deformation for single peak behavior and boundary impingement for multiple peak case.

Following deformation, restoration occurs and softens the material. Three different processes are responsible for this softening: static recovery, static recrystallization and metadynamic recrystallization. Static recovery is defined as the decrease in density and change in distribution of the dislocations and other defects that take place after deformation, where these changes do not involve the sweeping of the deformed material by migrating high angle boundaries. Metadynamic recrystallization is the continuation of the growth of dynamically recrystallized nuclei after deformation has been interrupted. Static recrystallization involves the nucleation and growth of new grains following deformation and is associated with the migration of high angle boundaries in the deformed material after deformation. As with dynamic recrystallization, it is a dual process of nucleation and growth. The nucleation stage is thermally activated and requires an incubation time before the nuclei become detectable, and takes place preferentially where the local deformation is the highest, that is, on grain boundaries, deformation bands, precipitates and inclusions.

Recrystallization is characterized by the fraction recrystallized \( X_{\text{rex}} \) at a given set of conditions (temperature, strain and strain rate) and time, as well as by the recrystallized grain size \( d_{\text{rex}} \). To quantify these two parameters \( (X_{\text{rex}} \text{ and } d_{\text{rex}}) \), a combination of flow stress and metallographic analysis is often employed. For dynamic recrystallization, the fraction recrystallized can be obtained from single-hit isothermal tests performed under conditions where the flow stress manifests a clear peak, by assuming that recrystallization is close to completion \( (X_{\text{rex}} \approx 1.0) \) when the flow curve attains a steady-state after the peak. An alternative approach is to measure the dynamically recrystallized fraction from metallographic examination of quenched
specimens. For static and metadynamic recrystallization, double-hit isothermal tests are utilized to determine the kinetics of restoration processes during rolling. This method involves recording the stress-strain curve (ABC) obtained by deformation at a constant strain rate and temperature, as shown in Figure 2.7. This is followed by an isothermal holding time, \( t \), after which the specimen is reloaded and the second flow curve described by curve (EF) is obtained. The restoration is determined from the observed changes in stress and in some cases, from the fraction recrystallized obtained from metallographic examination of quenched specimen.

For static and metadynamic recrystallization, the fraction recrystallized at a given time, \( t \), is expressed by the Avrami equation of the form:

\[
X_{\text{ex}} = 1 - \exp\left(-0.693\left(\frac{t}{t_{0.5}}\right)^{k}\right)
\]

(2.10)

where \( t_{0.5} \) is the time required to attain 50 pct. recrystallization. Similar expressions has been developed for dynamic recrystallization with the applied strain as a variable instead of time:

\[
X_{\text{ex}} = 1 - \exp\left(-0.693\left(\frac{\varepsilon - \varepsilon_{0.5}}{\varepsilon_{0.5}}\right)^{k}\right)
\]

(2.11)

The recrystallized grain size is usually obtained from metallographic measurements. The recrystallization rate and the recrystallized grain size of a given steel is dependent on the prior austenite grain size and the deformation conditions such as temperature, strain and strain rate. Hence, the time or strain for 50% recrystallization (\( t_{0.5} \) or \( \varepsilon_{0.5} \)), the peak strain (\( \varepsilon_p \)) and the recrystallized grain size (\( d_{\text{ex}} \)) are commonly expressed as functions of measurable rolling parameters (temperature, strain, strain rate and initial grain size) as follows:

\[
t_{0.5}, \varepsilon_{0.5}, \varepsilon_p, d_{\text{ex}} = A_t d_0^{m_t} \varepsilon_n^{n_t} \exp\left(\frac{Q_{\text{ex}}}{RT}\right)
\]

(2.12)
The critical strain for dynamic recrystallization is then expressed as a fraction of the peak strain \( (\varepsilon_c \approx 0.85 \varepsilon_p) \). For modeling purposes, the recrystallization equations developed from isothermal tests are applied to actual rolling where the temperature changes continuously by invoking the additivity principle\(^8\). Even though dynamic and metadynamic recrystallization are dependent on the occurrence of a peak in the flow curve, no comprehensive quantitative description of peak occurrence in steels is known to the author.

All the recrystallization equations in the literature are based on steel samples that have been cooled to room temperature and reheated. For hot direct rolling, as practiced in the CSP process, the correct starting point should be freshly cast, hot austenite just after solidification. Differences have been reported between the recrystallization behavior of CCR and hot charged strips\(^2\). For C-Mn steels, Priestner et al.\(^8\) found obvious differences between the two. For instance, at 30% reduction, a temperature of 1170 °C was needed to cause complete recrystallization in Is for the as-cast C-Mn steel, compared to 1000 °C in the reheated sample. Furthermore, the transition from dynamic to static recrystallization was different for the two types of treatment. In the case of microalloyed steel grades, it appears that the microstructure and constitution of the as-cast austenite cannot be reconstituted by reheating a cold slab due to major differences in grain size, microsegregation and precipitation between the virgin and reheated austenite\(^8\). Considerable doubt must exist, therefore, whether microstructural models derived using data from experiments with reheated austenite will reliably predict the recrystallization behavior of austenite during rolling of thin-cast slabs.

### 2.3.2 Grain Growth

Although recrystallization of austenite removes the relatively high internal energy imparted by hot deformation, the structure is still metastable. Further reduction in the overall internal energy occurs by a reduction of the total austenite grain boundary area through growth. Grain growth is
the motion of grain boundaries in a solid material such that mean grain size is increased. By increasing the mean size of grains, the number of grains per unit volume and total grain boundary area must be reduced. The loss of grain boundary area is the driving force for growth since it is accompanied by a liberation of grain boundary energy. Growth of a new grain is always accompanied by the consumption of non-growing grains since grain growth within an object is not associated with a net change in shape or volume. Larger grains grow at the expense of smaller grains such that a criterion for critical grain size for growth is necessary. During grain growth, grain boundary motion can be influenced by particles or precipitates embedded in the matrix and by solute atoms in the matrix.

Grain growth is governed by some geometrical and topological requirements since the grains must always fill a fixed space. In a two dimensional grain network, only two types of changes can occur, (a) neighbor switching and (b) face disappearance. In a three dimensional structure, cell disappearance can also occur. Without pinning, normal grain growth has been found to obey a parabolic law in the form:

\[ d_{g_0} = \sqrt{d_{g_0}^2 + Kt} \]  \hspace{1cm} (2.13)

where \( K = 3\gamma_{gb} \cdot D_{gb} \cdot b^2 / (kT) \)

However, pinning is a common feature of grain growth in steel rolling since precipitation invariably occurs even in plain carbon steels. The quantitative description of pinning has not been fully established. Therefore, empirical power law equations have been commonly used to quantify grain growth as follows:

\[ d_{g_0}^{in} = d_{g_0}^{in} + At \exp\left(\frac{-Q}{RT}\right) \]  \hspace{1cm} (2.14)
where \( m \) ranges from 4 to 10. Data from isothermal tests are usually fitted to the power law and then applied to non-isothermal conditions in stepwise fashion.

Some statistical grain growth models which consider growth by size or shape class have been developed but their application is more complex than the power law and consequently have not been fully exploited\(^5\). Monte Carlo simulations have also been applied to grain growth and provide useful insight into grain growth phenomenon\(^5\).

Mixed grain size has been a problem in the hot direct rolling process when the slab temperature control is not adequate\(^9\). The coarser grains in the mixture often appear near the surface which has been attributed to strain induced grain growth during rolling\(^9\). Grain growth in direct rolling of thin slabs has been reported to be more retarded than equivalent CCR process by the pinning effect of fine MnS and Ti(C,N) precipitates\(^16\).

### 2.3.3 Precipitation

Microalloying elements, such as Nb, V, Ti, and Mo which form carbide/nitride precipitates during the hot working process exert some influence on recovery, recrystallization and grain growth. Grain boundary mobility has been shown to be quite sensitive to the presence of small concentrations of impurities. The speed of migration is drastically reduced by the impurities which segregate to the grain boundaries. Several quantitative treatments of the interaction between grain boundaries and solute atoms have been performed by various investigators\(^88,89\). It has been shown that at high concentrations or low temperatures, the impurities are dragged along by the grain boundaries and the boundary speed is controlled by the diffusion of the impurities behind the boundaries. At low concentrations or high temperatures, the boundary breaks away from its atmosphere and moves faster.

The precipitates encountered in austenite can be separated into three types: (i) those which are not dissolved during reheating, (ii) precipitates formed dynamically during deformation, and
(iii) strain-induced precipitates formed after deformation. Undissolved precipitates have little effect on recrystallization, for they are too coarse. However, both dynamic and strain-induced precipitates can be responsible for the retardation of recrystallization. The interaction between precipitates and recrystallization is often analyzed by recrystallization-precipitation-temperature-time diagram as shown in Figure 2.8 (a). The figure displays recrystallization start and finish curves ($R_s$ and $R_f$) as well as the precipitation start curves with or without strain ($P_{SD}$ and $P_s$). Recrystallization is shifted to longer times ($R_s^P$ and $R_f^P$) when precipitation commences before or during recrystallization. The temperature at which precipitation retards recrystallization is referred to as the temperature of 'no-recrystallization', $T_{nr}$. This temperature ($T_{nr}$) depends on the nature and the amount of alloying elements present as indicated by Figure 2.8 (b). Nb has the most profound effect in retarding recrystallization and this accounts for its selection as the most common microalloying element in hot rolling.

Strain induced precipitation models have been developed for Nb (CN), Ti (CN) and MnS$^{90,93}$. Recently a model based on coupling of thermodynamics, dislocation theory, classical nucleation and growth theories has been proposed to predict the precipitation behavior in austenite during solution treatment and hot working of Nb-Ti bearing steels$^{94}$. In this model, the thermodynamic module calculates the equilibrium fraction of complex precipitates during solution treatment and estimates the chemical driving force of strain induced precipitates from hot worked supersaturated austenite. The nucleation and growth module predicts the time dependence of volume fraction and particle radius of strain induced precipitates. The dislocation theory estimates the change in density of dislocations which act as nucleation sites of strain induced precipitates. Strain-free precipitation readily occurs during transformation or in the α phase because of the low solubility of various elements in ferrite. Although this type of precipitation has been studied, the
quantitative analysis of the precipitation behavior and its interaction with ferrite transformation is rather scarce.

Differences have been observed between the precipitation behavior of direct rolled Nb-bearing steels and CCR samples. Kamada and Hashimoto reported a more homogeneous distribution of fine Nb(CN) precipitates for the direct rolling process when compared to the CCR equivalent where both coarse and fine Nb(CN) precipitates were present. The coarser precipitates in CCR samples were attributed to the undissolved Nb(CN) which lead to a coarser ferrite grain size with a consequent deterioration of the mechanical properties, especially the tensile strength. AlN precipitation has been reported to occur during both casting and rolling in the CSP process. During continuous casting, aluminum nitride precipitates especially below 900 °C and this embrittles austenite grain boundaries and promotes the formation of transverse cracks at the base of oscillation marks under the influence of axial tensile strains. In the case of microalloyed steel grades, co-precipitation of NbCN and AlN extends the low ductility region to even higher temperatures.

A few studies have been conducted on the precipitation phenomena in direct strip rolling of thin-cast slabs. Direct rolling from thin slabs have been found to be characterized by a more homogeneous distribution of fine MnS and Ti(C,N) precipitates than in equivalent CCR steels, leading to enhanced retardation of recrystallization and grain refinement. Owing to short residence times of thin slabs in the soaking furnace and low reheating temperature of 1100 °C, any precipitates formed during casting might not dissolve and therefore, reduce the amount of precipitation during rolling. This will ultimately lead to non uniform distribution of microstructure and consequently, undesirable variation in the mechanical properties of the coil. Full dissolution of solute elements is desired since this allows for uniform precipitation of fine
particles that retards recrystallization, resulting in finer grain size and improved mechanical properties.

2.3.4 Phase Transformation

On the runout table, hot rolled austenite transforms to ferrite, pearlite, bainite and martensite, in order of increasing cooling rate and decreasing transformation start temperatures. Besides the cooling rate, the steel composition and properties of the parent austenite also determine the products of transformation. For example, the amount of pearlite in the transformed microstructure increases with increasing carbon content in plain carbon steels.

The transformation of ferrite and pearlite proceeds by a nucleation and growth mechanism, the nucleation commencing preferentially at austenite grain boundaries (corner, edge and surface). Experimentally, transformation kinetics is determined by constructing temperature-time-transformation (TTT) curves from isothermal tests or continuous cooling-transformation curves (CCT) from continuous cooling tests. Quantitatively, transformation is often characterized by the transformation start temperature ($T_s$), the fraction transformed ($X_t$) and the ferrite grain size ($d_f$). The transformation start temperature is often expressed as a function of composition, cooling rate, austenite grain size and retained strain (if any). The transformation start temperature can be computed by considering nucleation and early growth as follows:

$$c^* - c^0 = \left(\frac{c_f - c^0}{\varphi^{1/2} d_f} \right) \sqrt{\frac{M}{\tau_s}} \left(\frac{c_f - c^0}{c_f - c_a} \right) dT$$  \hspace{1cm} (2.15)

Although various nucleation and growth models have been proposed, it is still common to quantify the fraction transformed with the well known semi-empirical Johnson-Mehl-Avrami-Kolmogorov (JMAK) equation in conjunction with additivity rule:

$$X_{(T)} = 1 - \exp \left( -\frac{1}{d^n} \left( \frac{K_{(T)}^{1/n}}{\tau - dT / dt} \right)^n \right)$$  \hspace{1cm} (2.16)
Theoretically, the ferrite grain size is dependent on the number of grains nucleated per unit area, the nucleation rate and austenite grain size. The influence of austenite grain size have been found to be dependent on the nucleation site, being more pronounced for edge nucleation than for surface and homogeneous nucleation. In practice, the ferrite grain size can be expressed as a semi-empirical function of cooling rate (or transformation start temperature) and austenite grain size as follows:

$$d_a = A(F \exp(Bd_\gamma^9 - E/T))^{1/3}$$

(2.17)

Further details on phase transformation are reviewed in a separate Ph.D. thesis. Besides the effects of tramp elements from EAF steelmaking, the transformation behavior of steels produced from CSP mills is expected to be similar to that of a conventional mill under equivalent runout table cooling conditions and austenite grain size.

### 2.4 Room Temperature Mechanical Properties

The room temperature mechanical properties of steel is determined by the final microstructure and the hardening effects of solid solution and precipitation. There are no comprehensive fundamental-based equations for room temperature mechanical properties since the physical mechanisms by which these properties evolve is not yet clear. For a single phase alloy with homogeneous grain size and random crystallographic orientation, the yield strength is given by the well-known Hall-Petch equation:

$$\sigma = \sigma_0 + k/\sqrt{d}$$  \hspace{1cm} (2.18)

where $\sigma_0$ is the internal stress, $d$ is the grain size and $k$ is a constant.

Unfortunately, as-rolled steel in most cases, is composed of different amounts of various phases (ferrite, pearlite, bainite, martensite), and exhibit a range of grain sizes and preferred orientation. Furthermore, the incorporation of the hardening/softening effects from the many
alloying and tramp elements that are present in every steel is a herculean task. Therefore most available mechanical property equations are based on the law of mixture (aggregation of various phases and elements) with parameters established from experimental data.

Typical equations for yield and ultimate tensile strength are as follows:

\[
YS = 63 + YS_{ss} + k_y X_f d_a^{-1/2} + (360 + 2600[C]^2 X_p) \\
YS_{sw} = 23 Mn + 53 Si + 700 P + 5000 N_x \\
k_y = 15.4 - 30 C + 6.094 / (0.8 + Mn)
\] (2.19)

\[
TS = 237 + TS_{ss} + 7.24 X_f d_a^{-1/2} + 500 X_p \\
TS_{sw} = 29 Mn + 79 Si + 700 P + 5369 N
\] (2.20)

Vinokur has developed equations for yield strength, ultimate tensile strength, percent elongation, reduction in area and impact toughness that incorporate more alloying and tramp elements than in the above equations. The mechanical properties equations developed for conventional rolling will be applicable to CSP mills although the contribution of tramp elements is expected to be higher due to EAF steelmaking. It was found that the presence of small amounts of Ti (0.01 to 0.03%) in C-Mn steels, which has little or no effect on the strength after conventional rolling, is very effective for strengthening of steels produced by direct rolling of thin slabs due to enhanced precipitation hardening. Even in the absence of Ti, precipitation hardening from fine and homogeneous distribution of MnS precipitates has been reported.

2.5 Effect of Alloying and Tramp Elements

Alloying elements are intentionally added to steel to enhance processing/property while residual or trace/tramp elements are those elements (mainly Cu, Ni, Cr, Mo, Sn, Sb, etc) in steel that are not intentionally added, originating from raw materials employed in steelmaking (ore, flux, coke, scrap, etc) and are often difficult to remove during steelmaking and secondary
refining. In general, the addition of alloying elements or the presence of tramp elements in steel can influence the microstructural evolution and mechanical properties of steel in the following ways:

(i) the element can dissolve interstitially or substitutionally to form single phase solid solutions leading to solid solution hardening and affecting the stacking fault energy.

(ii) the element may combine with carbon or nitrogen to form precipitates resulting in precipitation hardening and grain refinement of austenite.

(iii) the element can segregate to the grain boundary of austenite, leading to grain boundary strengthening or softening as well as grain boundary pinning (P, S, Cu, Sn, As, Sb).

(iv) the element may stabilize a particular phase in steel thereby affecting transformation behavior.

The influence of the common alloying elements in steel (C, Mn, Si, Nb, Ti, V) are known to some extent. However, the exact nature of the effects of tramp elements on the processing, properties and service behavior of steel, is not clear. The growth of scrap-intensive EAF steelmaking, a backbone of thin slab steel production, has elevated the importance of tramp elements in steel production. This is because each recycling stage is generally associated with a melting loss of 3-4 pct. Fe and since the residual elements do not oxidize in the presence of Fe, their level gradually increase. It has been estimated that the residual content doubles when steel is recycled 25 times. Table 2.4 lists some of the possible metallurgical effects of tramp elements on processing and properties of steel.

Prior to rolling, alloying and tramp elements contribute to precipitation, segregation, hot shortness, hot embrittlement and cracking during continuous casting. At the reheating stage, these elements can contribute to hot shortness, surface embrittlement and adherent scale formation. Modification of flow stress behavior, precipitation, recrystallization, grain growth
and solid-state segregation during rolling has been traced to these elements\textsuperscript{105,106}. Rolling defects such as rolled-in scale and surface imperfections (slivers, laps, cracks, stickers, etc) are also attributable to these elements\textsuperscript{106}.

The effects of various alloying and tramp elements on the hot strength of steel were depicted in Figure 2.6. With respect to the recrystallization behavior, the effects of various elements is summarized in Figure 2.9\textsuperscript{105,107,108}. It is noted that even the effect of carbon on hot strength and recrystallization is not completely understood\textsuperscript{10,110}. The activation energy for static recrystallization was found to be dependent on composition as follows\textsuperscript{111}:

\[
Q_{\text{rex}} (kJ / mol) = 124.714 + 28.386[\% Mn] + 64.717[\% Si] + 72.775[\% Mo] + 76.830[\% Ti]^{0.213} + 121.1[\% Nb]^{0.1}
\] (2.21)

The constant \(A_1\) in the \(t_{0.5}\) equation was also found to be dependent on composition, generally increasing as \(Q_{\text{rex}}\) decreases\textsuperscript{112}.

In the runout table, precipitation and transformation behavior can be modified by the alloying and tramp elements. C, Ni, Mn, Co, Cu and N are \(\gamma\) stabilizes which retard \(\gamma\) decomposition by lowering the \(\gamma\rightarrow\alpha\) transformation start temperature \((A_{r3})\). Si, Al, P, Cr, V, Mo, Ti and Nb are \(\alpha\) stabilizers which raise the \(A_{r3}\) but may also retard decomposition by slowing down the diffusivity of carbon in \(\gamma\) and some other reason such as grain boundary enrichment\textsuperscript{70}. The effect of composition on the equilibrium \(\gamma\leftrightarrow\alpha\) transformation temperatures \((A_{r3}\) and \(A_{e3}\)) has been quantified as follows\textsuperscript{113}:

\[
A_{r3} (^{\circ}C) = 910 - 203\sqrt{C} - 15.2Ni + 44.7Si + 104V + 31.5Mo + 13.1W - (30Mn + 11Cr + 20Cu - 700P - 400Al - 120As - 400Ti)
\] (2.22)

\[
A_{e3} (^{\circ}C) = 723 - 10.7Mn - 16.9Ni + 29.1Si + 16.9Cr + 290As + 6.38W
\] (2.23)

Even the ferrite grain size that results from transformation as well as the softening after high temperature coiling has been found to be dependent on composition\textsuperscript{102}.
At room temperature, the mechanical properties, formability and texture are affected by these elements. It has been shown that the yield strength, ultimate tensile strength, percent elongation, reduction in area and impact toughness can all be expressed as direct functions of the various alloying and tramp elements in steel.\textsuperscript{103}
Table 2.1 Typical heat transfer correlations for spray cooling.\(^{42}\)

<table>
<thead>
<tr>
<th>Heat Transfer Regime</th>
<th>Correlation</th>
</tr>
</thead>
</table>
| Boiling (quenching) regime                          | \[
| \[
| Transition boiling regime                           | \[
| \[
| Critical heat flux                                  | \[
| Nucleate boiling regime                             | \[
| Onset of single-phase regime                        | \[
| Single-phase regime                                 | \[

Units of the parameters are: \(q '[W \, m^{-2}], \, T[K], \, Q'[m^3 \, s^{-1} \, m^{-1}], \, U_m[m \, s^{-1}], \, d_11[m], \, k_f[W \, m^{-1} \, K^{-1}], \, \rho_i[kg \, m^{-3}], \, \rho_i[kg \, m^{-3}], \, c_p[J \, kg^{-1} \, K^{-1}], \, h_i[J \, kg^{-1}], \, \sigma[N \, m^{-1}]\).

Table 2.2 Friction conditions for solid-solid contact.\(^{63}\)

<table>
<thead>
<tr>
<th>Friction Condition</th>
<th>Equation</th>
<th>Remark</th>
</tr>
</thead>
</table>
| Constant frictional Stress  | \[
| No-slip condition           | \[
| Coulomb friction            | \[

Covers the whole range of slipping and sticking. Accuracy depends on the shear factor (m) and velocity.

Simple to use but cannot handle slipping condition. Accuracy depends on the friction coefficient (\(\mu\)) and pressure.

Covers the whole range of slipping and sticking. Accuracy depends on the friction coefficient (\(\mu\)) and pressure.
Table 2.3 Some constitutive relations for steels at high temperatures.

<table>
<thead>
<tr>
<th>σ-Function</th>
<th>Example</th>
<th>Ref.</th>
<th>Remarks</th>
</tr>
</thead>
<tbody>
<tr>
<td>( \sigma = f(T, \varepsilon) )</td>
<td>( \sigma = \frac{1}{n} \sinh^{-1} \left[ \frac{\varepsilon \exp(Q_{def} / RT)}{A} \right]^{1/n} )</td>
<td>70</td>
<td>Stress can only be evaluated at a specific strain. The effects of strain and grain size can be included in the fitting parameters.</td>
</tr>
<tr>
<td></td>
<td>( \sigma = \frac{1}{n} \left[ \varepsilon \exp(Q_{def} / RT) / A_1 \right]^{1/n} )</td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>( \sigma = \frac{1}{n} \left[ \ln(\varepsilon / A_1) + \frac{Q_{def}}{RT} \right] )</td>
<td></td>
<td></td>
</tr>
<tr>
<td>( \sigma = f(T, \varepsilon, \varepsilon) )</td>
<td>( \sigma = A_3 \varepsilon^p \varepsilon^q \exp(Q_{def} / RT) )</td>
<td>70,71</td>
<td>The effect of grain size can be included in the fitting parameters.</td>
</tr>
<tr>
<td></td>
<td>( \sigma = \sigma_0 + (\sigma_{ss} - \sigma_0)(1 - \exp(-k\varepsilon))'' )</td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>( \sigma_0, \sigma_{ss}, k = f(T, \varepsilon) )</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

\( \Delta \sigma = f(\sigma, \varepsilon) \)

\( \sigma_{drx} = \sigma - \Delta \sigma \)

\( \Delta \sigma = (\sigma_{ss1} - \sigma_{ss2}) \left\{ 1 - \exp(-C \left( \frac{\varepsilon - \varepsilon_c}{\varepsilon_{ss2} - \varepsilon_c} \right)^k) \right\} \)

Table 2.4. Effect of tramp elements on processing and properties of steel\(^{106}\)

1. The processing conditions in terms of:
   - Recrystallization and rolling forces in the hot strip mill: Mo, Cr, Sn, etc.
   - Austenite-to-ferrite transformations, hardenability: all
   - Hot ductility during hot deformation: Zn, Sn, etc.
   - Recrystallization during annealing: Mo, Cr, Sn, etc.
2. The surface aspect of the hot rolled and pickled strip: Cu, Ni, As, Sn, etc.
   - Due to hot shortness
   - Due to possible synergy of Cu and Sn in hot shortness
3. The embrittlement of grain boundaries: Sb, Sn and As
   - During strip coiling
   - During batch or continuous annealing of low carbon steels
4. The precipitate/matrix interface segregation phenomena: Sn
   - Ostwald ripening, precipitate growth, texture control
   - Sn on Fe\(_N\), Sn on MnS, Sb on TiC, etc.
5. The mechanical properties of the end products: all
   - Hot strips and cold rolled sheets
   - Plates
   - Long products
6. The coating by hot dip or electrodeposition
7. The weldability of high strength steel grades: Mo, Cr, Cu and Ni
Figure 2.1  Water cooling systems in the final stand of a finishing mill and the runout out table\textsuperscript{37}. 

43
Figure 2.2  Comparison of the heat transfer coefficient and the cooling rate for three water cooling systems\textsuperscript{39}.
Figure 2.3 Schematic of the boiling curve for a saturated liquid.41
Figure 2.4  Comparison of strain distribution predicted by three finite-element models at the roll bite exit for a 20 mm transfer bar reduced 45 percent.  

---

63
Figure 2.5 Schematic representation of stress-strain curves during the deformation of austenite.\textsuperscript{70}
Figure 2.6 Effect of alloying and tramp elements on the hot strength of austenite\textsuperscript{70,105}. (a) 0.25Si-1.10Mn base steel, (b) 0.10C base steel (c) 0.10C-1.10Mn base steel, (d) 0.02C-0.2Mn-0.05Al base steel.
Figure 2.7  Schematic representation of stress-strain curve in a double-hit test.\textsuperscript{70}
Figure 2.8  Effect of precipitation on recrystallization\textsuperscript{89}: (a) recrystallization-precipitation-temperature-time (RPTT) diagram, (b) effect of solute on recrystallization-stop temperature for a 0.07C-1.40Mn-0.25Si base steel.
Figure 2.9  Effects of alloying and tramp elements on recrystallization$^{105,107}$: (a) dynamic recrystallization at 1000 °C, (b) and (c) static recrystallization of 0.02C-0.2Mn-0.05Al base steel, (d) static recrystallization of 0.05C-0.2Mn-0.2Si-0.04Nb base steel.
Chapter 3

OBJECTIVES AND METHODOLOGY

The literature review presented in the last section clearly shows that although extensive work has been done in understanding the various aspects of hot strip rolling, the extension of the available knowledge into the realms of thin-cast slabs is yet to be studied. Furthermore, there are still gaps in knowledge and understanding of the complex phenomena of heat transfer, deformation and structure modifying processes, and their intricate interactions during hot strip rolling. It is clear that hot strip rolling via the thin-cast slab route exhibits some unique characteristics that distinguish it from a conventional hot strip mill operation. The study of the influence of these unique features on the heat transfer, deformation, structure modifying processes, the final microstructure and mechanical properties of the rolled material will be invaluable in the optimization and control of this new technology.

3.1 Objectives

The present research was directed at accurate prediction of the temperature, deformation behavior (roll forces, stresses, strains and strain rates) and microstructural evolution (recovery, recrystallization, grain growth, austenite and ferrite grain sizes) during CSP rolling, as well as the final mechanical properties of hot rolled strips. Utilizing the A36 (AISI 1018) and DQSK (AISI 1005) steel grades, and a range of steel grades (0.06-0.09 C, 0.16-0.9 Mn) produced at HYLSA's CSP mill at Monterrey (Mexico), the present work intended to accomplish the following specific objectives:

(i) To study the details of deformation, heat transfer and microstructural evolution from the tunnel furnace exit to the downcoiler in the CSP mill with the aid of an integrated mathematical model.
(ii) To simulate the rolling process using available laboratory equipment in order to quantify the flow stress and structural changes during CSP rolling with emphasis on the wide range of grain sizes covered in the CSP mill.

(iii) To link the CSP process parameters to the final microstructure and mechanical properties.

(iv) To validate the model and laboratory results with mill measurements from an operating CSP mill.

(v) To combine (i) to (iv) in order to elucidate, and to highlight, the similarities and differences between CCR and CSP rolling in terms of the thermal history, deformation behavior and microstructure evolution.

It was envisaged that the results of this work will provide some direction on the right combination of CSP process variables required to achieve the desired targets as shown in Table 3.1. This will go a long way in not only optimizing CSP technology but also its application in the higher quality end of the flat steel market. It is noted that the present effort is an extension of the ongoing UBC/AISI/DOE microstructural engineering project which is aimed at the development of an efficient integrated model for adequate control of hot rolling of steel strip in order to achieve optimum product quality. The UBC/AISI/DOE microstructural engineering project covers conventional rolling of a range of steel grades including A36 (close to AISI 1018), DQSK (close to AISI 1005), microalloyed (Nb,V) and interstitial free (IF) steels.

### 3.2 Methodology

Figure 3.1 depicts schematically the methodology that was adopted to achieve the above objectives. It is seen from this figure that integrated process modeling (i.e. mathematical modeling and experiments) was at the heart of the present effort. Comprehensive mathematical modeling of
the rolling process was carried out employing finite difference and finite element analysis. This was achieved by the modification and adaptation of three existing UBC models:

(i) an integrated finite-difference model that assumes uniform through-thickness strain to predict the thermal and microstructural evolution in the finishing mill; Sim's equation was added for the prediction of roll forces.

(ii) an Eulerian finite-element deformation model based on the flow formulation method to predict the through-thickness strain and strain rates as well as the mill load.

(iii) a newly developed finite-difference runout table model which utilizes a semi-theoretical analysis of jet impingement and parallel flow boiling to compute strip temperature and the consequent $\gamma \rightarrow \alpha$ phase transformation. The model also computes the yield and ultimate tensile strengths based on some structure/property equations.

In the laboratory, compression tests (both single and double-hits) were carried out on the Gleeble thermomechanical simulator in order to quantify the flow stress and recrystallization behavior of the steels studied under typical CSP mill conditions. Mechanical property tests were performed on an Instron machine according to the standard ASTM tension test procedure. Metallographic examinations were conducted on test specimens and grain sizes measured with the image analyzer.

For model validation purposes, an industrial campaign was carried out at a CSP mill (HYLSA, Monterrey, Mexico) where interstand and inter-bank temperature measurements were conducted. CSP coil samples and complete dynamic data (roll force, FM exit temperature, FM exit thickness, FM exit width and coiling temperature) were also acquired. The coil samples were subjected to metallographic examination, ferrite grain size measurement and mechanical tests. The mechanical tests were utilized to measure the yield strength (YS), ultimate tensile strength (UTS) and percent elongation.
Table 3.1 Desired targets of various CSP process stages and their influencing process variables.

<table>
<thead>
<tr>
<th>CSP Process Stage</th>
<th>Influencing Process Variables</th>
<th>Desired Target</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tunnel Furnace</td>
<td>Temperature, time, furnace atmosphere</td>
<td>Precipitate dissolution, minimal/flaky scale, desired grain size</td>
</tr>
<tr>
<td>Descaling</td>
<td>Water pressure, water flow rate, water coverage, strip speed, strip speed.</td>
<td>Efficient scale removal, minimal slab cooling</td>
</tr>
<tr>
<td>Rolling</td>
<td>Number of stands, reduction per stand, speed, temperature, interstand cooling, roll type, roll diameter, roll cooling.</td>
<td>Optimal number of stands, optimal shape/profile, minimal roll wear, minimal rolling defects, desired finishing temperature, desired finishing grain size, rolling load within mill capacity.</td>
</tr>
<tr>
<td>Runout Table Cooling</td>
<td>Number of banks, number of active jets, location of active jets, strip and water temperature, speed, water flow rate, water coverage, water quality</td>
<td>Desired coiling temperature, desired microstructure, desired mechanical properties, minimal thermal distortion.</td>
</tr>
<tr>
<td>Coiling and Cooling to room temperature</td>
<td>Coiling temperature, cooling rate, time</td>
<td>Desired microstructure, desired mechanical properties.</td>
</tr>
</tbody>
</table>
LABORATORY EXPERIMENTS

CCR & CSP Samples

Instron
- Yield Strength
- Ultimate Tensile Strength
- % Elongation

Gleeble 1500
- Grain Growth
- Flow Stress
- Recrystallization Kinetics
- Transformation Kinetics

Metallography
- Image Analyzer
- Grain Size
- Transformed Phase (Ferrite, Pearlite, etc)

MATHEMATICAL MODELS

Thermal
- 1-D Finite Difference

Deformation
- Eulerian FEM

Microstructure
- Recrystallization
- Grain Growth
- Transformation

Temperature

Deformation Resistance
- Roll Forces
- Strain, Strain Rate

Industrial Trial
- Measured Temperature
- Measured Roll Force
- Ferrite Grain Size
- Yield Strength
- Ultimate Tensile Strength
- % Elongation

Structure/Property Model
- Yield Strength
- Ultimate Tensile Strength
- % Elongation

Model Validation

Model Predictions
- Temperature, Roll Forces, Strain, Strain Rate, \gamma Grain Size, \alpha Grain Size, YS, UTS, % Elongation

Figure 3.1  Flow chart of the research methodology
Chapter 4

EXPERIMENTAL MEASUREMENTS

The first part of the laboratory experiments was designed to elucidate the effect of coarse austenite grain size on the flow stress and recrystallization behavior of plain carbon steels. Transfer-bar samples of two plain carbon steels (A36 and DQSK) were utilized in the experiments; coarse grain sizes greater than 500 µm were examined. The nominal composition of the two steels is listed in Table 4.1. As seen from the table, the BOF-A36 steel with carbon equivalent of 0.3 is a structural quality (SQ) steel with enough Mn to ensure a minimum room-temperature tensile strength of 36 ksi (250 MPa) and is used mainly in consumer goods and structural applications. The DQSK steel with carbon equivalent of 0.1 is a drawing quality (DQ) steel used for deep-drawn parts and other parts requiring severe deformation such as stamping, automobile bodies and some household appliances. Specifically, grain growth experiments, and single-hit and double-hit tests were carried out in the Gleeble 1500 thermomechanical simulator and the resulting grains from the tests were measured with the image analyzer. It is acknowledged that reheated transfer-bar samples differ from virgin as-cast austenite but the influence of coarse grain size on deformation and recrystallization behavior can be effectively investigated with the former.

In the second part, three CSP slab samples, cut from the head end by the flying shear located just before the descaling unit and quenched immediately in water, were used to measure the flow stress, recrystallization and transformation behavior of 0.06-0.075C CSP steels. The specific compositions of the three slabs are also listed in Table 4.1. The measurements were compared with DQSK values since the carbon content of the three CSP steels are closer to DQSK than A36.
4.1. Grain Growth - Development of Coarse Austenite

The grain growth of isothermally reheated A36 and DQSK steel grades up to grain sizes of 250 µm has been investigated\textsuperscript{86}. It was found that in order to obtain a homogeneous microstructure, the heating rate at various temperature ranges has to be optimized. In both steel grades, the heating rates in three temperature regions have been identified as crucial:

(i) the $\alpha \rightarrow \gamma$ transformation regime

(ii) region of abnormal grain growth attributed to coarsening and dissolution of AlN

(iii) normal grain growth region

Fast heating rate in (i) will lead to fine initial austenite grains while slow heating rate will result in coarse initial austenite. Fast heating rate in (ii) limits the extent of inhomogeneity while slow heating rate will enhance inhomogeneity. Fast heating rate in (iii) will result in fine grain size while slow heating rate yields coarse grains.

Based on these results\textsuperscript{86}, a multi-stage heat treatment which is depicted in Figure 4.1 was found necessary to ensure coarse homogeneous austenite structure. Solid cylindrical specimens, 10 mm in diameter and 15 mm in height, were heat treated in the Gleeble 1500 thermomechanical simulator. It was observed that soaking at 1300 °C for longer than 300 seconds or soaking at higher temperatures often resulted in local melting while soaking for longer times at temperatures lower than 1300 °C resulted in extensive surface oxidation and grain size inhomogeneity. After soaking to different times, the samples were either water or helium quenched on the Gleeble 1500 thermomechanical simulator. The fast cooling rate ensured that only a small amount of austenite-ferrite transformation occurs at austenite grain boundaries. The samples were then ground, polished and etched in 2% Nital. The microstructure observed revealed ferrite outlining the initial austenite grain boundaries.
Figure 4.2 depicts the microstructures obtained after soaking A36 steel specimens for 150 and 300 seconds respectively at 1300 °C, following the heat treatment of Figure 4.1. The austenite grains thus revealed were measured with the image analyzer employing the standard ASTM planimetric or Jeffries’ procedure (ASTM Designation E 112) and the results are listed in Table 4.2. The mean equivalent-area diameter (EQAD) measured for the micrographs depicted in Figure 4.2 are 727 and 1110 μm for the soaking time of 150 and 300 seconds respectively. Due to the very fast transformation kinetics of DQSK, the parabolic equation (Equation 2.13) was utilized to estimate the coarse grain size after heat treatment rather than from measurements since the prediction from the equation compares reasonably well with measured values for the A36 steel. Mean EQAD grain sizes of 918 and 1152 μm were estimated for the soaking time of 150 and 300 seconds respectively. These grain sizes were utilized for single-hit tests for the measurement of flow stress; the peak strain at which dynamic recrystallization occurs was also determined. Quenched samples from single-hit tests were examined metallographically and the recrystallized grain size after a given holding time was measured. Double-hit tests were also carried out on these coarse austenite grains to assess the softening kinetics at a specified deformation condition of temperature, strain and strain rate.

4.2. Measurement of Flow Stress of Coarse Austenite

As discussed in Chapter 2, the flow stress of any material is generally dependent on such parameters as composition, temperature, strain, strain rate and microstructure. The coarse austenite grain sizes developed in section 4.1 above were utilized in single-hit compression tests in the Gleeble 1500 thermomechanical simulator. Solid cylindrical specimens, 10 mm in diameter and 15 mm in height, were mounted on the deformation anvils as shown schematically in Figure 4.3. With respect to the measurement of deformation load and sample displacement, the following data were acquired during deformation.
(1) the instantaneous diameter of the specimen measured by a diametral LVDT attached to the central plane of the test sample.

(2) displacement of the ram after initial contact measured by LVDT attached to the ram.

(3) the load on the ram (referred to as the standard load) measured by a load cell at the end of the ram fixture.

(4) the load on the fixed anvil (referred to as the auxiliary load) measured by a load cell at the end of the fixed anvil.

The instantaneous temperature of the sample was measured by a thermocouple (chromel-alumel type K) attached to the central plane of the sample. Data was acquired in the frequency range of 1000 to 8000 Hertz such that 1000 and 800 data points were acquired for 1 and 10 s\(^{-1}\) strain rates respectively. The tests were performed under vacuum with a small amount of argon left in the chamber (argon flushing is used before vacuum pumping) to dilute any air leaks through the chamber seals.

The diametral measurements were converted to C-strain \(\varepsilon = 2\ln(d/d_0)\) after correcting for the thermal expansion just before the commencement of deformation. This method was considered more accurate than the L-Strain \(\varepsilon = \ln(l_0/l)\) obtained from the ram displacement measurement since the latter presumes perfect transmission of mechanical energy from the ram to the sample. However, the accuracy of this method is still limited by barreling which is maximum at the sample center. The stress was computed by converting the measured diameters into area and dividing the auxiliary load converted into force by this area. The auxiliary load measurement was considered a more accurate measurement because the motion of the ram could lead to inaccuracies in standard load measurement due to the friction associated with moving parts. The operating strain rate is computed by dividing the strain with time. From the measurements, it was found that it takes about 2 to 5 percent of the total deformation time to attain a constant
strain rate, a standard deviation of 3 to 8 percent remains in the strain rate measurement beyond this initial transition. At low strain rates ($< 10 \text{ s}^{-1}$), an average of 2 percent error is associated with the measured flow stress while this error could rise to about 10 percent at higher strain rates ($\geq 10 \text{ s}^{-1}$). The increasing error in measurement at higher strain rate results from the difficulty in controlling the ram motion within extremely short times (the flow stress is measured in a time frame of only 0.1 second at 10 $\text{s}^{-1}$ in the strain range of 0.0 to 1.0). With respect to reproducibility, 2 to 15 MPa difference in flow stress values has been obtained for tests performed under similar conditions.$^{115,116}$

The experimental conditions were designed to cover the range of operation of the first and second stands of the CSP mill where coarse grain size plays a role. It is noted that the flow stress of austenite in the grain size range of 14-250 $\mu$m has been measured and analyzed for both steels.$^{117}$ Hence, in combination with the present efforts, the flow stress covering the whole range of operating conditions in the CSP mill has been captured. The measured flow curves were utilized as a basis for the quantitative determination of the deformation resistance at any given set of mill operating conditions. The behavior of the stress-strain curve in terms of hardening and softening phenomena was also delineated from the measured flow curves. Furthermore, the peak strain which is a very important parameter in the flow stress and microstructural calculations, was extracted from those curves that exhibited dynamic recrystallization. The influence of various process variables on the measured flow stress are shown in Figures 4.4 through 4.6.

4.2.1 Effect of Temperature

Figure 4.4 (a) depicts the stress strain curves for the 1110 $\mu$m grain size of A36 steel at a strain rate of 1 $\text{s}^{-1}$ and temperatures of 1050, 1100 and 1150 $^\circ\text{C}$ respectively. Figure 4.4 (b) shows the stress strain curves for the 1152 $\mu$m of DQSK steel at the same strain rate and
temperatures. As expected, the flow stress at a given strain increases with decreasing temperature for both steels. Furthermore, decreasing temperature retards dynamic recrystallization manifested by the increasing strain at which a peak in flow stress occurs. A 100 °C decrease in temperature is seen to result in a 10 to 40 MPa increase in the flow stress at a given strain. The effect of temperature on flow stress can be explained by the fact that the two restoration mechanisms, dynamic recovery and dynamic recrystallization, are thermally activated. The rate of formation of cells or subgrains by dislocation rearrangement (recovery) as well as the rate of nucleation and growth of strain-free grains (recrystallization) increase with increasing temperature as a result of faster grain boundary mobility and diffusivity. Hence, softening dominates at high temperatures while working hardening dominates at lower temperatures. It is noted that these temperatures (1050 to 1150 °C) fall within the operating temperatures at the first and second stands of the CSP mill where coarse grain size plays a role, as discussed in later sections.

4.2.2 Effect of Strain Rate

Figure 4.5 (a) shows the stress strain curves for the 1110 μm grain size of A36 steel at a temperature of 1050 °C and strain rates of 1, 5 and 10 s⁻¹ respectively. Figure 4.5 (b) shows the stress strain curves for the 918 μm of DQSK steel at the same temperature and strain rates. It is evident that the flow stress at a given temperature, strain and austenite grain size increases with increasing strain rate. In contrast to the effect of temperature, higher strain rates shift the curves away from dynamic recrystallization. It is seen that for the two coarse grains shown in Figure 4.5, dynamic recrystallization occurs only at strain rates of 1 s⁻¹. An increase in strain rate from 1 to 10 s⁻¹ raises the flow stress by up to 45 MPa. The effect of strain rate on flow stress results from the increased rate of work hardening (dislocation pile up) at high strain rates coupled with a reduction in the time available for dynamic recrystallization. These strain rates (1 to 10 s⁻¹) reflect
the operating strain rates at the first and second stands of the CSP mill where coarse grain size plays a role, as discussed in later sections.

4.2.3 Effect of Austenite Grain Size

Figure 4.6 (a) portrays the stress strain curves for three grain sizes of A36 steel (244, 727 and 1110 μm) at a temperature of 1100 °C and strain rate of 10 s⁻¹. Figure 4.6 (b) displays the stress strain curves for three grain sizes of DQSK steel (249, 918 and 1152 μm) at the same temperature and strain rate. It is evident from these curves that the flow stress at a given strain decreases with increasing grain size provided that the softening mechanism remains the same. An increase in grain size from 244 to 1110 μm which is typical of the first stands of a conventional finishing mill and CSP hot-strip mill respectively, can result in a 30 MPa decrease in flow stress as seen from Figure 4.6 (a). It has been suggested that a finer grain size increases the work hardening rate in the initial stages of deformation by accelerating the process of subgrain formation, the rate controlling step in the initial stages of deformation. Beyond the initial stage of deformation, a fine grain size promotes dynamic recrystallization, thereby lowering the flow stress after the peak as evidenced in curve 1 of Figure 4.6 (a). On the other hand, a coarse grain size reduces the tendency toward dynamic recrystallization, ensuring that work hardening predominates with a consequent increase in flow stress as illustrated by curves 2 and 3 of Figure 4.6 (a).

4.2.4 Peak Occurrence and Peak Strain

It is evident from the foregoing that temperature, strain rate and austenite grain size are the major factors that control the occurrence of dynamic recrystallization. A large number of flow stress curves (close to 100 for both steels) for five grain sizes of each steel representing the grain
size range for the CCR and CSP processes were carefully inspected and classified in three categories:

(i) curves with a peak which indicates the onset of dynamic recrystallization
(ii) curves without peaks where the flow stress clearly continues to increase with increasing strain.
(iii) curves that are at the borderline between (i) and (ii).

The results are displayed in Figures 4.7 and 4.8 which are a series of strain rate versus temperature graphs. The two figures indicate that a boundary exists between flow curves with a peak and those without a peak. This is a very important finding, because it is the first attempt known to the author to quantitatively delineate the occurrence of a peak in the flow curve for any given set of deformation conditions. The practical implications of the aforementioned boundary are two fold. Firstly, with respect to the flow stress, it shows the presence or absence of a peak in the flow curve for any given temperature and grain size; a peak can only occur when the applied strain rate is lower than the boundary strain rate. This enables the relevant portion of the stress-strain curve (work hardening, steady-state and dynamic recrystallization) to be modeled effectively. Secondly, with respect to the recrystallization kinetics, it predicts the presence or absence of dynamic or metadynamic recrystallization in conjunction with the critical strain. Dynamic recrystallization will occur during deformation when the applied strain rate is lower than the boundary strain rate and the corresponding applied strain is greater than the critical strain. Between stands, metadynamic recrystallization will occur when the applied strain rate is lower than the boundary strain rate and the corresponding applied strain is greater than the critical strain; otherwise static recrystallization takes place at interstand locations.

The measured peak strains for coarse austenite are listed Table 4.3. As expected, the peak strain increases with decreasing temperature, increasing strain rate and increasing grain size.
4.3 Kinetics of Recrystallization and Recrystallized Grain Size

Double-hit compression tests were performed to evaluate the effect of coarse grain size on the softening rate following deformation. Figures 4.9 displays the stress-strain curves for the first and second hits obtained for 727 and 1110 μm of A36 steel at 1100 °C and holding times of 1, 2 and 4 seconds respectively; the corresponding 0.2 % offset yield stress values for each hit are shown in the figure ($\sigma_{y1}$ and $\sigma_{y2}$). Similar results for 918 and 1152 μm of DQSK steel at the same temperature and holding times are shown in Figure 4.10. A target strain rate of 5 s$^{-1}$ was employed in the tests; the actual strain rates are listed in Table 4.4. It is observed that the stress-strain curves for the second hit are quite different from the first one, the difference indicating the structural state prior to each deformation stage. The ratio of $\sigma_{y1}$ to $\sigma_{y2}$ increases with increasing delay time between the two hits, reflecting the increasing degree of restoration. Fractional softening was computed according to the procedure depicted earlier in Figure 2.7. Only static recrystallization was considered since it has been shown that the metadynamic recrystallization kinetics for both steels is independent of grain size$^{119}$. Static recrystallization was assumed to start after 20 pct. softening (end of the incubation period) as suggested in the literature and applied to both steels in earlier studies$^{118,119}$. Table 4.4 lists the computed fraction recrystallized at various holding times for the stress-strain curves of Figures 4.9 and 4.10. For the grain sizes measured and strain range, it is evident that complete recrystallization ($F_x \geq 0.95$) occurs between 2 and 4 seconds at 1100 °C and strain rate of about 5 s$^{-1}$.

Single-hit tests were performed for similar deformation conditions to confirm that complete recrystallization occurs close to 4 seconds and to determine the recrystallized grain size. The samples from these tests were quenched and subjected to metallographic examination and grain size measurement as outlined in section 4.1. Figure 4.11 depicts the recrystallized grain size after 4 seconds for the 727 and 1110 μm of A36 steel. It is observed that complete recrystallization
did occur as evidenced in the fully equiaxed grain sizes. The average recrystallized grain size was measured (EQAD) to be 67 and 142 μm for initial austenite grain sizes of 727 and 1110 μm respectively.

4.4 Flow Stress, Recrystallization and Transformation of CSP Steels.

Grain growth, single-hit, double-hit and continuous cooling transformation (CCT) tests were performed on the three CSP steel grades listed in Table 4.1 to evaluate the difference between their deformation, recrystallization and transformation behaviors and that of DQSK. The grain growth tests were performed utilizing exactly the same heat-treatment illustrated in Figure 4.1, a holding time of 150 seconds at 1300 °C was used. Similar to DQSK, the resulting austenite grain size could not be measured due to the very fast transformation kinetics of these CSP steels. Hence, the 918 μm grain size estimated for DQSK under the same heat treatment was assumed to apply to the three CSP steels.

Single-hit tests were carried out to compare the flow stress of the three CSP steels with DQSK. The results of these tests are presented in Figure 4.12 for two strain rates and temperatures. It is observed that the 0.062C-0.155Mn (7061) and 0.075C-0.325Mn (7092) grades exhibited similar hot strengths as DQSK at 1100 °C and 1 s⁻¹ but higher strengths at 1050 °C and 10 s⁻¹. It is further observed that both the 7061 and 7092 steels exhibited lower peak strain than DQSK, indicating faster dynamic recrystallization rate. The 0.071C-0.899Mn (7094) grade showed higher strength than DQSK and the other two grades at both test conditions. This trend is a reflection of the carbon equivalent of the steels (0.1 for DQSK, 0.12 for 7061, 0.15 for 7092 and 0.24 for 7094), the increased hot strength of CSP steels resulting not only from the higher carbon but also the higher residual contents of the steels as reported in literature 70,105.

Double-hit tests were carried out to evaluate the rate of recrystallization. The result is depicted in Figure 4.13 for a holding time of 2 s at 1100 °C (~ 5 s⁻¹ strain rate) and the fractional
softening results are listed in Table 4.5. It is observed that under the test conditions, the 7061 and 7092 grades exhibited peaks and therefore, displayed the relatively faster metadynamic recrystallization at the interstand which was already completed before 2 seconds. On the other hand, the 7094 grade manifested static recrystallization which appears to be on the verge of completion at 2 seconds, a situation that is similar to A36 and DQSK under the same test conditions (Table 4.4).

With respect to transformation, the samples were heated to 1000 °C, held for 5 minutes and cooled in air (~20 °C/s) in the Gleeble 1500 thermomechanical simulator. The measured dilation is plotted against temperature in Figure 4.14. It is observed that the transformation behavior of the 7061 grade closely matches that of DQSK; the transformation start temperature (T_s) is only 5 °C lower than that of DQSK while the transformation finish temperature (T_f) is 12 °C lower. Transformation starts and finishes at relatively lower temperatures in the 7092 grade (35 °C lower in T_s and 42 °C lower in T_f) while the transformation behavior of the 7094 grade is vastly different from DQSK (76 °C lower in T_s and 92 °C lower in T_f). The measured transformation start temperatures reflect the difference in the calculated equilibrium T_Ae3 temperature (874 °C for 7061, 865 °C for 7092 and 854 °C for 7094).

These three tests (single-hit, double-hit and continuous cooling transformation) indicate that two of the three CSP steels, 7061 and 7092, can effectively be modeled as DQSK steel grade with some error but not the 7094 grade. In the following sections, the rolling of four CSP steels (2061, 7061, 2082, 7092) were simulated with DQSK equations. The 2061 is similar to the 7061 grade except for a slightly higher Mn (0.3 instead of 0.18) and Si (0.1 instead of 0.03). The 2082 grade is similar to 7092 except for slightly higher Si (0.1 instead of 0.03) and lower Cu (0.1 instead of 0.2).
Table 4.1. Composition of steels utilized in the tests.

<table>
<thead>
<tr>
<th>Steel Grade</th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Si</th>
<th>Cu</th>
<th>Ni</th>
<th>Cr</th>
<th>Sn</th>
<th>Al</th>
<th>N</th>
<th>Ceq</th>
</tr>
</thead>
<tbody>
<tr>
<td>DQSK</td>
<td>0.038</td>
<td>0.30</td>
<td>0.01</td>
<td>0.008</td>
<td>0.009</td>
<td>0.015</td>
<td>0.025</td>
<td>0.033</td>
<td>≤0.005</td>
<td>0.04</td>
<td>0.0052</td>
<td>0.097</td>
</tr>
<tr>
<td>A36</td>
<td>0.17</td>
<td>0.74</td>
<td>0.009</td>
<td>0.008</td>
<td>0.012</td>
<td>0.016</td>
<td>0.01</td>
<td>0.019</td>
<td>≤0.005</td>
<td>0.04</td>
<td>0.0047</td>
<td>0.299</td>
</tr>
</tbody>
</table>

I. BOF steel composition

II. Composition of three CSP steels (EAF)

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Si</th>
<th>Cu</th>
<th>Ni</th>
<th>Cr</th>
<th>Sn</th>
<th>Al</th>
<th>N</th>
<th>Ceq</th>
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<tr>
<td>7094</td>
<td>0.071</td>
<td>0.899</td>
<td>0.009</td>
<td>0.003</td>
<td>0.178</td>
<td>0.12</td>
<td>0.055</td>
<td>0.019</td>
<td>0.008</td>
<td>0.036</td>
<td>0.005</td>
<td>0.237</td>
</tr>
<tr>
<td>7092</td>
<td>0.075</td>
<td>0.325</td>
<td>0.008</td>
<td>0.003</td>
<td>0.047</td>
<td>0.184</td>
<td>0.056</td>
<td>0.017</td>
<td>0.01</td>
<td>0.045</td>
<td>0.007</td>
<td>0.15</td>
</tr>
<tr>
<td>7061</td>
<td>0.062</td>
<td>0.155</td>
<td>0.005</td>
<td>0.004</td>
<td>0.029</td>
<td>0.073</td>
<td>0.043</td>
<td>0.011</td>
<td>0.007</td>
<td>0.043</td>
<td>0.007</td>
<td>0.118</td>
</tr>
</tbody>
</table>

\[ C_{eq} = \text{carbon equivalent} = \%C + \frac{\%Mn}{6} + \frac{\%Cr + \%Mo + \%V}{5} + \frac{\%Ni + \%Cu}{15} \]

Table 4.2 Measured and estimated austenite grain size.

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Time (s)</th>
<th>Measured (d_{\text{r}}) (μm)</th>
<th>Estimated (d_{\text{r}}) (μm) (Equation 2.13)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>(A) A36  Steel</td>
<td></td>
<td></td>
</tr>
<tr>
<td>1300</td>
<td>150</td>
<td>727</td>
<td>857</td>
</tr>
<tr>
<td>1300</td>
<td>300</td>
<td>1110</td>
<td>1076</td>
</tr>
<tr>
<td></td>
<td>(B) DQSK</td>
<td></td>
<td></td>
</tr>
<tr>
<td>1300</td>
<td>150</td>
<td>-</td>
<td>918</td>
</tr>
<tr>
<td>1300</td>
<td>300</td>
<td>-</td>
<td>1152</td>
</tr>
</tbody>
</table>
Table 4.3 Measured peak strain for coarse austenite with corresponding peak stress.

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Strain Rate (s⁻¹)</th>
<th>Grain Size (μm)</th>
<th>Peak Strain</th>
<th>Peak Stress (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1100</td>
<td>4</td>
<td>727</td>
<td>0.46</td>
<td>92</td>
</tr>
<tr>
<td>1050</td>
<td>1</td>
<td>1110</td>
<td>0.44</td>
<td>89</td>
</tr>
<tr>
<td>1100</td>
<td>1</td>
<td>1110</td>
<td>0.4</td>
<td>67</td>
</tr>
<tr>
<td>1150</td>
<td>1</td>
<td>1110</td>
<td>0.37</td>
<td>58</td>
</tr>
<tr>
<td>1150</td>
<td>5</td>
<td>1110</td>
<td>0.46</td>
<td>78</td>
</tr>
<tr>
<td>(A). A36</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>1050</td>
<td>1</td>
<td>918</td>
<td>0.33</td>
<td>80</td>
</tr>
<tr>
<td>1100</td>
<td>1</td>
<td>918</td>
<td>0.31</td>
<td>64</td>
</tr>
<tr>
<td>1150</td>
<td>1</td>
<td>918</td>
<td>0.24</td>
<td>55</td>
</tr>
<tr>
<td>1050</td>
<td>1</td>
<td>1152</td>
<td>0.41</td>
<td>76</td>
</tr>
<tr>
<td>1100</td>
<td>1</td>
<td>1152</td>
<td>0.31</td>
<td>66</td>
</tr>
<tr>
<td>1100</td>
<td>5</td>
<td>1152</td>
<td>0.39</td>
<td>81</td>
</tr>
<tr>
<td>1150</td>
<td>1</td>
<td>1152</td>
<td>0.29</td>
<td>59</td>
</tr>
<tr>
<td>1150</td>
<td>5</td>
<td>1152</td>
<td>0.39</td>
<td>73</td>
</tr>
<tr>
<td>(B). DQSK</td>
<td></td>
<td></td>
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</table>
Table 4.4 Measured fractional softening for coarse austenite at 1100 °C.

<table>
<thead>
<tr>
<th>Time (s)</th>
<th>Strain</th>
<th>( \dot{\varepsilon} ) (s(^{-1}))</th>
<th>( \sigma_m ) (MPa)</th>
<th>( \sigma_y1 ) (MPa)</th>
<th>( \sigma_y2 ) (MPa)</th>
<th>( F = \frac{\sigma_m - \sigma_r}{\sigma_m - \sigma_y1} )</th>
<th>( F = \frac{F - 0.2}{0.8} )</th>
</tr>
</thead>
<tbody>
<tr>
<td>I (a). A36, ( d_r = 727 , \mu m )</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>1</td>
<td>0.36</td>
<td>4.96</td>
<td>98</td>
<td>43</td>
<td>48</td>
<td>0.91</td>
<td>0.89</td>
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<tr>
<td>2</td>
<td>0.41</td>
<td>5.21</td>
<td>92</td>
<td>43</td>
<td>47</td>
<td>0.92</td>
<td>0.90</td>
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<tr>
<td>4</td>
<td>0.43</td>
<td>5.35</td>
<td>96</td>
<td>44</td>
<td>44</td>
<td>1</td>
<td>1</td>
</tr>
<tr>
<td>I (b). A36, ( d_r = 1110 , \mu m )</td>
<td></td>
<td></td>
<td></td>
<td></td>
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<td>1</td>
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<td>5.17</td>
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<tr>
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<tr>
<td>4</td>
<td>0.42</td>
<td>5.27</td>
<td>101</td>
<td>43</td>
<td>45</td>
<td>0.97</td>
<td>0.96</td>
</tr>
<tr>
<td>II (a). DQSK, ( d_r = 918 , \mu m )</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>1</td>
<td>0.39</td>
<td>4.68</td>
<td>95</td>
<td>44</td>
<td>50</td>
<td>0.88</td>
<td>0.85</td>
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<tr>
<td>2</td>
<td>0.35</td>
<td>4.95</td>
<td>83</td>
<td>45</td>
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<td>0.90</td>
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<td>0.35</td>
<td>4.95</td>
<td>79</td>
<td>41</td>
<td>41</td>
<td>1</td>
<td>1</td>
</tr>
<tr>
<td>II (b). DQSK, ( d_r = 1152 , \mu m )</td>
<td></td>
<td></td>
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<tr>
<td>1</td>
<td></td>
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<td></td>
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<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>0.34</td>
<td>4.67</td>
<td>90</td>
<td>43</td>
<td>49</td>
<td>0.87</td>
<td>0.84</td>
</tr>
<tr>
<td>4</td>
<td>0.34</td>
<td>4.78</td>
<td>89</td>
<td>44</td>
<td>44</td>
<td>1</td>
<td>1</td>
</tr>
</tbody>
</table>

\( \sigma_m \) = maximum stress at the end of the first hit, \( \sigma_y1 \) = 0.2 % offset yield stress for the first hit \( \sigma_y2 = 0.2 \% \) offset yield stress for the second hit

Table 4.5 Measured fractional softening of the CSP steels at 1100 °C (\( d_r = 918 \, \mu m \)).

<table>
<thead>
<tr>
<th>Time (s)</th>
<th>Strain</th>
<th>( \dot{\varepsilon} ) (s(^{-1}))</th>
<th>( \sigma_m ) (MPa)</th>
<th>( \sigma_y1 ) (MPa)</th>
<th>( \sigma_y2 ) (MPa)</th>
<th>( F = \frac{\sigma_m - \sigma_r}{\sigma_m - \sigma_y1} )</th>
<th>( F = \frac{F - 0.2}{0.8} )</th>
</tr>
</thead>
<tbody>
<tr>
<td>(a). 0.06C-0.155Mn (7061)</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>0.48</td>
<td>4.85</td>
<td>88</td>
<td>53</td>
<td>50</td>
<td>1</td>
<td>1</td>
</tr>
<tr>
<td>(b). 0.075C-0.325Mn (7092)</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>0.45</td>
<td>4.73</td>
<td>88</td>
<td>53</td>
<td>50</td>
<td>1</td>
<td>1</td>
</tr>
<tr>
<td>(c). 0.071C-0.899Mn (7094)</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>0.44</td>
<td>4.67</td>
<td>103</td>
<td>57</td>
<td>58</td>
<td>0.98</td>
<td>0.97</td>
</tr>
</tbody>
</table>
Figure 4.1 Multi-stage heat treatment utilized to measure the effect of coarse grain size on flows stress and recrystallization (the test temperature, $T_{\text{test}}$, ranged from 1000 to 1150 °C at strain rates of 1 to 10 s$^{-1}$).
Coarse austenite grains of A36 steel obtained with the multi-stage heat treatment (see Figure 4.1) after soaking for (a) 150 s and (b) 300 s respectively at 1300 °C.
Figure 4.3  Schematic diagram of the gleeble jaws, anvil and specimen assembly.
Figure 4.4  Effect of temperature on the flow stress of coarse austenite at the strain rate of $1 \text{ s}^{-1}$ for (a) A36 and (b) DQSK steels.
Figure 4.5 Effect of strain rate on the flow stress of coarse austenite at 1050 °C for (a) A36 and (b) DQSK steels.
Figure 4.6 Effect of grain size on the flow stress of (a) A36 and (b) DQSK steels.
Figure 4.7  Strain-rate versus temperature plot for five grain sizes of A36 steel showing regions of peak and no-peak curves.
Figure 4.8 Strain-rate versus temperature plot for five grain sizes of DQSK steel showing regions of peak and no-peak curves.
Figure 4.9  Double-hit flow curves for two coarse grain sizes of A36 steel at 1100 °C and holding times 1, 2 and 4 seconds respectively.
Figure 4.10  Double-hit flow curves for two coarse grain sizes of DQSK at 1100 °C and holding times 1, 2 and 4 seconds respectively.
Figure 4.11  Recrystallized grain sizes of A36 steel 4s after deforming to a strain of 0.5 at 1100 °C.
Figure 4.12 Comparison of the flow stress of three CSP steel grades with DQSK.
Figure 4.13  Double-hit results showing the softening rate of three CSP steel grades in comparison with DQSK.
Figure 4.14 Comparison of the transformation behavior of three CSP steel grades with DQSK.
Chapter 5

MATHEMATICAL MODELING

As mentioned in the methodology section of Chapter 3, comprehensive mathematical modeling of the rolling process was carried out employing finite difference and finite element analysis. The availability of earlier models developed at UBC was of immense help in this study, allowing for the elimination of the time and effort required to develop new computer codes. Three existing UBC models were modified and adapted to the present research:

(i) an integrated finite-difference model that assumes uniform through-thickness strain to predict the thermal and microstructural evolution in the finishing mill\textsuperscript{19,120}.

(ii) an Eulerian finite-element deformation model based on the flow formulation method to predict the through-thickness strain and strain rates as well as the mill load\textsuperscript{63,122}.

(iii) a newly developed finite-difference runout table model which utilizes a semi-theoretical analysis of jet impingement and parallel flow boiling to compute strip temperature and the consequent $\gamma \rightarrow \alpha$ phase transformation\textsuperscript{45,47}. The model also computes the yield and ultimate tensile strengths based on the IRSID-SOLAC equations\textsuperscript{102}.

Besides the general adaptation of the original code to the CSP mill configuration, the following specific features were incorporated into the original code:

(a) the addition of Sims's equation\textsuperscript{121} to the original code for the prediction of the mill load.

(b) the incorporation of the new constitutive equations developed in this study into the original code.

(c) the replacement of the earlier literature-based microstructural equations\textsuperscript{120} with new grain growth and recrystallization equations developed specifically for the two steels (A36 and DQSK)\textsuperscript{86,118,128}.
The application of these models to the various sub-units of the CSP mill is schematically illustrated in Figure 5.1. The thermal module of the integrated finite-difference model is seen to be applicable at all locations in the mill before the runout table. The recrystallization and grain growth modules of the integrated finite-difference model are applied after each stand based on the temperature calculations. In the roll gap, the strain and strain rate are computed with either the Eulerian finite-element deformation model or the integrated finite-difference model (with uniform through-thickness strain assumption) based on the predicted strip temperature and the associated deformation resistance, including dynamic softening effects (dynamic recovery and recrystallization). At the runout table, the finite-difference runout table model is used to compute the temperature and the kinetics of austenite decomposition. Finally, the mechanical properties of the strip is calculated based on the predicted coiling temperature and the ferrite grain size.

The important mathematical formulations employed in the three models are discussed below, the detailed numerical discretization and solution procedure has been discussed in detail in the referenced thesis and publications.

5.1 Integrated Finite-Difference Rolling Model

The integrated finite-difference rolling model utilizes uniform through-thickness strain assumption to predict the thermal and microstructural evolution in the finishing mill. The original code was developed by Chris Devadas\textsuperscript{120}. Sims's equation\textsuperscript{121} was added to the modified code for the prediction of roll forces. There are three modules in the model: the heat transfer, the deformation and the microstructural modules.

5.1.1 Heat Transfer During Rolling

The governing heat conduction equation for the strip at steady state is given by:
where \( T \) is the temperature, \( \rho \) is the density, \( k \) is the conductivity, and \( C_p \) is the specific heat per unit volume, \( u \) and \( v \) are velocities along the two coordinate axes, \( x \) and \( y \). \( Q_p \) is the heat generation from plastic deformation as defined in Equation (2.4).

Two valid assumptions were utilized to simplify the above equation:

(i) Strip width is far greater than its thickness such that temperature gradients across the width is negligible compared to through-thickness gradient.

(ii) The heat transfer by bulk motion predominates over the conduction along the strip length due to the high speeds utilized in rolling.

Employing both assumptions and the transformation, \( y = vt \), Equation (5.1) reduces to:

\[
\frac{\partial}{\partial x} \left( k \frac{\partial T}{\partial x} \right) + \frac{\partial}{\partial y} \left( k \frac{\partial T}{\partial y} \right) - \rho C_p u \frac{\partial T}{\partial x} - \rho C_p v \frac{\partial T}{\partial y} + Q_p = 0 \quad (5.2)
\]

A similar equation for the work rolls is given in cylindrical coordinates by:

\[
\frac{1}{r} \frac{\partial}{\partial r} \left( r k_r \frac{\partial T_r}{\partial r} \right) = \rho_r C_{pr} \frac{\partial T}{\partial t} \quad (5.3)
\]

where \( t \) is the time taken for an elemental volume of the rolls to rotate through an angle measured from a reference point.

It is noted that the cyclic temperature variation that occurs during each revolution of the roll is confined to a surface boundary layer (\( \delta \)). In the present study, a value of 0.0135 m was established for \( \delta \) based on the analysis of Tseng et al.\(^{123} \) where the surface boundary layer was defined as the depth where the temperature changes cyclically by greater than 1%.

With respect to heat generation by plastic deformation, the temperature increment at each stage of deformation was computed as follows:\(^{120} \):

\[ \text{\textit{Equation (5.1)}} \]
\[
\Delta T_{def} = \frac{\sigma_f e}{\rho_s C_{ps}}
\] (5.4)

while the frictional component was evaluated from 120:

\[
\Delta T_{fric} = \frac{4\mu \nu_r P_r \alpha \Delta t}{kh \cos \theta}
\] (5.5)

A constant value of 0.35 was assumed for the friction coefficient in this model based on an earlier analysis performed on the finishing mill 120.

Equations (5.2) and (5.3) were discretized and solved numerically with the implicit finite-difference technique, only the top half of the strip thickness was employed in the solution procedure as symmetry was assumed. The thermophysical properties of the steel strip and work rolls employed in the model are listed as functions of temperature in Appendix A.

The following boundary conditions were applied in the computation:

1. Roll-Gap Heat Transfer Coefficient 20,34

\[
-k \frac{\partial T}{\partial x} = h(T_x - T_r)
\] (5.6)

\[
h(W/ m^2.k) = 696.5\overline{p} - 34396
\] (5.7)

or

\[
h(W/ m^2.k) = 28571.43 k_{rs} \left(\frac{\overline{p}}{\sigma_f}\right)^{1.7}
\] (5.8)

where \( k_{rs} = \frac{k_r k_s}{k_r + k_s} \). The flow stress is evaluated at an effective interface temperature given by

\[
T_{interface} = T_s + (T_r - T_s) \frac{\rho_r C_{pr}}{\rho_r C_{pr} + \rho_s C_{ps}(T_s)}.
\]

\( \rho_r \) and \( \rho_s \) are the density of the work roll and strip respectively. \( C_{pr} \) and \( C_{ps} \) are the specific heat of the work roll and strip respectively. Equations (5.7) and (5.8) give similar results except when the effective interface temperature in (5.8) is
lower than $T_{Ac3}$. Therefore, the former (Equation (5.7)) is preferred when there is uncertainty in the flow stress calculation.

(2) Strip Cooling by Water Sprays (Descalers, Back Wash and Interstand Sprays)

Equation (5.6) is applied with the heat transfer coefficient given by:

$$h(W/m^2K) = 10^3 \times \left[ 708.0 \left( \frac{F_w}{A} \right)^{0.75} T_s^{-1.2} + 0.166 \right]$$

(5.9)

In the case that a water film persisted on the top of the strip beyond the impingement zone in the interstand, the recent film boiling correlation of Filipovic et al. was employed. By computing the heat flux for the range of rolling conditions in the finishing mill, it was found that the heat flux does not deviate tangibly from a value of 200 kW/m$^2$. Hence, the heat transfer coefficient is given by

$$h(W/m^2K) = \frac{2.0 \times 10^5}{T_s - T_w}$$

(5.10)

(3) Strip Cooling - Radiation and Convection

Equation (5.6) also applies with the heat transfer coefficient given by:

$$h_{rad} = h_{rad} + h_{conv} = \sigma \epsilon(T) \left( \frac{(T_s + 273.1)^4 - (T_w + 273.1)^4}{(T_s - T_w)} \right) + C \left( \frac{k_{air}}{L} \right) R_e^{n} P_{St}^{1/2}$$

(5.11)

where $\epsilon(T) = \frac{T_s}{1000} \left( 0.125 \frac{T_s}{1000} - 0.38 \right) + 1.1$, $C=0.332$, $n=0.5$ for $R_e \leq 5 \times 10^4$ and $C=0.0288$, $n=0.8$ for $R_e > 5 \times 10^4$. $\sigma$ is the Stefan-Boltzmann constant ($5.68 \times 10^{-8}$ Wm)

(4) Strip Centerline

$$-k \frac{\partial T}{\partial x} = 0, \text{ at } x = 0$$

(5.12)

(5) Roll Cooling - Impingement Zone
The cylindrical coordinate form of Equation (5.6) is applied with the heat transfer coefficient given by:

\[ h(W/m^2.K) = 1.29 \times 10^5 \times \left( \frac{F_w}{A} \right)^{0.521} \]  \hspace{1cm} (5.13)

(6) Roll Cooling - Boiling/Convection

The cylindrical coordinate form of Equation (5.6) is applied with the heat transfer coefficient given by the relevant boiling curve correlations.

6.1 Natural and Forced Convection \(^{120}\) \(T_{sur} < T_{sat}\)

\[ h(W/m^2.K) = 0.11 x \left( \frac{k_w}{D} \right) \times [0.5 Re^2 + Gr Pr]^{0.315} \]  \hspace{1cm} (5.14)

6.2 Nucleate Boiling \(^{120}\) \(T_{sat} < T < T_{max}\)

\[ h = \frac{\dot{q}}{(T_{rs} - 100.0)} \]  \hspace{1cm} (5.15)

where

\[ \dot{q} = \mu_w h_{fg} \sqrt{g(\rho_w - \rho_s)} \left( \frac{C_{pi} \Delta T_x}{h_{fg} \times Pr^n \times C_f} \right)^3 \]  \hspace{1cm} (5.16)

6.3 Unstable Film Boiling \(T > T_{max}\)

In this case, the heat transfer coefficient is obtained by interpolating the experimental data of Nukiyama et al. \(^{124}\) which is employed as an input data in the model.

(7) Roll Cooling - Radiation and Air Convection \(^{120}\)

\[ h(W/m^2.K) = \sigma e A((T_x + 273.1)^4 - (T_\infty + 273.1)^4) / (T_x - T_\infty) \]

\[ + 0.11 \left( \frac{k_{air}}{2R} \right) \left( \left( g_r + 0.5 R_e^2 \right) Pr \right)^{0.3} \]  \hspace{1cm} (5.17)

(8) Roll Surface Boundary Layer (\(\delta\)).

\[ -k_r \frac{\partial T}{\partial r} = 0 \quad , \quad r = R^*, \quad R_\infty - R^* = \delta \]  \hspace{1cm} (5.18)
5.1.2 Deformation

For any given reduction in the roll bite (r), the uniform strain is calculated as follows:

$$\varepsilon = \frac{2}{\sqrt{3}} \ln \left( \frac{1}{1-r} \right)$$  \hspace{1cm} (5.19)

The uniform strain rate ($\dot{\varepsilon}$) is calculated by dividing this strain by the corresponding time.

From entry into the roll bite to exit, the intermediate thicknesses and velocities along the arc of contact are obtained from the conservation of mass

$$h_1v_1 = h_2v_2$$  \hspace{1cm} (5.20)

Sims's equation\textsuperscript{121} was employed to calculate the roll forces as follows:

$$F = \frac{2}{\sqrt{3}} \sigma f Q_p \sqrt{R} \Delta h$$  \hspace{1cm} (5.21)

The deformed work roll radius, $R$, was calculated from Hitchcock formula as follows\textsuperscript{121}:

$$R = R \left( 1.0 + \frac{16F(1 - v^2)}{\pi E \Delta h} \right)$$  \hspace{1cm} (5.22)

$Q_p$ in Equation (5.21) is given by\textsuperscript{121}:

$$Q_p = \frac{\pi}{2f(h)} \tan^{-1} f(h) - \frac{\pi}{4} \left( \frac{\sqrt{R' \Delta h}}{f(h)} \right)$$

$$\times \left[ \ln \left( \frac{2R'(1 - \cos \phi) + h_2}{h_2} \right) + \frac{1}{2} \ln \left( 1.0 - \frac{h_1 - h_2}{h_1} \right) \right]$$  \hspace{1cm} (5.23)

$$\phi = \tan^{-1} \left[ \frac{\pi \ln \left( \frac{h_1 - h_2}{h_1} \right)}{8\sqrt{R'/h_2} + 0.5 \tan^{-1} \left( f(h) / \sqrt{R'/h_2} \right)} \right]$$  \hspace{1cm} (5.24)

where $\Delta h = h_1 - h_2$ and $f(h) = \sqrt{\frac{(h_1 - h_2) / h_1}{1 - (h_1 - h_2) / h_1}}$

The development of constitutive equations for the prediction of the deformation resistance ($\sigma_0$) will be discussed in the following section.
5.1.2.1 Modeling of Flow Stress - Effect of Coarse Grain Size

The first attempt at modeling the deformation resistance (σ₀) for the two steel grades was based on the hyperbolic sine equation, utilizing measurements on a limited number of grain sizes and strain rates\(^{125}\). The following constitutive equations were obtained\(^{125}\):

\[ Z = \dot{\varepsilon} \exp\left(\frac{Q_{\text{def}}}{RT}\right) \quad (5.25) \]

\[ \sigma_f = \frac{f(\varepsilon)}{\alpha} \sinh^{-1}\left[\left(\frac{Z}{k}\right)^n\right] \quad (5.26) \]

where \( f(\varepsilon) = 1 - 0.5 \exp\left(-\varepsilon / \varepsilon_{ch}\right) \) and \( \varepsilon_{ch} = 0.5201 - 7.42 \times 10^{-4} T + 3.0 \times 10^{-7} T^2 \). The fitting parameters for these equations are presented in Table 5.1.

There are three major limitations of Equation (5.26):

(i) the reduction in flow stress due to dynamic recrystallization was not accounted for.

(ii) the effect of grain size on flow stress as observed in measurements was not included.

(iii) strain is not a fundamental variable of the hyperbolic sine equation, hence the use of the fitting strain function, \( f(\varepsilon) \).

In order to include the effect of grain size and dynamic recrystallization, a new set of constitutive equations based on measured flow stress data from UBC, NIST and CANMET were developed to cover the conventional cold charge rolling conditions for both A36 and DQSK steels\(^{117}\). These equations were rooted on the Voce formulation\(^{126}\) and the hyperbolic sine equation as follows:

\[ \sigma = \sigma_0 + \left(\sigma_{s1} - \sigma_0\right)\left(1 - \exp(-\beta \varepsilon)\right)^n \quad (5.27) \]

\[ \sigma_0 = c_0 \exp\left(c_1 \left(\frac{T_0}{T} - 1\right)\right)\left(\frac{\dot{\varepsilon}}{\dot{\varepsilon}_0}\right)^{d_1} \quad (5.28) \]

\[ \sigma_{s1} = \frac{1}{\alpha_1} \sinh^{-1}\left[c_2 \left(\dot{\varepsilon} \exp(Q_{s1} / RT)\right)^{n_1}\right] \quad (5.29) \]

\[ \beta = c_3 + c_4 \left(\frac{T_0}{T} - 1\right) + c_5 \left(\frac{\dot{\varepsilon}}{\dot{\varepsilon}_0}\right) + c_6 \left(\frac{d}{d_0} - 1\right) \quad (5.30) \]
where $\sigma_{ss1}$ is the measured steady-state stress or the apparent value obtained from curve-fitting of flow stress when steady-state is not attained. The reference values for temperature ($T_0$), strain rate ($\dot{\varepsilon}_0$) and grain size were taken to be 1273 K, 10 s$^{-1}$ and 0.08 mm respectively. In the case of dynamic recrystallization, the reduction in flow stress is computed from the following equations\textsuperscript{117}:

$$\Delta\sigma = (\sigma_{sv1} - \sigma_{ss2}) \left\{ 1 - \exp\left( -c_7 \left( \frac{\varepsilon}{\varepsilon^*} \right)^{n_4} \right) \right\} \tag{5.31}$$

$$\varepsilon^* = \exp \left( c_8 + c_9 \left( \frac{T_0}{T} - 1 \right) \right)(\dot{\varepsilon})^{n_5} d^{n_6} \tag{5.32}$$

$$\sigma_{sv2} = \frac{1}{\alpha_2} \sinh^{-1} \left[ c_{10} \left( \dot{\varepsilon} \exp(Q_d / RT) \right)^{n_7} \right] \tag{5.33}$$

where $\sigma_{ss2}$ is the measured steady-state value attained after the peak stress. The values of the fitting parameters for Equations (5.27) to (5.33) obtained for the two steels are listed Table 5.2\textsuperscript{117}.

Equations (5.27) to (5.33) were found to be valid for the range of operating conditions in the conventional finishing mill ($15 \leq \gamma \leq 250 \ \mu m, 900 \leq T \leq 1100 \ ^\circ C, 0 \leq \varepsilon \leq 1.0, 1 \leq \dot{\varepsilon} \leq 200 \ s^{-1}$). One important limitation of these equations is the assumption that the steady-state stress in both recovery ($\sigma_{ss1}$) and dynamic recrystallization ($\sigma_{ss2}$) conditions are independent of grain size. All the experimental data such as the one shown in Figure 4.6 indicate that this is not the case; values of steady-state stress ($\sigma_{ss1}$ and $\sigma_{ss2}$) are consistently higher for finer grain sizes than for coarser grains under the same deformation and softening conditions. Furthermore, the equations break down when the austenite grain size is greater or equal to 350 $\mu m$ at high temperatures and low strain rates ($\beta$ in Equation (5.30) exhibits a negative value at $Z \leq 1.33 \times 10^{14}$ and $Z \leq 1.47 \times 10^{14}$ for A36 and DQSK respectively, and consequently, Equation (5.27) becomes indeterminate), therefore cannot be applied to the first two stands of the CSP mill or the first stand of a
conventional roughing mill where the austenite grain size can range from 500 to 1400 \( \mu m \). Hence, a new set of constitutive equations has to be found for these conditions.

Based on curve-fitting of experimental flow stress curves for coarse grain sizes and a revision of the data for finer grain sizes, Equations (5.27) to (5.33) were modified. An attempt was made to reduce the number of fitting parameters and rationalize some of them. By applying Taylor’s equation \( \sigma = \alpha_0 M G \beta p^{0.5} \) to the fundamental expressions for hardening and recovery, Sellars\(^71\) has shown that \( n_0 \) in Equation (5.27) should have the value of 0.5, and this was adopted in the modified equations. The apparent activation energy \( Q_{def} \) in C-Mn steels has been shown to depend on the C, Mn and Si contents\(^73\) (Equation 2.9). Hence, the apparent activation energy is expected to be higher for A36 steel than for DQSK. Values of 330 and 315 kJ/mol were obtained for A36 and DQSK respectively, reflecting the trend obtained earlier for the two steels listed in Table 5.1.

The revised equations are as follows:

\[
\sigma = \sigma_0 + (\sigma_{ss1} - \sigma_0) \left(1 - \exp(-\beta \varepsilon)\right)^{0.5} \tag{5.34}
\]

\[
Z = \dot{\varepsilon} \exp\left(\frac{Q_{def}}{RT}\right) \tag{5.35}
\]

\[
f(d) = a_0 + b_0 \ln(d) \tag{5.36}
\]

\[
\sigma_0 = a_1 \ln(Z) * f(d) \tag{5.37}
\]

\[
\sigma_{ss1} = \frac{f(d)}{\alpha} \sinh^{-1}\left[k_i Z^n\right] \tag{5.38}
\]

\[
\beta = a_2 \ln Z * f(d) \tag{5.39}
\]

In the case of dynamic recrystallization, the reduction in flow stress is computed from an equation proposed by Sellars\(^71\) as follows:

\[
\Delta \sigma = (\sigma_{ss1} - \sigma_{ss2}) \left\{1 - \exp(-2.996 \left(\frac{\varepsilon - \varepsilon_c}{\varepsilon_s - \varepsilon_c}\right)^{3/2})\right\} \tag{5.40}
\]
\( \sigma_{\text{m}2} = \frac{f(d)}{\alpha} \sinh^{-1} \left[ k_2 Z^n \right] \)  

(5.41)

The use of the critical strain in Equation (5.40), which is obtained from measurement \((\varepsilon_c = 5 / 6 \varepsilon_p)\), eliminates the parameter \(\varepsilon^*\) utilized in Equation (5.32). The values of the fitting parameters for Equations (5.34) to (5.41) obtained for the two steels are listed in Table 5.3.

Employing Equations (5.34) to (5.41), the predicted flow stress curves are compared with the measured values for coarse grain sizes and selected values of finer grain sizes [computed with Equations (5.27) to (5.33)] in Figure 5.2. It is seen that there is good agreement between prediction and measurement; most predictions are within ±10 pct. of measurement. The predicted effects of temperature, strain rate and austenite grain size are shown in Figures 5.3 and 5.4 for A36 and DQSK respectively. It is seen that these equations correctly capture the expected and observed effects of temperature, strain rate and grain size. The flow stress at a given strain increases with decreasing temperature for a given austenite grain size and strain rate as seen in Figures 5.3 (a) and 5.4 (a) respectively. Furthermore, decreasing temperature retards the rate of softening. In contrast to the effect of temperature, higher strain rate increases the flow stress at a given temperature, strain and austenite grain size, and shift the curves away from dynamic recrystallization as in Figures 5.3 (b) and 5.4 (b) respectively. Figures 5.3 (c) and 5.4 (c) predict that the flow stress at a deformation condition decreases with increasing grain size provided that the softening mechanism remains the same; the tendency towards dynamic recrystallization being higher for finer grain sizes. The strength of these constitutive equations lies in their application to all the possible conditions operating in the CSP mill as well as the roughing and finishing stands of a conventional mill. In fact, further sensitivity analysis \((850 \, ^\circ C \leq T \leq 1300 \, ^\circ C, \, 0.1 \, s^{-1} \leq \dot{\varepsilon} \leq 200 \, s^{-1}, \, 5 \, \mu m \leq d_r \leq 5000 \, \mu m)\) indicated that the predictions of Equations (5.34) to (5.41) remain consistent and stable beyond the experimentally validated range.
5.1.2.2 Quantitative Description of Peak Occurrence in a Flow Stress Curve

As discussed in the last chapter (Section 4.2.4), a novel attempt was made in this study to quantitatively delineate the occurrence of peaks in flow curves for a given set of deformation conditions (temperature, strain rate and austenite grain size). This was achieved by establishing an empirical relationship between the temperature, strain rate and grain size, for the flow curves without a clear peak that exhibited negligible change of stress with strain ($\partial \sigma / \partial \varepsilon \approx 0$) at relatively large strain values. The following equation for the boundary strain-rate ($\dot{\varepsilon}_b$) was obtained for both steels:

$$\dot{\varepsilon}_b = \exp(a_3 - a_4 d_r - a_5 c / T)$$ (5.42)

The results are depicted in the form of Zener-Hollomon ($Z_b$) versus grain size plot for both steels in Figure 5.5.

$$Z_b = \dot{\varepsilon}_b \exp(Q_{def} / RT) = a_6 \exp(-a_7 d_r)$$ (5.43)

The fitting parameters for both equations are depicted in Table 5.4. The invaluable contribution and assistance of Bin Chau in the analysis of the flow curves and the consequent development of these equations is greatly appreciated and acknowledged. It is noted that Equation (5.42) provides a greater accuracy than (5.43) but it is not continuous through the range of grain sizes considered. Furthermore, the high value of $Z_b (>10^{10})$ makes relatively large variations negligible during curve fitting (e.g., $10^8$ is only 1 pct. of $10^{10}$). Equations (5.42) and (5.43) are the first attempt known to the author to quantitatively delineate the occurrence of peak in a flow curve for a given set of deformation conditions, and they carry important practical implications. In terms of deformation resistance during rolling, it is now only necessary to compute the reduction in flow stress due to dynamic recrystallization (Equations (5.40) and (5.41)) when the applied strain rate is lower than the boundary value ($\dot{\varepsilon}_b$), otherwise Equations
(5.34) to (5.39) are sufficient for the determination of the deformation resistance. With respect to the recrystallization kinetics, the equations predict the presence or absence of dynamic/metadynamic recrystallization in conjunction with the critical strain. Dynamic or metadynamic recrystallization will occur when the applied strain rate is lower than the boundary value \( (\dot{\varepsilon}_h) \) and the corresponding applied strain is greater than the critical strain; otherwise static recrystallization takes place at interstand locations. Hence, the flow chart for the computation of recrystallization is as depicted in Figure 5.6.

To fully appreciate the importance of this finding, a review of the common practice in deciding whether dynamic recrystallization is occurring during rolling or metadynamic recrystallization at the interstand reveals the inadequacy of the status quo. Literature tend to suggest that dynamic and metadynamic recrystallization will not occur only when the applied strain is less than the critical strain or if the calculated peak strain is ridiculously high \(^{30}\) (a rather subjective judgment). Utilizing this approach and the peak strain equation for both steels (see Table 5.5 below), dynamic and metadynamic recrystallization would have been predicted for strains greater than 0.45 when a 100 \( \mu \)m grain size of either A36 or DQSK steel is deformed at 1000 °C and strain rate of 30 s\(^{-1}\) \( (\varepsilon_c = 0.427 \text{ for A36 and } \varepsilon_c = 0.392 \text{ for DQSK}) \). However, a simple calculation based on Equation (5.42) for the given deformation conditions reveals that the boundary strain rate is 21 and 20.7 s\(^{-1}\) for A36 and DQSK respectively. Therefore, the flow stress curve for both steels do not exhibit a peak at the given conditions, rendering the computation of \( \Delta \sigma \) and other dynamic/metadynamic recrystallization parameters meaningless.

### 5.1.3 Microstructural Evolution

Equations (2.10), (2.12) and (2.13) were utilized to evaluate the fraction recrystallized, the time for 50% recrystallization, the peak strain, recrystallized grain size and grain growth after
recrystallization. Table 5.5 lists these equations and the values of the fitting parameters for the two steels, with due recognition given to Sun et al.\textsuperscript{118,119} who developed the recrystallization equations and Militzer et al.\textsuperscript{86} who developed the grain growth equations.

When there is incomplete recrystallization during the interpass period, a uniform softening method has been used and the kinetics are predicted assuming a single average microstructure with the effective strain of

$$\varepsilon_i = \varepsilon_i^0 + \lambda(1.0 - X_{\text{rec}})\varepsilon_{i-1}$$ \hspace{1cm} (5.44)

where $\lambda$ is a constant ($\approx 0.5$ for low carbon steel). $X_{\text{rec}}$ is the fraction recrystallized between passes $(i-1)$ and $(i)$.

The equation for 50 pct. recrystallization ($t_{0.5}$) listed in Table 5.5 which was developed from double-hit tests of finer grain sizes (< 250 \textmu m), was extrapolated to coarse grain sizes and compared with the measurements listed in Table 4.4. The results for both steels are presented in Figure 5.7, showing that the recrystallization kinetics obtained for fine austenite grains cannot be extrapolated to the coarse grain range of A36 (Figure 5.7 (a)) while this is not a problem for DQSK (Figure 5.7 (b)). It was found that the grain size exponent ($p$) in the equation

$$t_{0.5} = Ad_0^b e^q \varepsilon^r \exp\left(\frac{Q}{RT}\right)$$ \hspace{1cm} (5.45)

should decrease with increasing grain size in the coarse austenite range of A36, if $A$ and $Q$ and the exponents $q$ and $r$ remain constant. For example, a value of $p=1.2$ was found to fit the experimental measurement for 1110 \textmu m (A36) compared to the average value of $p=1.5$ reported earlier for 14 to 244 \textmu m grain size. This difference appears inconsequential by itself, but it should be noted that for $d_0=1110$ \textmu m, $d_0^{15}$ is eight times greater than $d_0^{12}$, resulting in a $t_{0.5}$ that is eight times the measured value. The variation of $p$ with grain size in the coarse austenite range ($> 250$ \textmu m) is proposed as follows:
5.2 Finite Element Deformation Model

This model utilizes an Eulerian finite-element deformation analysis based on the flow formulation method to predict the through-thickness strain and strain rates as well as the mill load. The Eulerian frame of reference implies the use of a coordinate system that is fixed in space, while the flow formulation refers to the dependence of deformation resistance (flow stress) on the rate of deformation. Visco-plastic material behaviour was assumed. The original code was developed by Kumar et al.\textsuperscript{122} and modified to its present form by Jin et al.\textsuperscript{63}

The governing equation for deformation was derived by applying the virtual work principle as follows:

$$\int_V \{\delta \varepsilon\}^T \{\sigma\} dV - \int_V \{\delta \varepsilon\}^T \{F\} dV - \int_{\Gamma} \{\delta \varepsilon\}^T \{t\} dS = 0 \quad (5.47)$$

For Von Mises type of visco-plastic materials, the constitutive equation describing the relationship between strain and stress can be written for an isotropic incompressible non-Newtonian fluid as\textsuperscript{122}:

$$\dot{\varepsilon}_{ij} = \frac{1}{2\mu} s_{ij} ; \quad s_{ij} = \sigma_{ij} - \delta_{ij}\nu ; \quad \nu = \sigma_{ii} / 3 \quad (5.48)$$

where the nonlinear viscosity ($\mu$) is given by\textsuperscript{122}:

$$\mu = \frac{\sigma_{y} + (\dot{\varepsilon} / \gamma \sqrt{3})^{\frac{1}{n}}}{\sqrt{3\dot{\varepsilon}}} \quad (5.49)$$

$n$ is the work hardening coefficient and $\gamma$ is the viscous coefficient which becomes infinity in the case of pure plastic flow. $\sigma_{y}$ is the uniaxial yield stress of the material.

The deviatoric stress $\{s\}$ is related to the stress $\{\sigma\}$ and the mean stress ($\sigma_m$) as follows\textsuperscript{122}:

\[ p = 3.021d^{-0.1289} \quad (5.46) \]
The deviatoric stress is also related to the strain rate for a linear fluid in the form:

\[
\{s\} = [D]\{\dot{\varepsilon}\}
\]

(5.51)

where \([D] = \begin{bmatrix} 2 & 0 & 0 \\ 0 & 2 & 0 \\ 0 & 0 & 1 \end{bmatrix}\)

The variation of the viscosity (\(\mu\)) with strain rate (Equation (5.49)) allows the application of Equation (5.51) to non-Newtonian (non-linear) fluids. Utilizing Equation (5.51), the governing equation can be rewritten as:

\[
\int \{\delta \dot{\varepsilon}\}^T [D] \{\dot{\varepsilon}\} dV = \int \delta \dot{\varepsilon} \dot{p} dV - \int \{\delta \nu\}^T \{F\} dV - \int \{\delta \nu\}^T \{t\} dS = 0
\]

(5.52)

By using the following velocity and pressure shape functions:

\[
\{v\} = \sum_{i=1}^{n} N_i^v a_i^v = [N^v] \{a^v\}
\]

(5.53)

\[
\{p\} = \sum_{i=1}^{n} N_i^p a_i^p = [N^p] \{a^p\}
\]

(5.54)

the resulting equations for the velocity and pressure fields can be written in a matrix form as:

\[
\begin{bmatrix}
\begin{bmatrix} K^v \\ K^p \end{bmatrix} & \begin{bmatrix} 0 \\ 0 \end{bmatrix}
\end{bmatrix}
\begin{bmatrix}
\{a^v\} \\
\{a^p\}
\end{bmatrix}
+ \begin{bmatrix}
\{f^v\} \\
\{0\}
\end{bmatrix} = 0
\]

(5.55)

where
\[ K_{ij}^v = \int_{\Omega} \left( \left[ \mathbf{L} \right] \left[ N_i^v \right] \right)^T \mu \left[ D \right] \left( \left[ \mathbf{L} \right] \left[ N_j^v \right] \right) dV \]  
(5.56)

\[ K_{ij}^p = \int_{\Omega} \left( \left[ \mathbf{M} \right]^T \left[ \mathbf{L} \right] \left[ N_i^p \right] \right)^T \left[ N_j^p \right] dV \]  
(5.57)

\[ f_{i}^v = -\int_{\Omega} [N]^T \{ F \} dV - \int_{\Gamma} [N]^T \{ t \} dS \]  
(5.58)

\[ a^v, a^p \] are the nodal velocities and pressures respectively.

Similarly, by assuming the temperature shape function to be

\[ T = \sum_{i=1}^{n_t} N_i T_i = [N]^T \{ T \} \]  
(5.59)

the corresponding finite-element equations for temperature governing equation (Equation (5.1)) can be expressed as

\[ [K^T \{ T \} + \{ f^T \} = 0 \]  
(5.60)

or in the expanded form as

\[ \sum_{n=1}^{n_T} k K_{ij}^T T_j + \sum_{n=1}^{n_T} \rho C_p K_{ij}^T T_j = \sum_{n=1}^{n_T} \oint_{\Gamma} \mathbf{N}_j \left( \frac{\partial N_j}{\partial x} I_x + \frac{\partial N_j}{\partial y} I_y \right) \{ T \} dS \]  
(5.61)

where

\[ K_{ij} = \int_{\Omega} \left( \frac{\partial N_i}{\partial x} \frac{\partial N_j}{\partial x} + \frac{\partial N_i}{\partial y} \frac{\partial N_j}{\partial y} \right) dxdy \]  
(5.61)

\[ K_{vij} = \int_{\Omega} \left( u N_i \frac{\partial N_j}{\partial x} + v N_i \frac{\partial N_j}{\partial y} \right) dxdy \]  
(5.62)

\[ f_{i} = \oint_{\Gamma} \dot{Q} \mathbf{N}_i dxdy \]  
(5.63)

The thermal force vector \( (f_i) \) which contains the heat generation term couples the mechanical deformation to heat transfer.
Since the spatial coordinates of the Eulerian reference frame are not associated with a specific volume of material, it is necessary to integrate the history upstream of its current spatial coordinates to determine the strain and stress. The strain rate distribution in the roll bite was obtained from the continuity equation:\(^{122}\)

\[
\dot{\varepsilon}_{ij} = \frac{1}{2} (v_{i,j} + v_{j,i}) 
\]

The strain distribution in the roll-bite was obtained by integration of the strain rate along the direction of the metal flow:\(^{122}\)

\[
\varepsilon = \int_0^t \dot{\varepsilon} dt 
\]

### 5.2.1 Boundary Conditions and Solution Procedure

Figure 5.8 illustrates the finite-element discretization of the roll-bite. The thermal boundary conditions are as discussed in section 5.1.1.1. The mechanical boundary conditions are discussed with reference to Figure 5.8.

At the entrance to the roll bite, \(ab\) in Figure 5.8,

\[
v = 0 
\]  \hspace{1cm} (5.66)

At the exit from the roll bite and along the strip centerline, \(ef\) and \(af\) respectively in Figure 5.8, Equation (5.66) also applies.

The free surfaces, \(bc\) and \(de\) in Figure 5.8, have no surface tractions. Along the arc of contact, \(cd\) in Figure 5.7, the strip velocity is set equal to the roll velocity (a no-slip assumption). To allow for frictional effects, a narrow layer of friction elements is introduced along the boundary with the outer nodes attached to the roll surface. Friction is simulated by relating the shear stress \(\tau\) associated with flow to the normal pressure \(p\) through the friction coefficient \(\mu\) as follows:
A constant value of 0.35 was assumed for the friction coefficient.

The flow equations were solved subject to the mechanical boundary conditions and utilizing the temperature boundary conditions already computed with the finite-difference rolling model. Hitchcock's equation (Equation (5.22)) was employed to account for roll flattenning. A non-symmetric frontal equation solver based on the Gaussian-elimination method was employed since the matrix \([K^T]\) is non-symmetric due to the presence of convective terms. Weighting functions were used to aid convergence during iteration of the thermal routine. The program was coded in FORTRAN and run on either the SGI work station or a pentium PC. An 300-element mesh was employed since further refinement did not affect the result significantly.

### 5.3. Runout Table Model

This model utilizes a semi-theoretical analysis of jet impingement and parallel flow boiling to compute strip temperature and the consequent \(\gamma \rightarrow \alpha\) phase transformation with finite-difference analysis. This model was developed by V.H. Hernandez\textsuperscript{45,47}, while the transformation equations were developed by R. Pandi and M. Militzer\textsuperscript{99,100,128,129}. The model also computes the yield and ultimate tensile strengths based on the structure/property equations listed earlier\textsuperscript{102} (Equations (2.19) and (2.20)). As in the finishing mill, the model has three modules; (i) the temperature, (ii) transformation and (iii) mechanical property modules.

#### 5.3.1 Heat Transfer in the Runout Table.

The governing equation in this case is given by:

\[
\rho C_p \frac{\partial T}{\partial t} = \frac{\partial}{\partial x} \left( k \frac{\partial T}{\partial x} \right) + \dot{q} 
\]  
(5.68)
Unlike the rolling model, the whole strip thickness from top to bottom surface was included in the solution procedure since the top and bottom cooling in the runout table can be very different. The strip was discretized into 100 through-thickness nodes and the numerical equations were solved by the Crank-Nicholson finite-difference scheme. A variable time step that depends on the cooling zone was utilized with the total number of time steps not exceeding 4000. The thermophysical properties of the steel strip in the various phases (austenite, ferrite, pearlite) are listed as functions of temperature in the Appendix A.

Equation (5.68) was solved subject to the initial condition:

\[ t = 0, \ 0 \leq x \leq L, \ T = T_0(x) \quad (5.69) \]

and boundary conditions:

\[ x = 0; \ -k \frac{\partial T}{\partial x} = h_0(T - T_{w,0}) \quad (5.70) \]

\[ x = L; \ -k \frac{\partial T}{\partial x} = h_L(T - T_{m,L}) \quad (5.71) \]

The heat-transfer coefficients at each location was calculated based on three cooling mechanisms: (i) impingement zone, (ii) parallel flow zone and (iii) air cooling/radiation zone.

(a) Heat Transfer in the Impingement Zone\textsuperscript{45,47} (Transition Boiling).

\[ h_{TB} = \frac{q_{TB}}{\Delta T_s + \Delta T_{sub}} \quad (5.72) \]

\[ q_{TB} = q_{1-s}F + q_{v-s}(1 - F) \quad (5.73) \]

\[ q_{1-s} = q_{NB} = q_{ip} \quad (5.74) \]

\[ q_{v-s} = q_{FB} \quad (5.75) \]

\[ F = \frac{L}{L_B} \quad (5.76) \]
\[ q_{ip} = m_p H_{lv} 2 \pi R_s \left( T_{ip} - T_{sat} \right) \frac{N}{A_{ns}} \]  
(5.77)

\[ q_{FB} = q_{FB} + \frac{3}{4} q_{rad} \]  
(5.78)

\[ L = \frac{\rho_l \delta c_0 (H_{lv} + \Delta H_{sub}) U_m (\delta c_0)}{q_{NB} \left( 1 - \rho_l (H_{lv} + \Delta H_{sub}) U_m (\delta c_0) / q_{NB} \right)} \]  
(5.79)

\[ L_B = \lambda = \frac{4 \pi}{A + \sqrt{A^2 + \frac{4 g (\rho_l - \rho_v)}{\sigma}}} \; ; \; A = \frac{1}{\alpha} \frac{\rho_l \rho_v}{\rho_l + \rho_v} \left( U_v - U_l \right)^2 \]  
(5.80)

\[ q_{FB} = 0.054 \times 10^6 u_j^{0.607} \left( 1 + 0.527 \Delta T_{sub} \right) \]  
(5.81)

(b) Heat Transfer in the Parallel Flow Region\textsuperscript{45,47}

Equations (5.72) to (5.79) above is applied with different L and q\textsubscript{FB}.

\[ L = \frac{\rho_l \delta c_0 (H_{lv} + \Delta H_{sub}) U_m (\delta c_0)}{q_{NB}} \]  
(5.82)

\[ \text{Nu}_x = \frac{q_{FB} x}{k_v \Delta T_s} = 0.0195 \left( \frac{\mu_f}{\mu_v} \right) \left( \frac{\beta (2 \overline{u}_x + 7)^{0.2}}{\overline{u}_x^{0.6} \; \text{Re}_x^{0.8} \; \text{Pr}_x^{1/3}} \right) \; , \; \text{for} \; \mu_p > \mu_w \]  
(5.83)

\[ \text{Nu}_y = \frac{q_{FB} x}{k_v \Delta T_s} = 0.0195 \left( \frac{\mu_f}{\mu_v} \right) \left( \frac{\beta (2 \overline{u}_y + 7)^{0.2}}{\text{Re}_x^{0.8} \; \text{Pr}_x^{1/3}} \right) \; , \; \text{for} \; u_p \leq u_w \]  
(5.84)

(c) Heat Transfer in the Air Cooling/Radiation Zone\textsuperscript{45,47}

\[ q = q_{rad} + q_{conv} \]  
(5.85)

q\textsubscript{rad} is computed as in Equation (5.11) while q\textsubscript{conv} is evaluated from the following correlations\textsuperscript{45,47}:

\[ u_p \geq u_{air} \; ; \; \text{Nu}_x = \frac{q_{conv} x}{k_{air} \Delta T_s} = \frac{\text{Re}_x^{1/2} \; \text{Pr}_x^{1/2}}{\left( \frac{10}{3} + \frac{20 r}{27 (\text{Pr} A_t)^{1/2}} \right)} \]  
(5.86)
or

\[ u_p \geq u_{avr}; \quad \text{Re}_x \geq 5 \times 10^3; \quad Nu_x = \frac{q_{\text{conv}} x}{k_{\text{air}} \Delta T_s} = 0.019(9 - 7r)^{0.2} \text{Re}_x^{0.8} \text{Pr}^{1/3} \]  \hspace{1cm} (5.87)

where \( r = 1 - u_x / u_p; \quad A_1 = 1 / (0.3 - 0.0074r) \)

5.3.2 Austenite Decomposition

(a) Transformation Start Temperature (T_s)

The \( \gamma \rightarrow \alpha \) is assumed to start when the transformation start temperature (T_s) is attained. \( T_s \) is calculated from the following equation:\(^{128}\):

\[
c^* - c^0 = \frac{(c_\gamma - c^0)\sqrt{2M}}{\Phi^{0.5} \partial y} \sqrt{r \int_0^\gamma D \frac{(c_i - c^0)}{(c_\gamma - c^0)} dT} \]  \hspace{1cm} (5.88)

The austenite to pearlite transformation is assumed to commence when the \( \alpha - \gamma \) interface velocity \( \left( \frac{dR_{\alpha,i}}{dt} \right) \) is less than the critical velocity for cementite nucleation:\(^{129}\)

\[
\frac{dR_{\alpha,i}}{dt} = \left\{ \left(1 - F_{\alpha,i+\Delta t}\right)^{-2/3} \frac{(F_{\alpha,i+\Delta t} - F_{\alpha,i})}{\Delta t} \right\} \]  \hspace{1cm} (5.89)

\[
\nu_{\alpha\beta} = 0.00318 \times 10^{12} T_{C_i} D_C \left\{ \ln \left( \frac{c_i}{c_p} \right) \right\}^2 \]  \hspace{1cm} (5.90)

(b) Growth of Transformed Phase

The kinetics of the growth of ferrite nuclei is given:\(^{129}\)

\[
X = \left[ 1 - \exp(-bt^{0.9}) \right] x \frac{c_\gamma - c^0}{c_\gamma - c_\alpha} \]  \hspace{1cm} (5.91)

where \( \ln(b) = (b_1 + b_2 d_y) (T_{A_1} - T) - (b_3 + b_4 d_y) \)

For the growth of pearlite nuclei, the following equation is used:\(^{129}\)
\[ X_p = \left[1 - \exp(-bt^{0.9})\right]x(1 - F_a) \]  \hspace{1cm} (5.92)

The growth kinetics is solved by assuming additivity rule during continuous cooling.

(c) Ferrite Grain Size

The equation for ferrite grain size is given by:\textsuperscript{129}:

\[ d_r = \left[ F \exp\left( B d_r^n - \frac{51000}{T_{\text{start}}} \right) \right]^{\frac{1}{n}} \]  \hspace{1cm} (5.93)

The fitting parameters for Equations (5.91) to (5.93) are listed for the two steels in Table 5.6

5.3.3 Mechanical Properties

Equations (2.19) and (2.20) are used for polygonal structures\textsuperscript{102} \((X_f \geq 60\%)\). The nitrogen in solution \((N_s)\) and the percent elongation were computed from\textsuperscript{130,131}:

\[ \%N_s = N_r - 1.28 \times 10^{-7} (T_{\text{coil}} - 500.0)^2 \]  \hspace{1cm} (5.94)

\[ \%\text{Elongation} = 55.0 - 0.05 \times UTS \]  \hspace{1cm} (5.95)

In the case of non-polygonal structure \((X_f < 60\%)\), the yield strength (YS in MPa) and ultimate tensile strength (UTS in MPa) of A36 steel were computed from\textsuperscript{131}:

\[ UTS = 447 + 200(1 - F_a) \]  \hspace{1cm} (5.96)

\[ YS = 0.7 \times UTS \]  \hspace{1cm} (5.97)
Table 5.1 Values of the fitting parameters in the first set of constitutive equations\(^{125}\) (Equations (5.25) and (5.26)).

<table>
<thead>
<tr>
<th>Steel Grade</th>
<th>k (\times 10^{11})</th>
<th>m</th>
<th>(Q_{\text{def}}) (kJ/mol)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A36</td>
<td>9.6946</td>
<td>0.2304</td>
<td>314</td>
</tr>
<tr>
<td>DQSK</td>
<td>1.312</td>
<td>0.2457</td>
<td>293</td>
</tr>
</tbody>
</table>

Table 5.2 Values of the fitting parameters in the second set of constitutive equations\(^{117}\) (Equations (5.27) to (5.33)).

<table>
<thead>
<tr>
<th>Parameter</th>
<th>(\alpha_1)</th>
<th>(\alpha_2)</th>
<th>(c_0)</th>
<th>(c_1)</th>
<th>(c_2)</th>
<th>(c_3)</th>
<th>(c_4)</th>
<th>(c_5)</th>
<th>(c_6)</th>
<th>(c_7)</th>
<th>(c_8)</th>
<th>(c_9)</th>
<th>(Q_{d1}) (J/mol)</th>
<th>(Q_{d2}) (J/mol)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A36</td>
<td>0.013</td>
<td>0.013</td>
<td>70.051</td>
<td>2.106</td>
<td>0.006</td>
<td>3.332</td>
<td>4.653</td>
<td>0.151</td>
<td>-0.942</td>
<td>1</td>
<td>-0.180</td>
<td>-2.080</td>
<td>333616</td>
<td>323001</td>
</tr>
<tr>
<td>DQSK</td>
<td>0.015</td>
<td>0.013</td>
<td>63.502</td>
<td>2.192</td>
<td>0.107</td>
<td>1.169</td>
<td>9.483</td>
<td>-0.006</td>
<td>-0.360</td>
<td>0.560</td>
<td>-0.760</td>
<td>1.154</td>
<td>333890</td>
<td>336678</td>
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</table>

Table 5.3 Values of fitting parameters in the new set of constitutive equations (Equations (5.34) to (5.41)).

<table>
<thead>
<tr>
<th>Parameter</th>
<th>(\alpha)</th>
<th>(a_0)</th>
<th>(a_1)</th>
<th>(a_2)</th>
<th>(b_0)</th>
<th>(e_s)</th>
<th>(k_1)</th>
<th>(k_2)</th>
<th>(m)</th>
<th>(n)</th>
<th>(Q_{\text{def}}) (kJ/mol)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A36</td>
<td>0.013</td>
<td>1.1</td>
<td>2.297</td>
<td>0.1458</td>
<td>-0.03</td>
<td>0.9305</td>
<td>1.695x10^{-2}</td>
<td>2.59x10^{-4}</td>
<td>0.1783</td>
<td>0.2812</td>
<td>330</td>
</tr>
<tr>
<td>DQSK</td>
<td>0.013</td>
<td>1.1</td>
<td>2.205</td>
<td>0.1426</td>
<td>-0.03</td>
<td>0.6083</td>
<td>1.137x10^{-4}</td>
<td>3.328x10^{-4}</td>
<td>0.3294</td>
<td>0.2875</td>
<td>315</td>
</tr>
</tbody>
</table>

Table 5.4 Values of fitting parameters for the boundary strain rate\(^{127}\) (Equations (5.42) and (5.43)).

<table>
<thead>
<tr>
<th>Parameter</th>
<th>(a_3)</th>
<th>(a_4)</th>
<th>(a_5)</th>
<th>(a_6)</th>
<th>(a_7)</th>
<th>(Q_{\text{def}})</th>
</tr>
</thead>
<tbody>
<tr>
<td>A36</td>
<td>38.167</td>
<td>1.547x10^{-2}</td>
<td>42747</td>
<td>1.072x10^{14}</td>
<td>330 kJ/mol</td>
<td></td>
</tr>
<tr>
<td></td>
<td>10.97</td>
<td>3.322x10^{-3}</td>
<td>9295</td>
<td>1.147x10^{-2}</td>
<td>330 kJ/mol</td>
<td></td>
</tr>
<tr>
<td>DQSK</td>
<td>10.22</td>
<td>8.59x10^{-3}</td>
<td>8058</td>
<td>9.338x10^{-14}</td>
<td>315 kJ/mol</td>
<td></td>
</tr>
<tr>
<td></td>
<td>11.723</td>
<td>5.153x10^{-4}</td>
<td>12819</td>
<td>7.726x10^{-3}</td>
<td>315 kJ/mol</td>
<td></td>
</tr>
</tbody>
</table>
Table 5.5 Microstructural equations and the fitting parameters for A36 and DQSK steels.

<table>
<thead>
<tr>
<th>Steel Grade</th>
<th>Static</th>
<th>Metadynamic</th>
</tr>
</thead>
<tbody>
<tr>
<td>A36</td>
<td>k = 2</td>
<td>k = 1</td>
</tr>
<tr>
<td>DQSK</td>
<td>k = 2</td>
<td>k = 1</td>
</tr>
</tbody>
</table>

II. Peak and Critical Strain:\[ \varepsilon_{p} = A_{1}d_{0}^{m_{1}} \varepsilon_{\eta}^{n_{1}} \exp \left( \frac{Q_{\text{ rex}}}{RT} \right), \quad \varepsilon_{c} = \frac{5}{6} \varepsilon_{p} \]

<table>
<thead>
<tr>
<th>Steel Grade</th>
<th>A1</th>
<th>m₁</th>
<th>n₁</th>
<th>p₁</th>
<th>Q_{\text{ rex}}</th>
</tr>
</thead>
<tbody>
<tr>
<td>A36</td>
<td>1.318</td>
<td>0.173</td>
<td>-</td>
<td>0.165</td>
<td>24327</td>
</tr>
<tr>
<td>DQSK</td>
<td>5.6 x 10</td>
<td>0.169</td>
<td>-</td>
<td>0.158</td>
<td>32965</td>
</tr>
</tbody>
</table>

III. Time for 50% Recrystallization (Metadynamic):\[ t_{0.5} = A_{1}d_{0}^{m_{1}} \varepsilon_{\eta}^{n_{1}} \exp \left( \frac{Q_{\text{ rex}}}{RT} \right) \]

<table>
<thead>
<tr>
<th>Steel Grade</th>
<th>A1</th>
<th>m₁</th>
<th>n₁</th>
<th>p₁</th>
<th>Q_{\text{ rex}}</th>
</tr>
</thead>
<tbody>
<tr>
<td>A36</td>
<td>2.13 x 10^{-6}</td>
<td>-</td>
<td>-</td>
<td>-0.68</td>
<td>132800</td>
</tr>
<tr>
<td>DQSK</td>
<td>9.38 x 10^{-2}</td>
<td>-</td>
<td>-</td>
<td>-3/4</td>
<td>-9380</td>
</tr>
</tbody>
</table>

IV. Time for 50% Recrystallization (Static):\[ t_{0.5} = A_{1}d_{0}^{m_{1}} \varepsilon_{\eta}^{n_{1}} \exp \left( \frac{Q_{\text{ rex}}}{RT} \right) \]

<table>
<thead>
<tr>
<th>Steel Grade</th>
<th>A1</th>
<th>m₁</th>
<th>n₁</th>
<th>p₁</th>
<th>Q_{\text{ rex}}</th>
</tr>
</thead>
<tbody>
<tr>
<td>A36</td>
<td>8.31 x 10^{-15}</td>
<td>1.5</td>
<td>-3/2</td>
<td>-1/3</td>
<td>263000</td>
</tr>
<tr>
<td>DQSK</td>
<td>5.22 x 10^{-13}</td>
<td>1.0</td>
<td>-0.68</td>
<td>-1/3</td>
<td>248000</td>
</tr>
</tbody>
</table>

V. Recrystallized Grain Size:\[ d_{\text{ rex}} = A_{1}d_{0}^{m_{1}} \varepsilon_{\eta}^{n_{1}} \exp \left( \frac{Q_{\text{ rex}}}{RT} \right) \]

<table>
<thead>
<tr>
<th>Steel Grade</th>
<th>A1</th>
<th>m₁</th>
<th>n₁</th>
<th>p₁</th>
<th>Q_{\text{ rex}}</th>
</tr>
</thead>
<tbody>
<tr>
<td>A36</td>
<td>88.96</td>
<td>0.369</td>
<td>-0.368</td>
<td>-</td>
<td>-28060</td>
</tr>
<tr>
<td>DQSK</td>
<td>88.96</td>
<td>0.369</td>
<td>-0.368</td>
<td>-</td>
<td>-28060</td>
</tr>
</tbody>
</table>

VI. Grain Growth:\[ d_{\text{ gg}} = \sqrt{d_{\text{ rex}}^{2} + 3y_{\text{ gb}} \cdot t \cdot D_{\text{ gb}} \cdot b^{2} / (kT)}, \quad D_{\text{ gb}} = D_{0} \exp \left( -\frac{Q}{RT} \right) \]

\[ D_{0} = 8.9 \times 10^{-5} \text{ m}^{2} / \text{s}, \quad k = 1.38054 \times 10^{-23} \text{ J/K}, \quad b = 2.58 \times 10^{-10} \text{ m}, \quad Q = 159920 \text{ J/mol} \]

<table>
<thead>
<tr>
<th>Steel Grade</th>
<th>y_{\text{ gb}} (J/m²)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A36</td>
<td>0.7</td>
</tr>
<tr>
<td>DQSK</td>
<td>0.8</td>
</tr>
</tbody>
</table>
Table 5.6 Fitting parameters for the transformation equations\textsuperscript{99,129} (Equations (5.91) to (5.93))

<table>
<thead>
<tr>
<th>Steel</th>
<th>$T_{Ae3}$ ($°C$)</th>
<th>$T_N$ ($°C$)</th>
<th>B</th>
<th>$b_1$</th>
<th>$b_2$</th>
<th>$b_3$</th>
<th>$b_4$</th>
<th>F</th>
<th>$\kappa$</th>
<th>M</th>
<th>n</th>
<th>$\eta$</th>
</tr>
</thead>
<tbody>
<tr>
<td>A36</td>
<td>824</td>
<td>760</td>
<td>52.3</td>
<td>3.96</td>
<td>0.022</td>
<td>5.35</td>
<td>0.066</td>
<td>0.8</td>
<td>2.01</td>
<td>2.1</td>
<td>0.0286</td>
<td>0.70</td>
</tr>
<tr>
<td>DQSK</td>
<td>883</td>
<td>840</td>
<td>52.0</td>
<td>2.54</td>
<td>0.006</td>
<td>2.15</td>
<td>0.023</td>
<td>0.96</td>
<td>2.04</td>
<td>1.9</td>
<td>0.0171</td>
<td>0.43</td>
</tr>
</tbody>
</table>
Figure 5.1 Schematic illustration of the application of the different models to the various sub-units of the CSP mill.
Figure 5.2  Comparison of measured and calculated flow stress for (a) A36 and (b) DQSK steels.
Figure 5.3 The predicted effects of (a) temperature, (b) strain rate and (c) austenite grain size on the flow stress of A36 steel.
Figure 5.4 The predicted effects of (a) temperature, (b) strain rate and (c) austenite grain size on the flow stress of DQSK steel.
Figure 5.5  Zener-Hollomon parameter as a function of grain size in (a) A36 and (b) DQSK steels, showing the boundary between peak and no peak domains in the flow stress.
Figure 5.6 Computational flow chart of recrystallization showing the application of the boundary strain rate ($\dot{\varepsilon}_b$).
Figure 5.7  Effect of grain size on the recrystallization kinetics of (a) A36 and (b) DQSK steels.
Figure 5.8 Finite-element discretization of the roll bite\textsuperscript{122}. 
In order to validate the model and laboratory results with mill measurements from an operating CSP plant, an industrial trial was carried out at HYLSA’s CSP mill in Monterrey, Mexico. There were four major objectives of the industrial trial:

(a) to conduct intermediate temperature measurements for the verification of the submodels.

(b) to acquire CSP slab and coil samples with corresponding mill data for further metallurgical analysis.

(c) to confirm CSP mill set-up information and dimensions.

(d) to observe and record typical CSP mill practices.

The original plan which was to monitor and measure rolling parameters during the production of three steel grades (AISI 1005, AISI 1018 and a HSLA-Nb grade) could not be carried out since HYLSA’s CSP mill produces steel grades with carbon content in the range of only 0.04 to 0.09. Furthermore, no HSLA-Nb was produced during the trial although two coil samples with the corresponding rolling data were acquired.

6.1 Description of HYLSA’s CSP Mill

The general layout of HYLSA’s CSP plant is depicted in Figure 6.1 while some technical data of the plant are listed in Table 6.1. The plant which is 442.5 m long (electric arc furnace to hot rolled coil storage area) started operation in February, 1995. HYLSA produces 16 grades of steel in the CSP mill as listed in Table 6.2, comprising of low carbon (0.05-0.09 C), electrical steels (0.4 -1.20 Si) and Nb-HSLA (0.02-0.055 Nb) grades.

A charge of preheated scrap (55%) and DRI (45%) from HYLSA’s proprietary HYL direct reduction process is melted in the DC electric arc furnace. Liquid steel (135 ton per heat) is
tapped into a ladle furnace after a 55-70 minutes sojourn in the furnace, where secondary refining utilizing argon stirring is carried out for 20 to 30 minutes. The ladle furnace is shuttled to the casting deck, where the liquid steel is poured into a tundish from the ladle turret. The tundish feeds the liquid steel to the SMS’s patented funnel-shaped mold via a submerged entry nozzle where the mold level is maintained at ± 2 mm with the aid of radioactive control and a hydraulic tundish stopper rod. Solidification starts at the meniscus level in the copper-silver alloy mold and continues down the tapered cavity into a seven-segment spray zone. The slab then passes through the withdrawal and straightener unit consisting of three pinch-roll sets with two internally cooled rollers. Just before entry to the tunnel furnace, the slab is cut into the desired length by a pendulum shear.

The tunnel furnace, 206 m in length, soaks and equalizes the slab (with an entry surface temperature of 950 to 1050 °C) to a uniform temperature of 1070 to 1150 °C from the head to the tail end and over the entire width, within 12 to 20 minutes. The slab surface temperature at the exit of the tunnel furnace is measured by a pyrometer located about 1.5 m before actual furnace exit. From the exit of the tunnel furnace, the hot slab traveling at a preselected mill entry speed of 0.3 to 0.5 m/s enters the high-pressure descaling unit which utilizes a pair of spraying headers at the top and bottom surfaces. At HYLSA, the first spray header is operated at a lower pressure (~315 bars) than the second one (380 bars). There are three water catchers and an exit scale sweeper (backwash spray) at the top surface.

After descaling, the slab enters the first stand of the six-stand four-high tandem finishing mill. The CSP rolling mill is equipped with a profile and flatness control (PFC) system and continuous variable crown (CVC) technology. The first three stands (F1 to F3), which are equipped with chrome iron rolls (chrome steel are also used), are 5.4 m apart and 800 mm in diameter while the last three stands (F4 to F6), which are indefinite chill iron rolls, are 5.3 m apart and 500 mm in
diameter. Roll cooling is accomplished by two spray headers on the entry side and three spray headers on the exit side for the top rolls of F1 to F3, the number of the entry headers remaining the same for F4 to F6 but with only two exit headers per roll. The bottom rolls have three spray headers at both entry and exit sides of F1 to F3, the number of headers at the exit side reducing to two for F4 to F6. Anti-peel coolant is also used in the roll gap at all stands. Although there are five interstand sprays, interstand cooling is applied only between F1-F2 and F2-F3 at all times. This practice was reported to minimize rolled-in scale defects resulting from secondary scale growth. It was also reported that this practice has the detrimental effect of reducing the target F6 exit temperature by about 20 °C. Increased strip entry speed and higher tunnel furnace exit temperature (used mainly for thin gauges) are employed to compensate for this temperature decrease. With respect to roll life, rolls are reground after about 40 km or one week of operation which correspond to about 800 μm roll wear for F1 to F4. However, it was found that F3 wears faster than all other rolls as a result of high load and secondary scale pick-up, requiring regrinding after every 10-25 km of operation. To ensure consistent flatness and profile, F5 and F6 has to be reground after a maximum roll wear of about 500 and 300 μm for thick and thin gauges respectively.

The strip exiting the last stand passes through the measuring station where the temperature, thickness, width and speed are measured, before entry to the runout table. The runout table which is 70 m in length consists of seven segments or banks, each bank having four cooling headers on the top surface. Spray cooling is utilized in the first two cooling headers of the first bank while the remaining headers are laminar with two jetlines each. The two spray headers, located 1.5 m apart, are inclined at about 35° from the vertical axis and operate with 16.6 l/s water flow rate and about 200 mm coverage thickness each. The banks are 900 mm apart while the laminar jets are 600 mm apart and operate with 19.4 l/s water flow rate per header. In terms
of process control, the laminar jets can be classified into three zones: (a) a normal zone where two headers (4 jetlines) are controlled by one valve or switch, (b) a micro or fine zone where each header can be individually switched on or off, and (c) a trim zone where each jetline can be individually switched on or off. On the bottom face, there are four jetlines corresponding to each header of the top surface and with similar controls.

With respect to the locations of the jetlines along the runout table, the two spray headers are followed by a single normal zone (4 jetlines on the top and 8 jetlines at the bottom) which in turn are followed by 12 micro or fine zones (24 jetlines on the top and 48 jetlines at the bottom), 5 normal zones (20 jetlines on the top and 40 jetlines at the bottom) and lastly, 8 trim zones (8 jetlines on the top and 16 jetlines at the bottom). Cross sprays located at the end of each bank and also the entry and exit of the runout table cooling zone, are employed to sweep water off the strip surface. The strip travels on 218 rollers (260 mm diameter) located 0.32 m apart (centerline to centerline) and rotating at 15 m/s. The maximum allowable strip speed in the runout table is about 13 m/s but the mill currently operates at a maximum speed of 11 m/s.

It was observed that the runout table cooling is very dynamic in nature, continuously changing from the head end of the strip until the coiling temperature is within ±15 °C of the target temperature. Furthermore, the strip surface is covered with water beyond the impingement point, until the nearest cross spray. The cross sprays located 300 mm from the last jetline in each bank, removes water from the strip surface, the efficiency of water removal is controlled by the water volume at the cross spray. The cross sprays are most efficient when all the headers before it are active, and least efficient when the nearest active header is further away. For coils less than 2 mm, it is common practice to turn on the first two spray headers and one or two headers of the trim bank. For these coils, water covers the strip surface for almost the entire cooling length of the runout table.
6.2 Temperature Measurements

Temperature measurements of strip surface were carried out at various locations in the mill with portable Minolta/Land CYCLOPS pyrometers which have been calibrated and tested in the mill before being used. The Minolta/Land CYCLOPS pyrometer is a dual-wavelength (two-colour) pyrometer which measures the ratio of infra-red energy emitted at two wavelengths, a feature that reduces the inaccuracies associated with changes in either the emissivity of the target material surface or that of the sight path between the sensor and the material. For the first few coils measured, only a few data points were obtained (with unknown frequency) due to the difficulty of focusing the pyrometer and writing down the temperature reading simultaneously, a limitation that arose from the fact that these pyrometers could not be connected to a data acquisition system. This problem was later solved by voice recording the pyrometer temperatures with a microcassette recorder. It is noted that these measurements are in addition to the tunnel furnace exit, F6 exit and coiler entry temperatures which are always measured and utilized in the on-line process control of the mill.

The interstand temperatures measured at various locations for six 2.54 mm strips of the same steel grade with similar thermal and reduction histories are displayed in Figure 6.2 in conjunction with the finishing temperatures. The mean temperature at each location and the associated standard deviation are listed in Table 6.3 (a). All the temperatures were measured halfway between two stands except for DS2-F1 which is measured at 2.16 m before F1. It is seen that the progressive decrease in temperature from mill entry to exit as well as the temperature decrease due to water cooling from descale sprays are captured in these measurements. After rebounding from the cooling effects of descaling, the surface temperature of the strip remains almost unchanged until after F3. Thereafter, there is a gradual decrease to the final exit temperature of 883 °C. Figure 6.3 shows some runout table temperature measurements at a location, 36.2 m
from the coiling pyrometer, in conjunction with finishing mill and coiling temperatures. The mean temperature at each location and the associated standard deviation are listed in Table 6.3 (b). At an average strip speed of 7 m/s and with only the first ten jetlines active, the figure suggests an average cooling rate of 30 to 40 °C/s in the active cooling zone (that is, between the run out table entry and the portable pyrometer) and 18 to 25 °C/s in the radiation/air cooling zone (that is, from the portable pyrometer to the coiler). The temperature measurements for two coils of 0.06C-0.18Mn (7061) steel grade is shown in Figure 6.4. The mean temperature at each location and the associated standard deviation are listed in Table 6.3 (c). It is seen that a temperature drop of 60 to 80 °C occurs between F4-F5 interstand and F6 exit, mainly from roll chilling and air radiation. The figure suggests an average cooling rate of 20 to 25 °C/s between the run out table entry and the portable pyrometer and 33 to 40 °C/s from the portable pyrometer to the coiler, noting that there are active jets before and after the portable pyrometer. In this case, the strip speed was 8 m/s.

To gain some understanding of the increase in water temperature as a function of distance from the impingement zone, water temperatures at the runout table were measured for 6 coils with type-K chromel-alumel thermocouples inserted into the water boxes opposite the cross sprays. The thermocouples were connected to a Toshiba T3200 data acquisition system and data was acquired at a frequency of 5Hz. The results are presented in Table 6.4. It is seen that the water heats up from its initial temperature of 36 °C to about 60 °C in a distance of 300 mm and up to 76 °C after 2700 mm. The importance of this lies in the fact that heat removal from strip by the cooling water decreases as the water temperature increases. It is envisaged that these measurements are good representations of the increase in water temperature from the preceding jet although the water reaching the thermocouple actually originate from three sources; (i) water from the last jet, (ii) water from the cross spray and (iii) a small amount of water from several
preceding jets. This is because the water flow rate from the cross spray is less than one tenth of that of the jets and it was observed that only a negligible fraction of water from a preceding jetline flows past the next jetline due to obstruction from back flow.

6.3 Data Acquisition and Analysis

Engineering log data and the more dynamic data from the mill ‘level 3 control’ were acquired for 55 coils. The engineering log records single values of relevant rolling and runout table cooling parameters (reduction, roll force, temperature, thickness, width, speed, active jets, etc.) in three installments: (i) precalculation or setup, (ii) post calculation or feedback and (iii) final adaptation. The dynamic ‘level 3 control’ data records all the measured parameters (thickness, width, temperature, roll force, speed, steel chemistry, etc.) in alternate 2 and 3 m intervals, for the entire strip length or at least, 100 m from the head end.

These data were analyzed to study the range and variability of mill parameters, and the effects of steel grade, strip speed, temperatures, thickness and width on mill operation and product quality. A summary of this analysis is listed in Table 6.5. Typical finishing exit and coiling temperatures have been presented in Figures 6.2 through 6.4. Figure 6.5 shows typical gauge and width performance while the combination of strip entry speed and entry temperature (measured at the tunnel furnace exit) is depicted in Figure 6.6. It is observed that a variability of 4 to 12 pct. is tolerated in the final strip thickness, this variability decreasing with increasing thickness. For the width, a maximum of 2 mm below the target width is allowed while a variability of 2.5 to 3.6 pct. is tolerated above the target width, this variability decreasing with increasing width. Starting from the tunnel furnace exit, the mill entry surface temperature was found to range from 1060 to 1150 °C, and decreases with increasing final strip thickness. The entry speed which range from 0.3 to 0.47 m/s is seen to decrease with decreasing strip thickness. The combination of higher entry temperature and high reduction per pass for thinner gauges
ensure that the desired FM exit temperature is attained. The low entry speed of thin strips compensates for the higher speeds emanating from high reductions such that the desired runout table speed is not exceeded.

Figure 6.7 shows the measured roll forces for three CSP strips, together with percent reduction at each stand. The mean value of the roll forces and the associated deviations are listed in Table 6.6. Figure 6.7 and Table 6.6 illustrate the link between reduction and roll force, the higher reduction and roll force being associated with the production of thin strip. It is observed that the standard deviation in the roll force measurement is higher for the thinner strips (1.54 and 2.54 mm) than for the thicker one (6.45 mm). A detailed analysis of the measured roll forces for most of the CSP strips revealed that a maximum of 5 to 10 pct. deviation from the mean value for a given coil is common, although the standard deviation of measurement is only about 2 pct. of the mean value at each roll stand.

6.4 The Challenge of Thin Strip Rolling - Cobbling and Strip Shape

The prospect of producing thin hot band coils (≤ 1 mm thickness) on CSP mills to compete with conventional cold rolled products is a major leap of faith that is being ardently pursued by most of the CSP mills. This can only be achieved by precise control of all hot rolling parameters to attain the stringent tolerances prescribed for cold rolled products. Cobbling is by far the biggest operational problem associated with production of thin strips in the CSP mill while the maintenance of acceptable strip shape and profile constitutes the major product quality challenge. Cobbling is the buckling of strip during rolling, leading to an uncontrolled formation of strip loops that makes further rolling impossible and is known to increase with increasing speed and decreasing thickness. Table 6.8 lists the cobbling rate and the origin of cobbling at HYLSA’s CSP mill in the month of July, 1996. This data is plotted in the form of bar charts in Figure 6.9. It is observed that almost all the cobbling occurred during the production of strips in the thickness
range of less than 2 mm. In fact, during the plant trial, the cobbling rate during the production of 1.24 mm strip was found to range from 20 to 40 pct. The usual mill practice is to extract the cobbled strip from the rolling line and to alter the rolling schedule to produce thicker strips for sometime before the next attempt at thin strip rolling. Table 6.7 and Figure 6.9 (b) indicate that cobbling results from a multitude of factors. For the month under review, strip buckling between stands, set up problems, looper control, coiler issues, mill tracking and automatic gauge control were the major causes of cobbling.

With respect to strip shape, it is more difficult to maintain tight tolerances in the profile and shape as the strip thickness decreases due to increased deformation inhomogeneity from higher reductions. In fact, the shape of the thinnest gauge produced at HYLSA’s CSP mill (a 0.91 mm thick strip) which was meant to replace equivalent rolled sheet was found to be inadequate for direct application that it has to be subjected to skin pass rolling before use. However, there is a microstructural and mechanical property advantage in producing very thin strips, the ferrite grain size can be refined to such an extent that the strength becomes closer to cold rolled strength. The microstructure at the centerline of the 0.91 mm strip after skin pass (skin pass does not effectively change the strip center since there is negligible center deformation involved) is compared with a 2.69 mm strip (the same 0.06C-0.3Mn steel grade) coiled at the same temperature in Figure 6.10. It is obvious that the final grain size of the 0.91 mm strip is finer than that of the 2.69 mm strip, values of 8.2 and 12 μm were estimated for the thin and thicker coils respectively. This refinement in grain size from 12 to 8.2 μm is expected to increase both YS and UTS by about 21 percent according to Equations 2.19 and 2.20.

6.5 Grain Size Evolution in the CSP Mill

In one case of cobbling witnessed during the plant trial, the cobbling occurred after a 100 m length of the strip from the head end had been coiled. In this particular case, the looper just
before F4 went out of control, raising the entire length of the looper arm such that big loops of strip started accumulating between F3 and F4. The data acquired before cobbling indicate that both the strip gauge and F6 exit temperature were out of range, the reduction schedule was producing a 3.02 mm strip instead of the targeted 2.57 mm with exit temperature of 859 °C instead of the 884 °C target. The strip was quenched in water (the descaler, interstand and roll cooling sprays were turned on at full blast to cool the strip quickly) before being extracted from the rolling line. Upon extraction from the rolling mill, the cobbled strip afforded a rare opportunity to study the grain size evolution in the CSP mill since the entrance and exit sections into all the stands were preserved. The detailed composition of this particular strip has been listed in Table 4.2 (the 0.062C-0.155Mn grade).

Specimens were cut from the cobbled strip and were utilized to investigate the microstructural evolution from the tunnel furnace exit to the runout table entrance. With respect to metallographic examination, the samples were ground, polished and etched as described in Chapter 4 to reveal the room temperature microstructure (ferrite, pearlite, bainite or martensite). The ferrite grain size was measured manually employing the standard ASTM planimetric or Jeffries’ procedure (ASTM Designation E 112), the measurement area was chosen to ensure that a minimum of 100 grains were counted in each case. The grain sizes from the manual measurement were found to be about 20 pct. coarser than the image analyzer measurements for the same specimen, a result of the limited number of grains counted compared to the image analyzer where at least 500 grains are utilized in the measurement. Hence, a 20 pct. reduction in the measured value was applied to any reported grain size where manual measurement was employed.

The room temperature microstructure which transformed from the as-cast slab at the entrance to the first stand (F1 entry) is displayed in Figure 6.11 (a), in conjunction with two other
slabs. Based on known characterization of ferritic microstructures\textsuperscript{98}, the micrographs suggest a mixture of Widmanstatten (lath or plate like) and quasi-polygonal ferrite (ferrite grains with irregular grain boundaries) that formed from the original solidified dendritic structure. It is very difficult to define a grain size in this case, although an estimate based on the quasi-polygonal ferrite structure was conducted. The estimated ferrite grain size for the cobbled 0.062C-0.155Mn (7061) slab is 43.3 μm, and 39.0 and 31.6 μm respectively for the 0.075C-0.325Mn (7092) and 0.071C-0.899Mn (7094) grades. The Widmanstatten to quasi-polygonal ferrite ratio in the microstructures appears to increase with increasing carbon and manganese contents. Attempt was made to reveal the prior as-cast austenite grain size utilizing various macroetching procedures. Satisfactory result was realized only in the case of the 0.071C-0.899Mn (7094) slab etched in 4 pct. picric acid, where an estimate of 990 μm was obtained.

The resulting strip surface and center microstructures at different locations in the mill for the cobbled strip are presented in Figures 6.12 and 6.13. The estimated grain sizes are listed in Table 6.8 and plotted in Figure 6.14. It is observed that the grain refinement that occurs from the mill entrance to exit is well manifested in these micrographs. The surface grain size decreased from the F1 entry value of 43.3 to 12.3 μm at F6 exit. At the strip center, the grain size reduced from the F2 entry value of 30.4 to 14.9 μm at F6 exit. The surface microstructure is expected to closely reflect the condition of austenite at the initiation of cobbling since the cobbled strip was quenched in water. The center would have taken more time to cool, particularly at locations where the strip is still relatively thick, and may have attained complete recrystallization and grown to a coarser austenite grain before cooling down to the transformation start temperature.

A striking feature of these micrographs is the extensive microstructural inhomogeneity at each stand exit, compared to a more homogeneous microstructure just before entry into the next stand, a situation that is so clearly reflected in the surface microstructures. Furthermore, the
average grain size at the exit of each stand is much smaller than the entry grain size into the next stand. This is thought to be a manifestation of the extent of recrystallization at each location. The fraction recrystallized at stand exit is minimal compared to the full recrystallization and some grain growth before entrance to the next stand. Transformation from the finer recrystallized austenite grains will result in a smaller ferrite grain size than that from a large unrecrystallized grain. However, the ferrite grain size from the coarse unrecrystallized grain is expected to be finer than that of an equivalent fully recrystallized grain due to enhanced strain-induced nucleation and associated grain refinement during transformation.

Another remarkable feature of the microstructures is the preservation of the Widmanstatten/quasi-polygonal ferrite structure until F2 exit, beyond which equiaxed polygonal ferrite is attained. The fact that the Widmanstatten/quasi-polygonal ferrite structure is associated with transformation from the original solidified dendritic structure seems to suggest that it takes the first two stands to break down the cast structure and subsequent stands to refine the resulting equiaxed microstructure through recrystallization as in the conventional rolling. This is a very important finding and has ramifications for CSP rolling practice. Firstly, it indicates that in sensitive grades where edge cracking occurs, the rolling conditions at F1 and F2 should be closely monitored to establish safe windows of operation. This might imply adding an additional edger between F1 and F2 in conjunction to the edger that is already being recommended ahead of F1 for these sensitive grades since it is known that a cast microstructure has a greater tendency towards edge cracking. Secondly, it sheds more light on the improvement of quality associated with increased number of stands in CSP rolling. The initial use of four stands implies that equiaxed grain refinement is only occurring at two stands, stands F3 and F4, which might not be enough for structural uniformity since it takes about twelve stands to achieve structural uniformity from reheated austenite in a conventional mill. This finding also has some implication for direct
strip casting, a reasonable amount of deformation is necessary to breakdown the solidified microstructure to attain the desired equiaxed grains of hot rolled strip or cold rolled sheet.

6.6 Microstructure and Mechanical Properties of CSP Steel Samples

Several coil samples were acquired during the industrial campaign for further metallurgical analysis in addition to the intermediate samples from the cobbled strip. The coils were obtained from the tail end immediately after coiling and allowed to cool to room temperature in air. The coil samples were employed in microstructural and mechanical property investigations. With respect to metallographic examination, the samples were ground, polished and etched as described in Chapter 4 to reveal the room temperature microstructure (ferrite, pearlite, bainite or martensite). The ferrite grain size was measured with the image analyzer employing the standard ASTM planimetric or Jeffries’ procedure (ASTM Designation E 112). In some cases, manual estimates were made as described in section 6.5.

The effects of coiling temperature, strip thickness and steel composition on the final ferrite grain size and mechanical properties of some CSP coils are listed in Table 6.9. The microstructures of some of these coils are shown in Figures 6.15 to 6.17. Figure 6.15 shows the final microstructure of two 0.06C-0.18Mn CSP strips coiled at 722 °C and 560 °C respectively. It is seen that while the coiling temperature of 722 °C produced polygonal ferrite, the coiling temperature of 560 °C resulted in a mixed polygonal and irregular ferrite. A mean ferrite grain size of 16.6 μm (ASTM 9.2) was measured for the strip coiled at 722 °C while a mean grain size of 9.0 μm (ASTM 11.0) was measured for the polygonal regions in the microstructure of the strip coiled at 560 °C. Intense cooling from 85 pct. of the headers was applied to achieve the coiling temperature of 560 °C while the coiling temperature of 722 °C was attained by utilizing the first few banks and a few headers of the trim banks (41 pct. of the headers).
Figure 6.16 depicts the final microstructures of two 0.06C-0.18Mn CSP strips (1.52 and 3.15 mm thick) coiled at 722 °C. It is observed that ferrite grain size of the 1.52 mm coil is definitely finer than that of the 3.15 mm coil; values of 12.6 and 16.6 μm were measured for the thin and thicker coils respectively. This is probably attributed to the high cooling rates attained in thinner strips for the same water flow rate at the runout table. Figure 6.17 displays the final microstructures of three CSP strips (0.06C-0.18Mn, 0.08C-0.35Mn and 0.06C-0.8Mn-0.02Nb grades), approximately 3 mm thick, and coiled between 600 and 615 °C. Grain sizes of 12.7, 14.3 and 11.9 μm were estimated for the 0.06C-0.18Mn, 0.08C-0.35Mn and 0.06C-0.8Mn-0.02Nb grades respectively. The amount of ferrite, pearlite and precipitates (in the case of HSLA-Nb) is observed to depend on the steel composition, there is a higher pearlite fraction in the microstructure with increasing carbon content.

The variation of microstructure across the width is presented in Figure 6.18 with the corresponding thickness measurements at five locations for a 2.69 mm strip of 0.06C-0.3Mn steel grade (2061) and a 6.35 mm strip of 0.07C-0.35Mn steel grade (2082). The measured thickness values indicate that the two strips were of simple convex strip profile, indicating a higher deformation at strip edges than at the center. The distribution of deformation across the width manifested in the strip microstructure where it is seen that the center of the two strips exhibited coarser grain size than the left and right edges. The grain size was estimated to be 8.6, 12 and 7.3 μm at the left edge, center and right edge respectively, for the 2.69 mm strip. Grain sizes of 8.2, 14.8 and 9.5 μm were estimated at the left edge, center and right edge respectively, for the 6.35 mm strip. The variation of microstructure across the thickness of CSP strips is displayed in Figure 6.19 for a 1.52 mm (0.06C-0.18Mn grade) and a 4.76 mm (0.07C-0.35Mn grade) strip. It is observed that there is no tangible through-thickness variation in these microstructures, a reflection of most CSP coils investigated. The grain size was estimated to be 11.5, 12.6 and 12.3
\[\mu m\text{ at the top surface, center and bottom surface respectively, for the 1.52 mm strip. Grain sizes of 14.8, 14.5 and 15.1 \mu m were estimated at the top surface, center and bottom surface respectively, for the 4.76 mm strip.}\]

The mechanical properties of the coils from the CSP plant trial were measured with the Instron machine according to the standard ASTM tension test procedure (ASTM Designation: A 370). Standard sheet-type specimens were machined from tail end samples that were cut from the CSP coil bundle immediately after coiling and cooled to room temperature in air. Both longitudinal (in the rolling direction) and transverse specimens were utilized. The tests were conducted at a cross-head speed of 6.35 mm per minute and the elongation were measured with a 1/2 inch extensometer that allows for a maximum of 50 percent extension. The load-elongation curves for these steels exhibited clear upper and lower yield stress points. The lower yield point was taken as the yield strength of the material in accordance with Section 13.1.2 of the ASTM A370 specification. The tensile strength was obtained by dividing the maximum load by the original cross-sectional area.

Table 6.9 list the measured values of yield stress, tensile stress and percent elongation for the CSP coils. It is observed that the yield and tensile strengths as well as the percent elongation are dependent on the coiling temperature, strip thickness and steel composition. The average values of some of the measured mechanical properties are presented in Figure 6.20, the error bars (2 pct. for YS, 1 pct. for UTS and 5 pct. for percent elongation) indicating the uncertainties obtained from reproducibility measurements made at UBC and the CSP plant.

It is observed that the yield and tensile strengths decrease with increasing coiling temperature while the percent elongation increases except in a few cases. This trend is shown in Figure 6.20 (a) where the average values of yield strength, tensile strength and percent elongation are plotted against the coiling temperature for three coils of 0.06C-0.18Mn grade with thickness
close to 3 mm. It is evident that by decreasing the coiling temperature from 722 to 560 °C, the YS and UTS increase by 20 and 47 MPa respectively while the percent elongation decreased by 5 pct. Similar trend was observed with respect to the effect of thickness as shown in Figure 6.20 (b) for three 0.06C-0.18Mn strips coiled at 722-726 °C; by decreasing the thickness from 3.15 to 1.52 mm, the YS and UTS increase by 54.5 and 47 MPa respectively while the percent elongation decreased by 5.7 pct. The tensile and yield strengths are also seen to increase with increasing alloying content while the percent elongation decreases as depicted in Figure 6.20 (c) for four 2.54-3.36 mm strips coiled at 600-615 °C. As expected, the HSLA-Nb steel exhibited the highest strength with moderate elongation due to precipitation hardening, the YS and UTS being 13 and 55 MPa higher than the strongest of the three plain carbon steels shown. Among the plain carbon steels, the YS and UTS increase by 55 and 34 MPa respectively while the percent elongation decreased by 6 percent as the carbon and manganese contents increase from 0.06C-0.18Mn to 0.08C-0.35Mn.

Figure 6.21 depicts a plot of the measured yield stress versus the inverse square root of the measured ferrite grain size for the five 0.06C-0.18 Mn (7061) coils listed in Table 6.9. It is observed that four of the five coils exhibited reasonable linear relationship between the yield stress and \(d^{-1/2}\), an indication that Hall Petch’s equation (Equation (2.18)) is applicable to these steels.

It is noted that the measured mechanical properties for strips coiled at relatively high temperatures (close to 700 °C) will likely be different from that of the cooled coil bundle, since the annealing effect of slow cooling is avoided in the specimens obtained immediately after coiling. Furthermore, transformation will continue at a slow pace in the coil bundle when it is not completed on the runout table. To gain some knowledge about this expected difference, tests were carried out at HYLSA to evaluate the change in mechanical properties for 2.54 to 3.15 mm strips of the 0.06C-0.18 Mn (7061 grade) coiled at 715 ± 15 °C. In these tests, samples were
removed from the tail end immediately after coiling and just before pickling for mechanical property evaluation. The results for 27 coils indicate a substantial difference in the mechanical properties, the samples taken at the pickling line exhibited on the average, a 25 percent decrease in yield strength, a 10 percent decrease in ultimate tensile strength and a 2 percent increase in percent elongation compared to the samples obtained immediately after coiling\textsuperscript{133}. For the particular case of the 3.15 mm thick strip of 0.06C-0.18Mn (7061) steel grade coiled at 722 °C (coil 140758 in Table 6.9), the measured yield strength listed in Table 6.9 dropped by 40 percent with a 12 percent decrease in UTS and a suprising 3 percent decrease in percent elongation.

Table 6.1 Some technical data for HYLSA’s CSP plant.

<table>
<thead>
<tr>
<th>CSP Melt Shop</th>
<th>CSP Soaking Furnace</th>
<th>CSP Rolling Mill</th>
</tr>
</thead>
<tbody>
<tr>
<td>Production Capacity</td>
<td>Length</td>
<td>Production Capacity</td>
</tr>
<tr>
<td>750,000 tons/year</td>
<td>206 m</td>
<td>1,500,000 tons/year</td>
</tr>
<tr>
<td>EAF Capacity</td>
<td>Fuel</td>
<td>Descaler</td>
</tr>
<tr>
<td>150 tons DC Furnace</td>
<td>Natural Gas</td>
<td>2 High-Pressure</td>
</tr>
<tr>
<td>Charge</td>
<td>No. of Zones</td>
<td>No. of Stands</td>
</tr>
<tr>
<td>55 % Scrap-45% DRI</td>
<td></td>
<td>6 (4-high)</td>
</tr>
<tr>
<td>Tapping</td>
<td>Tap-to-Tap Time</td>
<td>Strip Thickness</td>
</tr>
<tr>
<td>135 tons per heat</td>
<td>55 - 70 minutes</td>
<td>1 - 12.7 mm</td>
</tr>
<tr>
<td>Tap-to-Tap Time</td>
<td>Ladle Metallurgy</td>
<td>Strip Width</td>
</tr>
<tr>
<td>55 - 70 minutes</td>
<td>Bottom Argon Stirring</td>
<td>790 - 1350 mm</td>
</tr>
<tr>
<td>Ladle Metallurgy</td>
<td>LMF Processing Time</td>
<td>Work Roll Diameter</td>
</tr>
<tr>
<td></td>
<td>20 - 30 minutes</td>
<td>780 mm (F1-F3)</td>
</tr>
<tr>
<td>LMF Processing Time</td>
<td>CSP Caster</td>
<td>Runout Table Cooling</td>
</tr>
<tr>
<td></td>
<td>CSP Soaking Furnace</td>
<td>Top</td>
</tr>
<tr>
<td>CSP Caster</td>
<td>CSP Soaking Furnace</td>
<td>2 Spray Headers,</td>
</tr>
<tr>
<td>Mold Type</td>
<td>CSP Soaking Furnace</td>
<td>52 Laminar Jets</td>
</tr>
<tr>
<td>Funnel-Shaped</td>
<td>CSP Soaking Furnace</td>
<td>Bottom</td>
</tr>
<tr>
<td>Mold Material</td>
<td>CSP Soaking Furnace</td>
<td>Walking-Beam</td>
</tr>
<tr>
<td>Copper Silver Alloy</td>
<td>CSP Soaking Furnace</td>
<td>Type with Step Control</td>
</tr>
<tr>
<td>Mold Life</td>
<td>CSP Soaking Furnace</td>
<td>Maximum Coil Weight</td>
</tr>
<tr>
<td>~ 800 heats</td>
<td>CSP Soaking Furnace</td>
<td>20 tons</td>
</tr>
<tr>
<td>Number of Strands</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>1</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>Thin Slab Thickness</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>48 mm</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>Thin Slab Width</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>790 - 1350 mm</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>Tundish Capacity</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>28 tons</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>Mould Length</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>1100 mm</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>Casting Speed</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>2.5 to 6.0 m/min.</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>Metallurgical Length</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>6020 mm</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>Bending Radius</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>3000 mm</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>Slab Cutting</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>Pendulum Shear</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>Slab Length</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
<tr>
<td>Variable (≤ 50 m)</td>
<td>CSP Soaking Furnace</td>
<td></td>
</tr>
</tbody>
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135
Table 6.2 Nominal composition of steel grades produced at HYLSA’s CSP mill.

<table>
<thead>
<tr>
<th>Grade</th>
<th>% C</th>
<th>% Mn</th>
<th>% P</th>
<th>% S</th>
<th>% Si</th>
<th>% Nb</th>
<th>% Al</th>
<th>% Cu</th>
<th>% Ni</th>
<th>Application</th>
</tr>
</thead>
<tbody>
<tr>
<td>2061</td>
<td>0.06</td>
<td>0.20-0.35</td>
<td>0.015</td>
<td>0.012</td>
<td>0.10</td>
<td></td>
<td>0.025-0.045</td>
<td>0.10</td>
<td>0.10</td>
<td>Automotive</td>
</tr>
<tr>
<td>2082</td>
<td>0.05-0.08</td>
<td>0.30-0.45</td>
<td>0.015</td>
<td>0.012</td>
<td>0.10</td>
<td></td>
<td>0.025-0.045</td>
<td>0.10</td>
<td>0.10</td>
<td>Automotive</td>
</tr>
<tr>
<td>2085</td>
<td>0.05-0.08</td>
<td>0.75-0.90</td>
<td>0.015</td>
<td>0.012</td>
<td>0.10</td>
<td>0.02-0.03</td>
<td>0.025-0.045</td>
<td>0.10</td>
<td>0.10</td>
<td>Automotive</td>
</tr>
<tr>
<td>2092</td>
<td>0.05-0.09</td>
<td>0.30-0.45</td>
<td>0.015</td>
<td>0.012</td>
<td>0.10</td>
<td></td>
<td>0.025-0.045</td>
<td>0.20</td>
<td>0.10</td>
<td>Bearing Steel</td>
</tr>
<tr>
<td>2093</td>
<td>0.05-0.09</td>
<td>0.45-0.75</td>
<td>0.015</td>
<td>0.012</td>
<td>0.10</td>
<td></td>
<td>0.025-0.045</td>
<td>0.15</td>
<td>0.10</td>
<td>Structural</td>
</tr>
<tr>
<td>2095</td>
<td>0.05-0.09</td>
<td>0.75-0.90</td>
<td>0.015</td>
<td>0.012</td>
<td>0.10</td>
<td>0.02-0.03</td>
<td>0.025-0.045</td>
<td>0.15</td>
<td>0.10</td>
<td>API</td>
</tr>
<tr>
<td>7061</td>
<td>0.06</td>
<td>0.20-0.35</td>
<td>0.015</td>
<td>0.012</td>
<td>0.03</td>
<td></td>
<td>0.025-0.045</td>
<td>0.10</td>
<td>0.10</td>
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<tr>
<td>7067</td>
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<td>0.40-0.60</td>
<td>0.03</td>
<td>0.012</td>
<td>0.40-0.80</td>
<td></td>
<td>0.15-0.22</td>
<td>0.15</td>
<td>0.10</td>
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<td>0.25-0.45</td>
<td>0.04</td>
<td>0.012</td>
<td>0.0-1.20</td>
<td></td>
<td>0.025-0.045</td>
<td>0.15</td>
<td>0.10</td>
<td>Electric Motor</td>
</tr>
<tr>
<td>7082</td>
<td>0.05-0.08</td>
<td>0.30-0.45</td>
<td>0.015</td>
<td>0.012</td>
<td>0.03</td>
<td></td>
<td>0.025-0.045</td>
<td>0.10</td>
<td>0.10</td>
<td>Galvanized</td>
</tr>
<tr>
<td>7092</td>
<td>0.05-0.09</td>
<td>0.30-0.45</td>
<td>0.015</td>
<td>0.012</td>
<td>0.03</td>
<td></td>
<td>0.025-0.045</td>
<td>0.20</td>
<td>0.10</td>
<td>Galvanized Tube</td>
</tr>
<tr>
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<td>0.05-0.09</td>
<td>0.45-0.75</td>
<td>0.015</td>
<td>0.012</td>
<td>0.03</td>
<td></td>
<td>0.025-0.045</td>
<td>0.10</td>
<td>0.10</td>
<td>Galvanized</td>
</tr>
<tr>
<td>7094</td>
<td>0.05-0.09</td>
<td>0.85-1.00</td>
<td>0.015</td>
<td>0.012</td>
<td>0.20-0.30</td>
<td></td>
<td>0.025-0.045</td>
<td>0.15</td>
<td>0.10</td>
<td>API</td>
</tr>
<tr>
<td>7095</td>
<td>0.05-0.09</td>
<td>0.75-0.90</td>
<td>0.015</td>
<td>0.012</td>
<td>0.03</td>
<td>0.02-0.03</td>
<td>0.025-0.045</td>
<td>0.10</td>
<td>0.10</td>
<td>Galvanized</td>
</tr>
<tr>
<td>7096</td>
<td>0.05-0.09</td>
<td>0.80-0.95</td>
<td>0.015</td>
<td>0.012</td>
<td>0.10</td>
<td>0.04-0.055</td>
<td>0.025-0.045</td>
<td>0.15</td>
<td>0.10</td>
<td>API</td>
</tr>
<tr>
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<td>1.15-1.30</td>
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<td>0.012</td>
<td>0.10</td>
<td>0.04-0.055</td>
<td>0.025-0.045</td>
<td>0.15</td>
<td>0.10</td>
<td>API</td>
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</table>
Table 6.3. Average values of the measured intermediate temperatures and their associated standard deviation.

(a) Interstand locations (Figure 6.2)

<table>
<thead>
<tr>
<th>Location</th>
<th>DS1-F1</th>
<th>F1-F2</th>
<th>F2-F3</th>
<th>F3-F4</th>
<th>F4-F5</th>
<th>F5-F6</th>
<th>F6 Exit Pyro</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mean Temp., °C</td>
<td>931.3</td>
<td>941</td>
<td>940</td>
<td>939</td>
<td>933</td>
<td>913</td>
<td>883</td>
</tr>
<tr>
<td>(Standard Deviation)</td>
<td>(4.3)</td>
<td>(4)</td>
<td>(5.4)</td>
<td>(5)</td>
<td>(5)</td>
<td>(6)</td>
<td>(13.5)</td>
</tr>
</tbody>
</table>

(b) Runout table locations (Figure 6.3)

<table>
<thead>
<tr>
<th>Strip Identification</th>
<th>ID: 141856</th>
<th>ID: 141863</th>
<th>ID: 141864</th>
<th>ID: 141865</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mean F6 Exit Pyro, °C</td>
<td>884.3</td>
<td>887.6</td>
<td>883</td>
<td>869.2</td>
</tr>
<tr>
<td>(Standard Deviation)</td>
<td>(3.8)</td>
<td>(2.7)</td>
<td>(3.5)</td>
<td>(5.5)</td>
</tr>
<tr>
<td>Mean Temp. At 36.2 m from Coiler, °C</td>
<td>711.2</td>
<td>721.2</td>
<td>692.3</td>
<td>704.8</td>
</tr>
<tr>
<td>(Standard Deviation)</td>
<td>(9.2)</td>
<td>(11)</td>
<td>(9)</td>
<td>(7)</td>
</tr>
<tr>
<td>Mean Coiling Temp., °C</td>
<td>598.3</td>
<td>611.5</td>
<td>598.7</td>
<td>604.8</td>
</tr>
<tr>
<td>(Standard Deviation)</td>
<td>(19.1)</td>
<td>(8.8)</td>
<td>(14)</td>
<td>(10.6)</td>
</tr>
</tbody>
</table>

(c) Interstand and runout table locations (Figure 6.4)

<table>
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<th>ID: 141338</th>
</tr>
</thead>
<tbody>
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<td>Mean F4-F5 Temp., °C</td>
<td>990</td>
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</tr>
<tr>
<td>(Standard Deviation)</td>
<td>(8.1)</td>
<td>(4.6)</td>
</tr>
<tr>
<td>Mean F6 Exit Pyro, °C</td>
<td>926.6</td>
<td>910</td>
</tr>
<tr>
<td>(Standard Deviation)</td>
<td>(3.2)</td>
<td>(4.3)</td>
</tr>
<tr>
<td>Mean Temp. at 23.5 m from Coiler, °C</td>
<td>811</td>
<td>803</td>
</tr>
<tr>
<td>(Standard Deviation)</td>
<td>(8.6)</td>
<td>(5.8)</td>
</tr>
<tr>
<td>Mean Coiling Temp., °C</td>
<td>735</td>
<td>721</td>
</tr>
<tr>
<td>(Standard Deviation)</td>
<td>(5.5)</td>
<td>(13)</td>
</tr>
</tbody>
</table>

Table 6.4. Measured temperature of runout table cooling water for six 1204 mm wide strips.

<table>
<thead>
<tr>
<th>Strip Thickness (mm)</th>
<th>Strip Speed (m/s)</th>
<th>FM Exit Temp. (°C)</th>
<th>Active Headers</th>
<th>Distance from Last Active Jetline (mm)</th>
<th>Average Water Temperature (°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>3.05</td>
<td>6.06</td>
<td>899</td>
<td>S1, S2</td>
<td>2700</td>
<td>74.0</td>
</tr>
<tr>
<td>2.54</td>
<td>7.10</td>
<td>905</td>
<td>S1, S2</td>
<td>2700</td>
<td>76.0</td>
</tr>
<tr>
<td>2.54</td>
<td>7.17</td>
<td>912</td>
<td>S1, S2</td>
<td>2700</td>
<td>76.0</td>
</tr>
<tr>
<td>3.10</td>
<td>6.46</td>
<td>920</td>
<td>S1, S2</td>
<td>2700</td>
<td>71.7</td>
</tr>
<tr>
<td>2.50</td>
<td>7.12</td>
<td>884</td>
<td>S1,S2,N1,N2</td>
<td>300</td>
<td>68.0</td>
</tr>
<tr>
<td>2.49</td>
<td>7.06</td>
<td>905</td>
<td>S1,S2,N1,N2</td>
<td>300</td>
<td>60.6</td>
</tr>
</tbody>
</table>
Table 6.5 Summary of data analysis of various mill parameters.

<table>
<thead>
<tr>
<th>Measured Parameter</th>
<th>Range</th>
<th>Analysis</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tunnel Furnace Exit Temp. (°C)</td>
<td>1060-1160</td>
<td>Decreases with increasing final thickness. Close to 1150 for most coils less than 2.5 mm.</td>
</tr>
<tr>
<td>Mill Entry Speed (m/s)</td>
<td>0.29-0.47</td>
<td>Increases with increasing final thickness.</td>
</tr>
<tr>
<td>Roll Force (MN)</td>
<td></td>
<td>Highest load occurs at F3. Increases with increasing reduction and decreasing temperature.</td>
</tr>
<tr>
<td>Final Thickness (mm)</td>
<td>1.24 - 7.0</td>
<td>4 to 12 pct. deviation from target can be tolerated. Percent deviation increases with decreasing thickness.</td>
</tr>
<tr>
<td>Final Width (mm)</td>
<td>900 - 1300</td>
<td>A maximum of 2 mm below target is tolerated. 2.5 to 3.5 pct. above target width is tolerated. Percent deviation increases with decreasing width.</td>
</tr>
<tr>
<td>Mill Exit Temp. (°C)</td>
<td>870-942</td>
<td>A maximum of 55 °C above and 15 °C below target is acceptable.</td>
</tr>
<tr>
<td>RT Speed (m/s)</td>
<td>3 - 11.5</td>
<td>Increases with decreasing final thickness.</td>
</tr>
<tr>
<td>Coiling Temperature (°C)</td>
<td>530 - 730</td>
<td>± 15 °C deviation from target acceptable.</td>
</tr>
<tr>
<td>Coil Length (m)</td>
<td>630-1300</td>
<td>Depends on entry slab length and final thickness.</td>
</tr>
</tbody>
</table>
Table 6.6 Details of measured roll forces (the entry slab thickness is 48 mm in all cases).

| Strip 1: 7061 Steel Grade, Width = 1.167 m, $T_{\text{entry}} = 1152$ °C, $V_{\text{entry}} = 0.36$ m/s, $T_{\text{exit}} = 910$ °C |
|---|---|---|---|---|---|---|
| | F1 | F2 | F3 | F4 | F5 | F6 |
| Thickness (mm) | 21 | 9.8 | 5.8 | 3.4 | 2.1 | 1.54 |
| Reduction (%) | 56.28 | 53.13 | 41.51 | 41.55 | 36.59 | 27.78 |
| Mean RF/Width (kN/m) | 18124 | 16474 | 14901 | 11254 | 9859 | 8724 |
| Standard Deviation (kN/m) | 528 | 505 | 147 | 190 | 123 | 35 |
| Maximum Deviation (kN/m) | 1102 | 1045 | 451 | 440 | 633 | 111 |

| Strip 2: 2082 Steel Grade, Width = 0.948 m, $T_{\text{entry}} = 1098$ °C, $V_{\text{entry}} = 0.33$ m/s, $T_{\text{exit}} = 882$ °C |
|---|---|---|---|---|---|---|
| | F1 | F2 | F3 | F4 | F5 | F6 |
| Thickness (mm) | 22.5 | 10.9 | 6.0 | 3.7 | 2.54 | - |
| Reduction (%) | 53.17 | 53.36 | 44.88 | 38.08 | 31.80 | - |
| Mean RF/Width (kN/m) | 16065 | 19639 | 17238 | 11792 | 9236 | - |
| Standard Deviation (kN/m) | 201 | 641 | 520 | 1460 | 126 | - |
| Maximum Deviation (kN/m) | 511 | 1287 | 960 | 605 | 339 | - |

| Strip 3: 2082 Steel Grade, Width = 1.116 m, $T_{\text{entry}} = 1154$ °C, $V_{\text{entry}} = 0.42$ m/s, $T_{\text{exit}} = 897$ °C |
|---|---|---|---|---|---|---|
| | F1 | F2 | F3 | F4 | F5 | F6 |
| Thickness (mm) | 31.7 | 19.5 | 13.3 | 9.8 | 7.8 | 6.45 |
| Reduction (%) | 33.90 | 38.59 | 31.83 | 26.44 | 20.00 | 17.46 |
| Mean RF/Width (kN/m) | 11417 | 12968 | 11081 | 6790 | 5499 | 5317 |
| Standard Deviation (kN/m) | 131 | 170 | 111 | 46 | 43 | 48 |
| Maximum Deviation (kN/m) | 410 | 356 | 540 | 120 | 139 | 143 |

RF = Roll Force
Table 6.7 Cobbling statistics and causes for a month production at HYLSA’ CSP mill

<table>
<thead>
<tr>
<th>Cobbling Data</th>
<th>Identified Origin of Cobble</th>
<th>No. Of Cobbles</th>
<th>% of Total Cobbles</th>
<th>Thickness Range (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.91-1.09</td>
</tr>
<tr>
<td>Strip buckling at interstand</td>
<td></td>
<td>14</td>
<td>23.0</td>
<td>2</td>
</tr>
<tr>
<td>Set up problem</td>
<td></td>
<td>11</td>
<td>18.0</td>
<td>7</td>
</tr>
<tr>
<td>Looper control</td>
<td></td>
<td>8</td>
<td>13.1</td>
<td>1</td>
</tr>
<tr>
<td>Coiler related</td>
<td></td>
<td>7</td>
<td>11.5</td>
<td>1</td>
</tr>
<tr>
<td>Mill tracking</td>
<td></td>
<td>4</td>
<td>6.6</td>
<td>1</td>
</tr>
<tr>
<td>Automatic gauge control (AGC)</td>
<td></td>
<td>3</td>
<td>4.9</td>
<td>1</td>
</tr>
<tr>
<td>F6 flatness control</td>
<td></td>
<td>2</td>
<td>3.3</td>
<td>1</td>
</tr>
<tr>
<td>Strip deflection at the runout table</td>
<td></td>
<td>2</td>
<td>3.3</td>
<td>1</td>
</tr>
<tr>
<td>Strip stuck in the roll-gap</td>
<td></td>
<td>2</td>
<td>3.3</td>
<td>1</td>
</tr>
<tr>
<td>Problem with coiler pinch rolls</td>
<td></td>
<td>1</td>
<td>1.6</td>
<td>1</td>
</tr>
<tr>
<td>Test production</td>
<td></td>
<td>1</td>
<td>1.6</td>
<td>1</td>
</tr>
<tr>
<td>General operation problem</td>
<td></td>
<td>1</td>
<td>1.6</td>
<td>1</td>
</tr>
<tr>
<td>F5 entry guide</td>
<td></td>
<td>1</td>
<td>1.6</td>
<td>1</td>
</tr>
<tr>
<td>General control problem</td>
<td></td>
<td>1</td>
<td>1.6</td>
<td>1</td>
</tr>
<tr>
<td>F5 and F6 misalignment</td>
<td></td>
<td>1</td>
<td>1.6</td>
<td>1</td>
</tr>
<tr>
<td>Bending</td>
<td></td>
<td>1</td>
<td>1.6</td>
<td>1</td>
</tr>
<tr>
<td>Lubrication problem</td>
<td></td>
<td>1</td>
<td>1.6</td>
<td>1</td>
</tr>
<tr>
<td><strong>Total (Cobbles)</strong></td>
<td></td>
<td><strong>61</strong></td>
<td><strong>100</strong></td>
<td><strong>1</strong></td>
</tr>
<tr>
<td><strong>Total Production (No. of Strips)</strong></td>
<td></td>
<td><strong>3613</strong></td>
<td></td>
<td><strong>5</strong></td>
</tr>
<tr>
<td><strong>Cobbling Rate (%)</strong></td>
<td></td>
<td><strong>1.69</strong></td>
<td></td>
<td><strong>20</strong></td>
</tr>
</tbody>
</table>

Table 6.8 Measured grain sizes at various locations in the cobbled strip (0.062C-0.155Mn).

<table>
<thead>
<tr>
<th>Location</th>
<th>F1 Entry</th>
<th>F1 Exit</th>
<th>F2 Entry</th>
<th>F2 Exit</th>
<th>F3 Entry</th>
<th>F3 Exit</th>
<th>F4 Entry</th>
<th>F4 Exit</th>
<th>F5 Entry</th>
<th>F5 Exit</th>
<th>F6 Exit</th>
</tr>
</thead>
<tbody>
<tr>
<td>Surface $d_\alpha$ ($\mu$m)</td>
<td>43.3</td>
<td>20.5</td>
<td>28.9</td>
<td>9.8</td>
<td>12.8</td>
<td>11.6</td>
<td>8.7</td>
<td>8.9</td>
<td>13.5</td>
<td>5.8</td>
<td>12.3</td>
</tr>
<tr>
<td>Center $d_\alpha$ ($\mu$m)</td>
<td>28.5</td>
<td>30.4</td>
<td>14.5</td>
<td>17.9</td>
<td>11.9</td>
<td>15.6</td>
<td>9.0</td>
<td>12.0</td>
<td>9.8</td>
<td>14.9</td>
<td></td>
</tr>
</tbody>
</table>
Table 6.9  Measured ferrite grain size and mechanical properties* of CSP coils.

<table>
<thead>
<tr>
<th>Coil ID</th>
<th>Grade</th>
<th>F6 Exit Temp. (°C)</th>
<th>Coiling Temp. (°C)</th>
<th>Thickness (mm)</th>
<th>$d_a$ (μm)</th>
<th>YS (MPa)</th>
<th>UTS (MPa)</th>
<th>% Elong. (Average)</th>
</tr>
</thead>
<tbody>
<tr>
<td>140783</td>
<td>7061</td>
<td>920</td>
<td>560</td>
<td>3.15</td>
<td>9</td>
<td>318</td>
<td>397</td>
<td>30.8</td>
</tr>
<tr>
<td>140611</td>
<td>7061</td>
<td>880</td>
<td>615</td>
<td>2.54</td>
<td>12.7</td>
<td>316</td>
<td>379</td>
<td>34.3</td>
</tr>
<tr>
<td>140758</td>
<td>7061</td>
<td>900</td>
<td>722</td>
<td>3.15</td>
<td>16.6</td>
<td>299</td>
<td>350</td>
<td>35.73</td>
</tr>
</tbody>
</table>

(b) Effect of strip thickness

<table>
<thead>
<tr>
<th>Coil ID</th>
<th>Grade</th>
<th>F6 Exit Temp. (°C)</th>
<th>Coiling Temp. (°C)</th>
<th>Thickness (mm)</th>
<th>$d_a$ (μm)</th>
<th>YS (MPa)</th>
<th>UTS (MPa)</th>
<th>% Elong. (Average)</th>
</tr>
</thead>
<tbody>
<tr>
<td>141063</td>
<td>7061</td>
<td>910</td>
<td>722</td>
<td>1.524</td>
<td>12.6</td>
<td>354</td>
<td>395</td>
<td>30.0</td>
</tr>
<tr>
<td>141337</td>
<td>7061</td>
<td>914</td>
<td>726</td>
<td>2.16</td>
<td>14.9</td>
<td>328</td>
<td>384</td>
<td>35.6</td>
</tr>
<tr>
<td>140758</td>
<td>7061</td>
<td>900</td>
<td>722</td>
<td>3.15</td>
<td>16.6</td>
<td>299</td>
<td>350</td>
<td>35.7</td>
</tr>
</tbody>
</table>

(c) Effect of steel composition

<table>
<thead>
<tr>
<th>Coil ID</th>
<th>Grade</th>
<th>F6 Exit Temp. (°C)</th>
<th>Coiling Temp. (°C)</th>
<th>Thickness (mm)</th>
<th>$d_a$ (μm)</th>
<th>YS (MPa)</th>
<th>UTS (MPa)</th>
<th>% Elong. (Average)</th>
</tr>
</thead>
<tbody>
<tr>
<td>140611</td>
<td>7061</td>
<td>880</td>
<td>615</td>
<td>2.54</td>
<td>12.7</td>
<td>316</td>
<td>379</td>
<td>34.3</td>
</tr>
<tr>
<td>141120</td>
<td>2061</td>
<td>897</td>
<td>615</td>
<td>2.69</td>
<td>12.0</td>
<td>309</td>
<td>377</td>
<td>32.3</td>
</tr>
<tr>
<td>140905</td>
<td>7092</td>
<td>889</td>
<td>600</td>
<td>3.23</td>
<td>14.3</td>
<td>357</td>
<td>413</td>
<td>28.4</td>
</tr>
<tr>
<td>134455</td>
<td>2085</td>
<td>883</td>
<td>615</td>
<td>3.36</td>
<td>11.9</td>
<td>383</td>
<td>468</td>
<td>28</td>
</tr>
</tbody>
</table>

(c) Other coils

<table>
<thead>
<tr>
<th>Coil ID</th>
<th>Grade</th>
<th>F6 Exit Temp. (°C)</th>
<th>Coiling Temp. (°C)</th>
<th>Thickness (mm)</th>
<th>$d_a$ (μm)</th>
<th>YS (MPa)</th>
<th>UTS (MPa)</th>
<th>% Elong. (Average)</th>
</tr>
</thead>
<tbody>
<tr>
<td>140804</td>
<td>2061</td>
<td>891</td>
<td>715</td>
<td>3</td>
<td>13.1</td>
<td>306</td>
<td>371</td>
<td>33.4</td>
</tr>
<tr>
<td>141323</td>
<td>7092</td>
<td>883</td>
<td>665</td>
<td>1.57</td>
<td>10.4</td>
<td>335</td>
<td>398</td>
<td>27</td>
</tr>
<tr>
<td>140527</td>
<td>7092</td>
<td>886</td>
<td>665</td>
<td>4.762</td>
<td>14.5</td>
<td>353</td>
<td>407</td>
<td>28.7</td>
</tr>
<tr>
<td>140509</td>
<td>7094</td>
<td>880</td>
<td>665</td>
<td>7</td>
<td>-</td>
<td>375</td>
<td>441</td>
<td>27.9</td>
</tr>
<tr>
<td>139480</td>
<td>7097</td>
<td>890</td>
<td>615</td>
<td>7.6</td>
<td>-</td>
<td>412</td>
<td>503</td>
<td>32.7</td>
</tr>
</tbody>
</table>

* The mechanical properties (YS, TS, and % Elongation) were measured in both the longitudinal and transverse directions. The average values are listed.
2061 = 0.06C-0.30Mn, 7061 = 0.06C-0.18Mn, 7092 = 0.08C-0.35Mn, 7094 = 0.08C-0.90Mn, 2085 = 0.06C-0.80Mn-0.02Nb, 7097 = 0.06C-1.2Mn-0.04Nb
Figure 6.1  Schematic layout of HYLSA's CSP mill.
Figure 6.2 Measured temperatures at various locations during CSP rolling of 2.54 mm strips (0.07C-0.4Mn steel grade).
Figure 6.3  Measured temperatures at during runout table cooling of four CSP strips (0.07C-0.4Mn steel grade).
Figure 6.4 Measured temperatures during rolling and runout table cooling of two CSP strips (0.06C-0.18Mn steel grade).
Figure 6.5  Acceptable strip and width tolerances in CSP strips.
Figure 6.6  The combination of strip entry temperatures and speeds during CSP rolling.
Figure 6.7  Measured roll force per unit width during CSP rolling of three strips (1.54, 2.54 and 6.45 mm thick strips of 0.07C-0.4Mn steel), showing the effect of percent reduction on the roll force.
Figure 6.9 Cobbling in the CSP mill - (a) the effect of strip thickness on cobbling rate and (b) the causes of cobbling.
Figure 6.10 Comparison of the microstructure of the thinnest CSP strip (0.91 mm) with a 2.69 mm thick strip coiled at the same temperature (0.06C-0.3Mn 2061 steel grade).
Figure 6.11  Micrographs of three CSP slab samples showing the predominance of Widmanstatten (lath or plate like) and quasipolygonal ferrite (ferrite grains with irregular grain boundaries) that transformed from the original solidified dendritic structure.
Figure 6.12  Micrographs of the cobbled strip (0.062C-0.155Mn steel grade), showing the evolution of microstructure from F1 exit to F3 exit in the CSP mill (Figure 6.11 (a) is the F1 entry microstructure).
Figure 6.13  Micrographs of the cobbled strip (0.062C-0.155Mn steel grade), showing the evolution of microstructure from F4 entry to F6 exit in the CSP mill.
Figure 6.14  Measured ferrite grain size of the cobbled strip (0.062C-0.155Mn steel grade), showing the evolution of microstructure from F1 entry to F6 exit in the CSP mill.
Figure 6.15 Micrograph of two 0.06C-0.18Mn CSP coils (3.15 mm thick) showing the effect of coiling temperature on the final microstructure.
Figure 6.16 Micrograph of two 0.06C-0.18Mn CSP strips coiled at 722 °C showing the effect of thickness on the final microstructure.
Figure 6.17 Micrograph of three CSP strips (≈ 3 mm thick) coiled at 608±7 °C showing the effect of steel composition on the final microstructure.
Figure 6.18 Variation of microstructure and thickness across the width of two CSP strips.
Figure 6.19 The through-thickness microstructure of two CSP strips.

A. 1.54 mm thick strip of 0.06C-0.18Mn steel grade
B. 4.76 mm thick strip of 0.08C-0.35Mn steel grade

I. Top Surface
II. Center
III. Bottom Surface
Figure 6.20 Effect of (a) coiling temperature, (b) strip thickness and (c) steel composition on the mechanical properties of CSP coils.

2085 = 0.06C-0.8Mn-0.02Nb, 7092 = 0.08C-0.35Mn
7061 = 0.06C-0.18Mn, 2061 = 0.06C-0.3Mn
Figure 6.21  The effect of ferrite grain size on the measured yield strength of five 0.06C-0.18Mn coils. The linear fit indicates that four of the five coils obey Hall Petch's equation.
MODEL VALIDATION, RESULTS AND DISCUSSION

The mathematical models discussed in Chapter 5 were employed to predict the thermal, mechanical and microstructural evolution during CSP rolling. In the first part of this section, the CSP mill measurements discussed in Chapter 6 were utilized to validate model predictions of temperature, roll force, grain size and mechanical properties. The models were run with the exact mill setup and rolling schedules for each coil at HYLSA's CSP mill. Four CSP steel grades with carbon and manganese contents in the range of 0.06-0.075 and 0.16-0.35 respectively, were modeled as DQSK steel grade. All the calculations were made assuming a 1000 μm initial austenite grain size at the entry into the first stand, noting that an austenite grain size of 990 μm was measured at F1 entry for the 0.071C-0.9Mn slab.

In the second part, the models were employed to study the important factors influencing CSP rolling in order to elucidate the optimal processing route. In this case, the mill setup which is listed in Table 7.1, and the rolling schedules were provided by SMS (the inventors and suppliers of the CSP mill) and in some cases, extracted from literature. It is noted that HYSLA's CSP mill utilizes an extra descaler and two interstand sprays in addition to the mill setup listed in Table 7.1. The similarities and differences between CCR and CSP rolling in terms of the thermal history, deformation behavior and microstructural evolution were also highlighted. Two CSP steel grades, AISI 1018 and AISI 1005 were modeled essentially as A36 and DQSK respectively.
7.1 Model Validation

7.1.1 Validation of Temperature Predictions

Validation of the temperature predictions was carried out by comparing the predicted values with the CSP measurements discussed in Chapter six. The predicted F6 exit and coiling temperatures for ten CSP coils are listed in Table 7.2 and plotted in Figure 7.1. Good agreement is observed between model prediction and measurement, the difference between prediction and measurement is within the tolerance limit of 15 °C allowed in the CSP mill. It is also noted that the sum of the standard deviation and pyrometer measurement error (~1 pct. of reading) is approximately 15 and 25 °C for the F6 exit and coiling temperatures respectively, as listed in Table 6.3.

It was found that the uniform strain model consistently predicts lower temperatures than the target exit temperature for thin gauges, the discrepancy between prediction and target increasing as the thickness decreases. This was attributed to inadequate estimation of deformation heat when uniform strain distribution is assumed, and this will be discussed further in the following sections. The F6 exit temperatures predicted for coils less than 3 mm thick in Table 7.2 were corrected with finite-element predictions of heat of deformation in the roll bite. Figure 7.2 compares the model predicted F6 exit temperatures with pyrometer measurements for a 1.524 and 6.35 mm thick strip. It is observed that the F6 exit temperature predicted by the uniform strain model matches quite well with measurement for the 6.35 mm strip while the prediction for the 1.524 mm strip is 60 °C lower than the measured value. A good agreement is obtained between prediction and measurement for the 1.524 mm strip when the heat of deformation from inhomogeneous deformation is included in the calculations.

With respect to runout table cooling, the predicted thermal profiles for two 3.15 mm thick strips (0.06C-0.18Mn) are compared with the runout table entry and coiling temperature.
measurements in Figure 7.3, illustrating the typical cooling pattern for high and low coiling temperatures. It is also seen that the temperature predictions agree quite well with pyrometer measurements. Eight intermediate measurements made during the rolling of 2.54 mm strips (0.07C-0.4Mn steel grade) from a single heat are compared with model predictions in Figure 7.4. It is observed that the model prediction agree reasonably well with measurements when the heat of deformation from inhomogeneous deformation is accounted for.

An estimate of the heat extraction from the various mill sub-units was conducted from these calculations and measurements. The results are presented in the form of a pie chart in Figure 7.5. It was found that the descaling unit accounts for 4-6 percent of the total heat extraction, each spray header effectively extracting 6-9 °C from the strip despite the huge temperature drop at the surface. The two interstand sprays accounts for 3-4 percent of the total heat extraction, each spray header effectively removing 4-5 °C from the strip. The heat loss to the six rolls was 41-44 percent of the total, each roll effectively removing 17-23 °C from the strip despite the extensive chilling at the surface. Radiation constitutes the dominant heat loss by the strips, accounting for 48-51 percent of the total. This finding is comparable to that of Hollander\textsuperscript{25} who found that radiation accounts for 50 percent of the total heat loss in a conventional five-stand roughing and six-stand finishing mill compared to the 38 percent lost to the work rolls with the remaining 12 percent extracted by the cooling water. Chen et al.\textsuperscript{20} also estimated a 33 percent heat loss to the work rolls in a seven pass roughing mill.

### 7.1.2 Validation of Roll Force Predictions

The roll force per unit width, predicted by the finite-element and Sims models are compared with the measurements for three CSP strips (listed in Table 6.6) in Figure 7.6. It is observed that the predictions agree reasonably well with measurement except in the first stand, considering that a maximum deviation of about 10 pct. is associated with the measurements. The finite-element
predictions are only marginally better than the simpler Sims model in this case. The fact that the calculated values are lower than the measurements in the first stand, is an indication of a higher deformation resistance than the model assumed. This could be due to three possible factors:

(i) the entry grain size is finer than the 1000 \( \mu \text{m} \) assumed in the calculations

(ii) increased hot strength from the residual contents of the steels

(iii) there are dark sports on the strip surface emanating from the descaling unit.

### 7.1.3 Validation of Ferrite Grain Size Predictions

The measured ferrite grain size for ten CSP strips are compared in Table 7.3 with predicted values at the center of the strip, and plotted in the form of a bar chart in Figure 7.7. It is seen that the ferrite grain size is affected by the coiling temperature, strip thickness and steel composition. There is reasonable agreement between prediction and measurement, considering that the prediction was based on the assumption that the transformation behavior of these steels is exactly the same as the DQSK steel grade. It was shown from the continuous-cooling-transformation test (Figure 4.14) that these steels are different from DQSK, transformation starts and completes at lower temperatures in the 0.062C-0.155Mn (7061) and 0.075C-0.325Mn (7092) grades when compared to DQSK under the same conditions. Furthermore, a 100 percent polygonal structure is predicted for all cooling condition, a situation that was found to be true in laboratory measurements of DQSK steel\(^{101}\). Figure 6.15 indicates that this is not the case in these steels, the coiling temperature of 560 °C utilizing 85 pct. of the available jets was found to produce a mixture of polygonal and non-polygonal structures. These differences in the transformation behavior of these steels from DQSK reflect the higher carbon and alloy contents of the steels, and is expected to influence the ferrite grain size.

It is noted that the ferrite grain size prediction was based on Equation (5.93) which indicated that the ferrite grain size is controlled mainly by the transformation start temperature (\( T_s \)) with
minor contribution from the austenite grain size, the transformation start temperature being a
function of both cooling rate and austenite grain size according to Equation (5.88). The predicted
austenite grain size for these coils varied from 20.9 to 32.4 μm, a range that has a negligible
effect on the ferrite grain size prediction. Hence, the cooling pattern that produces the lowest
transformation start temperature is seen to result in the finest ferrite grain size.

Figure 7.7 (a) and (b) illustrate the measured and predicted effect of coiling temperature on
the ferrite grain size. Details of the predicted coiling temperature, transformation start
temperature and cooling rate are listed in Table 7.3 (a). The predicted ferrite grain increased from
11.9 to 15.1 μm as the coiling temperature is raised from 615 to 722 °C for the 0.06C-0.18Mn
steel grade. It is observed that ferrite grain size predicted for the strip coiled at 560 °C (15.3 μm)
is similar to that coiled at 722 °C, an indication of the comparable transformation start
temperatures of the two coils (812 and 813 °C). This similarity in the predicted ferrite grain size
resulting from comparable transformation start temperatures is also manifested in the two 0.06C-
0.3Mn strips coiled at 615 and 715 °C where ferrite grain sizes of 13.9 and 13.6 μm respectively,
were predicted. Ferrite grain size measurements for the two 0.06C-0.3Mn strips confirm this
trend, there is only a difference of 1 μm between the two coils.

Figure 7.7 (c) and (d) illustrate the measured and predicted effect of strip thickness on the
ferrite grain size. Again, details of the predicted coiling temperature, transformation start
temperature and cooling rate are listed in Table 7.3 (b). For the three 0.06C-0.18Mn strips coiled
between 715 and 726 °C, the predicted ferrite grain size decreased from 15.1 to 10.1 μm when
the strip thickness was reduced from 3.15 to 1.52 mm, reflecting the higher cooling rate of the
thinner strip (424.7 compared to 82.6 °C/s) which resulted in a lower transformation start
temperature (794.2 compared to 813.1 °C). The predicted ferrite grain size decreased from 15.6
to 11.2 μm when the strip thickness was reduced from 4.76 to 1.57 mm for the two 0.08C-0.35Mn strips coiled at 665 °C. In this case, the cooling rate at the start of transformation increased from 23.4 to 217.1 °C/s while the transformation start temperature decreased from 812.9 to 801.6 °C. Hence, it is seen that thinner coils attain higher cooling rates although they are usually cooled with fewer jetlines in the runout table. The measured and predicted ferrite grain sizes compare favorably.

Figure 7.7 (e) compares the predicted and measured ferrite grain size for three strips of different steel grades coiled at 600 to 615 °C, with details listed in Table 7.3 (c). It is noted that the model prediction did not take into account the difference in composition of the steels, hence, the predicted ferrite grain size simply reflects the change in the transformation start temperature. The ferrite grain size is expected to decrease as carbon and manganese contents increase, as reflected in the slight decrease in the measured grain size for the 0.06C-0.3Mn (12 μm) compared to the 0.06C-0.18Mn grade (12.7 μm). The fact the measured grain size of the 0.08C-0.35Mn grade is coarser than both the 0.06C-0.3Mn and 0.06C-0.18Mn grades is due to the reduced cooling rate (65 °C/s versus 154 and 227 °C/s) and consequently, a higher transformation start temperature (814 °C versus 809 and 799 °C). Furthermore, the 0.08C-0.35Mn strip is slightly thicker than the other two (3.23 mm compared to 2.69 and 2.54 mm).

To gain a further understanding of the occurrence of non-polygonal structure, as seen in microstructure of the 0.06C-0.18Mn strip coiled at 560 °C (Figure 6.15), the calculated cooling rate and ferrite fraction transformed at the strip center were compared for the two cooling patterns presented in Figure 7.3. The results are depicted in Figure 7.8. For the high coiling temperature (i.e. 722 °C), it is seen that transformation started at the cooling rate of 82.6 °C/s and decreased substantially from this value until transformation was completed. This allowed the growth of ferrite nuclei at fairly high temperatures and resulted in the coarse microstructure (16.6
μm). On the other hand, for the lower coiling temperature (560 °C), the cooling rate increased from its value of 86.1 °C/s at the start of transformation to over 100 °C/s at its completion, leading to Fe₃C nucleation and a non-polygonal microstructure. It is also noted that the finishing temperature for the strip coiled at 560 °C was 20 °C higher than the finishing temperature for the strip coiled at 722 °C, leading to a slightly coarser austenite grain size (31.7 compared to 29.9 μm) which contributed to non-polygonality.

Attempt were also made to simulate the variation of microstructure across the width as observed in Figure 6.18. It was found that in addition to the possibility of a colder strip at the entrance of the runout table, the combination of increased cooling (from water running off the strip edge in conjunction with top and bottom cooling), and the reduced thickness and austenite grain size (from the relative higher edge deformation) leads to a finer ferrite grain size at the strip edges. For example, the 28 percent decrease in the ferrite grain size at the left edge of the 2.69 mm thick strip (0.06C-0.3Mn steel grade) shown in Figure 6.18 (a) when compared to the center, can be simulated with a 30 °C decrease in the entry temperature, a 50 % increase in the water flow rate (simulating higher cooling) coupled with a 0.2 mm and 5 μm decrease in the strip thickness and austenite grain size respectively.

7.1.4 Validation of Predicted Mechanical Properties

Table 7.3 and Figures 7.9 and 7.10 compare the predicted and measured mechanical properties for ten CSP coils. Good agreement between prediction and measurement is observed. The predicted mechanical properties listed in Table 7.3 were computed by taking into account the specific compositions of the steels (including residual elements), employing published data on the effect of such elements as Ni, Cr, Mo, Cu and Sn which has been found to contribute up to 55, 90, 14, 78 and 170 MPa per weight percent respectively, to the room temperature yield strength of as-rolled and normalized low carbon steels. These upper limit values were adopted in the
estimation of the effect of residual elements on the mechanical properties of CSP steels, noting that these steels have 3 to 12 times more copper, 2 to 7 times more tin, and 1.5 to 3 times more nickel than the DQSK steel grade. The calculations indicate a maximum of 10 percent increase in yield strength, a value that was also assumed to apply to the UTS. With respect to the alloying elements that were implicitly included in the model (C, Mn, Si, P, N), it was found that the strength increase from the higher carbon and silicon contents of the CSP steels was compensated by the lower phosphorous and manganese (in the case of the 0.06C-0.18Mn) contents. There is a maximum of 8 MPa increase in strength from the nitrogen in solution since the nitrogen content of these coils ranged from 0.0052 to 0.007 weight percent. Hence the YS and UTS values listed in Table 7.3 are 10 pct. higher than that calculated for DQSK steel.

Figure 7.9 (a) and (b) illustrate the measured and predicted effect of coiling temperature on the yield strength (YS) and ultimate tensile strength (UTS), with details in Table 7.3 (a). There is reasonable agreement between prediction and measurement. The predicted YS increased from 287 to 297 MPa as the coiling temperature is decreased from 722 to 615 °C for the 0.06C-0.18Mn steel grade while the predicted UTS increased from 360 to 382 MPa, resulting in a one percent decrease in elongation (Figure 7.10 (a)). The YS and UTS predicted for the coiling temperature of 560 °C is expected to be lower than measurement since they were based on a polygonal ferrite grain size of 15.32 μm which is different from the mixed polygonal/non-polygonal structure observed in the strip (Figure 6.15). The non-polygonal structure also accounts for the relatively low elongation of this coil. For the two coils of 0.06C-0.3Mn grade, the predicted YS increased from 307 to 326 MPa as the coiling temperature is decreased from 715 to 615 °C while the predicted UTS increased from 378 to 387 MPa, resulting in a 0.5 percent decrease in elongation (Figure 7.10 (b)). It is seen that the increase in the strength of the strips coiled at lower temperatures results mainly from the contribution of nitrogen in solution.
There is 83.2, 62.9 and 61 percent Ns for predicted coiling temperatures of 557, 609 and 613 °C respectively, compared to zero percent at temperatures higher than 700 °C.

Figure 7.9 (c) and (d) illustrate the measured and predicted effect of strip thickness on the yield strength (YS) and ultimate tensile strength (UTS), with details listed in Table 7.3 (b). It is observed that the predicted values compare favorably with measurements. For the three 0.06C-0.18Mn strips coiled between 715 and 726 °C, the predicted YS increased from 287 to 298 MPa as the strip thickness decreased from 3.15 to 1.52 mm for the 0.06C-0.18Mn steel grade while the predicted UTS increased from 360 to 378 MPa, resulting in a one percent decrease in elongation (Figure 7.10 (c)). The predicted YS increased from 285 to 355 MPa as the strip thickness decreased from 4.76 to 1.57 mm while the predicted UTS increased from 366 to 398 MPa mm for the two 0.08C-0.35Mn strips coiled at 665 °C, resulting in a 0.7 percent decrease in elongation (Figure 7.10 (d)). The slight increase in the measured strength with increasing thickness might be attributed to measurement error.

Figure 7.9 (e) compares the predicted and measured yield strength (YS) and ultimate tensile strength (UTS) for three strips of different steel grades coiled at 600 to 615 °C, with details in Table 7.3 (c). The predicted YS increased from 297 to 328 MPa as the carbon and manganese contents are raised from 0.06C-0.18Mn to 0.08C-0.35Mn while the predicted UTS increased from 382 to 393 MPa, resulting in a one percent decrease in elongation (Figure 7.10 (e)). However, it is observed that the measured YS, UTS and elongation exhibited a more tangible change with increasing alloy content than the predicted values. The reason for this is not certain. The slight increase in strength for the 0.06C-0.18Mn over the 0.06C-0.3Mn might be due to the lower thickness of the former. The measured elongation followed the expected trend, decreasing with increasing alloying content.
7.2 Model Results and Discussion

7.2.1 Heat Transfer and Austenite Grain Size Evolution During Rolling

Employing the CSP rolling schedules listed in Table 7.3, the predicted thermal profiles and austenite grain size evolution (surface and center) for the rolling of a 5 mm thick strip of AISI 1018 is shown in Figure 7.11. This prediction was made assuming uniform through-thickness strain and strain rate. The 50 mm thick slab exited the tunnel furnace at a uniform temperature of 1100 °C and an as-cast austenite grain size of 1000 μm. It entered the descaler at a speed of 0.367 m/s where the surface temperature dropped rapidly to 659 °C due to spray cooling (the temperature rebounded after descaling) while the center temperature decreased by only 5 °C. In the roll bites (F1 to F6), the strip surface temperature decreased to below 600 °C as a result of roll chilling while the center temperature increased by 5 to 17 °C due to heat generation from plastic deformation. At the mill exit (5 m from F6), the model predicted strip surface and center temperatures were 875 and 879 °C respectively, compared to the 870±15 °C target.

With respect to recrystallization, grain refinement was dominated by metadynamic recrystallization in the first three stands (except at the surface where static recrystallization occurred due to low temperatures) and by static recrystallization in the last three stands. In the first interstand, the coarse grain size recrystallized completely at the strip center within the first one metre from F1 exit while complete static recrystallization occurred at the strip surface just before F2 (≈ 5 m from F1) due to the relatively low temperature as the strip surface temperature rebounded from 544 to 992 °C. The initial 1000 μm grain size recrystallized to 107 and 132 μm at the surface and center respectively, the grain size at the center increased by 4 μm as a result of grain growth before F2 entrance. Grain refinement by recrystallization continued after each pass, leading to the final grain size of 24 (107→49→34→26→24 μm) and 28 μm.
(136→74→50→33→28 μm) at the strip surface and center respectively. The interstand grain growth ranged from 1 to 9 μm, the maximum grain growth of 9 μm occurred at the strip center between F2 and F3 where the interstand time was 5.6 seconds at relatively high temperatures of 1066 to 1026 °C. It is noted that there was no change in austenite grain size after F5 until F6 entrance because of the slow recrystallization rate at relatively low strain (0.324) and temperatures. Furthermore the low strain at F6 (0.162) coupled with the low temperatures (< 900 °C) did not lead to any tangible recrystallization at the mill exit.

The predicted thermal profiles and austenite grain size evolution at the centerline for the three AISI 1018 steel strips of different thickness with rolling schedules listed in Table 7.4 are shown in Figure 7.12. It is observed that the mill exit target temperature of 870±15 °C is easily attained by a careful combination of mill entry speed and temperature. Similar results were obtained for the AISI 1005 steel grade. In general, the entry speed and/or temperature should increase with decreasing final thickness if the same target exit temperature is desired. The increase in entry speed and/or temperature compensates for the higher heat removal from thinner strips. An austenite grain size of 23 to 33 μm was predicted at the mill exit for the strip centerline, the grain size becoming finer as strip thickness decreases. Grain refinement in the three coils was dominated by metadynamic recrystallization in the first three stands and by static recrystallization in the last three stands. It is noted that with the same reduction in the first pass for these coils, the recrystallization rate increased with higher mill entry temperature.

A comparison of the thermal profiles for a 2.62 mm CCR and 2.87 mm CSP strips is presented in Figure 7.13, based on the rolling schedules listed in Table 7.5. The CSP schedule in this case was deduced from the data reported by Flemming et al.49. The entry temperature into the CSP mill was assumed constant through the thickness at 1130 °C compared to about 100 °C
difference between the surface and center of the CCR coil. It is evident that the predicted mill exit temperature is comparable to the measured value for the CCR and the reported range for the CSP. The major difference between the thermal histories of CSP and CCR rolling lies in the entry temperature, roll-bite heat-transfer coefficient and heat of deformation. The heat transfer coefficient and heat of deformation are higher in the CSP rolling than in the CCR due to the relatively large reductions per stand as listed in Table 7.6. The present simulation and an earlier one carried out for a 4-stand CSP mill show that the roll-bite heat-transfer coefficient for the CSP can be up to 25 pct. more than the equivalent CCR gauge while the temperature rise due to deformation can be 2 to 3 times higher. The high heat-transfer coefficient and the slower entry speed of the CSP process allows for enough heat extraction within the five stands that the mill exit temperature is comparable to conventional rolling (903 °C compared to 886 °C in this case).

The austenite grain size evolution at the strip center for two initial CSP grain sizes (500 and 1000 μm) and a CCR initial grain size of 180 μm, employing the uniform through-thickness strain assumption, is depicted in Figure 7.14. For the CSP 500 μm initial grain size, there is complete static recrystallization half-way through the first interstand compared to complete metadynamic recrystallization immediately after deformation in the CCR with an initial grain size of 180 μm. When the initial grain size in the CSP mill increased to 1000 μm, only partial static recrystallization ($X_{\text{re}} = 0.88$) occurs during the first interpass time. In both the CSP and CCR rolling, metadynamic recrystallization occurs in subsequent interstands until the fourth stand. Static recrystallization is completed after the last two stands in the CSP mill (F4 and F5) for both grain sizes compared to partial static recrystallization after the last two stands in the CCR mill (F6 and F7). Although partial recrystallization leads to a mixed grain size, only the predominant grain size (more than 70 percent of the total volume) is depicted in Figure 7.14.
The predicted recrystallized grain sizes in the first interstand were 77, 80 and 94 μm for the initial austenite grain sizes of 180, 500 and 1000 μm respectively. It is noted that the recrystallized grain size for the 1000 μm initial CSP grain size is finer in this case than the values reported in Figure 7.12 (94 μm compared to 136-140 μm). This is due to strain-induced grain refinement as result of the higher reduction in the first pass (50 pct. compared to 32 pct.). Grain refinement by recrystallization continued at the remaining interstands, leading to the final grain size of 24.7 μm in the CCR case (76.9→52.4→40.7→33.8→30.2→27.8→24.7 μm). The final austenite grain size resulting from CSP rolling is about 6 μm smaller than that of CCR rolling (for both 500 and 1000 μm initial grain sizes) despite the initial retardation of recrystallization from the coarse, as-cast grain size. The finer grain size in the CSP process results primarily from the high strains in the final stands (ε≥0.6 for F4 and F5 compared to ε≤0.3 for the CCR, F5, F6 and F7), which gives rise to complete recrystallization that consequently refines the microstructure at lower temperatures. The predicted austenite grain size of 24.7 μm for the CCR process matches the mean grain size of about 25 μm measured for A36 coils of various gauges\textsuperscript{135}.

7.2.2 Deformation During Rolling

7.2.2.1 Through-Thickness Strain and Strain Rate

The strain and strain rate contours in the last stand during CSP rolling of 2 and 8 mm coils predicted with the finite-element model, which allows for inhomogeneous deformation, are shown in Figures 7.15 and 7.16 respectively. It is seen that both the strain and strain rate are distributed non-uniformly through the strip thickness and from roll-gap entry to exit, both attaining their highest values close to the strip surface. In the case of strain rate, two peaks are observed near the surface, at the roll-gap entrance and exit, owing to the high redundant shear associated with constrained metal flow in and out of the roll bite. The through-thickness strain
ranged from 0.1 to 0.5 at roll-gap entry and 0.3 to 1.4 at the roll-gap exit for the 2 mm coil while the through-thickness strain rate changed from 150 to 2400 s\(^{-1}\) at the entry to the exit values of 150 to 854 s\(^{-1}\). For the 8 mm coil, the through-thickness strain ranged from 0.02 to 0.18 at roll-gap entry and 0.16 to 0.35 at the roll-gap exit while the through-thickness strain rate varied from 10 to 130 s\(^{-1}\) at the entry to the exit values of 10 to 70 s\(^{-1}\).

The through-thickness strain profiles at the exit of each roll-bite predicted with the Eulerian FEM model for both coils are shown in Figure 7.17. It is evident that for both cases, the strain is non-uniform through the thickness, increasing from close to the nominal value at the strip center to a maximum very near to the roll/strip interface. For example, the through-thickness strain at the exit of F4 varied from the 0.9 to a sub-surface maximum of 7.58 in the 2 mm coil compared to the nominal value of 0.81. In the 8 mm coil, the through-thickness strain at the exit of F4 varied from the 0.45 to a sub-surface maximum of 1.29 compared to the nominal value of 0.41. The surface region where the strain is substantially higher than the nominal value was found to constitute about 20 to 25 percent of the strip thickness. The through-thickness strain rate just before the roll-bite exit follows the same trend as the strain. Three major factors control the through-thickness strain inhomogeneity - reduction, strip speed and temperature gradient. The through-thickness strain inhomogeneity increases with increasing reduction, strip speed and temperature gradient from surface to center. The higher surface-to-center strain observed for the 2 mm gauge when compared to the 8 mm gauge arises mainly from the higher reductions and strip speed since the temperature gradient is less for thinner strips. It is noted that while the trend of strain distribution shown in Figure 7.17 has been confirmed experimentally\(^ {136}\), the absolute values of surface and subsurface strains are suspect since they change with slight variations in the roll/strip interface friction condition.
The through-thickness strain inhomogeneity has implications on the heat transfer through its contribution to heat generation from plastic deformation. Figure 7.18 depicts the computed temperature increment from plastic deformation at each stand for the 2 and 8 mm CSP coils. The temperature increment from plastic deformation ranged from 6 to 25 °C at the strip center for the 2 mm coil, compared to 20 to 175 °C at the strip surface. For the 8 mm coil, the temperature increment from plastic deformation varied from 3 to 11 °C at the strip center, compared to 14 to 43 °C at the strip surface. It is evident that these temperature increments follow the same trend as the through-thickness strain shown in Figure 7.17, increasing gradually from strip center and then more rapidly to a maximum near the surface. The surface region where the temperature increment was substantially higher than the centerline value was also found to constitute about 20 to 25 percent of the strip thickness.

The temperature increments computed at strip surface and center with the finite-element model are compared with the results of the uniform strain assumption in Figure 7.19. It is observed that for both coils, the uniform strain assumption largely underpredicts the heat generation at the surface, the largest difference between the finite-element and uniform strain predictions is 15 °C for the 8 mm gauge and 100 °C for the 2 mm gauge. Therefore, a model based on uniform strain assumption might not accurately predict the thermal profiles, particularly for thin coils where the contribution of deformation heat could be significant as a result of high reductions, as illustrated in Figures 7.2 and 7.4 for the 1.524 and 2.54 mm strips respectively.

The effect of through-thickness strain and strain rate variation on recrystallization and austenite grain size evolution for the two coils is shown in Figures 7.20 and 7.21 respectively. The contours in Figure 7.20 represent the through-thickness recrystallized fraction between F5 and F6 computed with the strains shown in Figure 7.17. It is observed that recrystallization is faster at the surface than at the center for the two coils because of the high surface strains and
strain rates, a situation that differs from the predictions made with the uniform-strain assumption where the center always recrystallized faster than the surface due to the lower surface temperatures. Also, recrystallization is completed throughout the thickness in a fraction of a second for the 2 mm coil while a good fraction of the 8 mm coil (near the center) did not fully recrystallize after the 2.5 s interstand time, reflecting the effect of the higher reduction utilized for the 2 mm coil compared to that for the 8 mm coil. Higher reduction implies higher strain and strain rate coupled with an elevated heat of deformation. With respect to final grain size at the mill exit, Figure 7.21 clearly shows that the through-thickness strain and strain rate variations result in inhomogeneous grain size distribution through the thickness, with the finer grains located near the surface. The predicted grain size was 23 μm at the center and 10 μm at the surface for the 2 mm coil, compared to values of 28 μm and 20 μm respectively for the 8 mm coil. The uniform strain assumption appears to provide a good estimate of grain size near the strip center for both coils. The greater disparity between the surface grain sizes predicted by uniform and non-uniform strain for the thinner gauge can be attributed to the higher deformation inhomogeneity introduced by the large reductions per pass in the thin gauge coil. This result agrees well with the trend reported in the literature on the effect of inhomogeneous deformation on microstructure.\textsuperscript{135}

The through-thickness strain profiles at the exit of the roll-bite predicted with the Eulerian FEM model for both CSP and CCR rolling are compared in Figure 7.22 utilizing the rolling schedules of Table 7.4. The nominal strains exiting the first and last stand of the CSP mill in this particular case are 0.81 and 0.61 respectively, compared to 0.69 and 0.16 for CCR rolling. Figure 7.22 (a) shows that the large reductions employed in CSP rolling create through-thickness strains greater than unity for most of the strip thickness which occurs to a lesser extent in the CCR case (Figure 7.22 (b)). The through-thickness strain rate at the roll-bite exit follows the same trend as
the strain, increasing from the nominal range of 4 to 95 s\(^{-1}\) at the centerline to about 3-4 times this value near the surface for CSP rolling. The strain rates at the exit of the first three stands of the CCR mill are approximately three times the CSP values as a result of the slower slab entry speed into the mill.

The effect of through-thickness strain and strain rate variation on austenite grain size evolution for the two processes is shown in Figure 7.23. It is seen that in both cases, through-thickness strain and strain rate variations result in inhomogeneous grain size distribution through the thickness, with the finer grains located near the surface. Uniform strain and strain rate assumptions appear to provide a good estimate of grain size near the strip center for the CCR process but they are not as satisfactory for the CSP process. The greater disparity between the grain sizes predicted by uniform and non-uniform strain for the CSP case can be attributed to the higher deformation inhomogeneity introduced by the large reductions per pass.

### 7.2.2.2 Roll Force Prediction

Values of the roll force per unit width, predicted by the finite-element model for three thicknesses of AISI 1018 steel grade, are compared with mill target estimates in Figure 7.24 (a); the vertical bar represents the range of rolling loads measured on the mill corresponding to conditions for which the targets were estimated. A detailed analysis of roll force measurement revealed that a maximum of 5 to 10 % deviation from the mean measured value for a given coil is common, although the standard deviation of measurement is only about 2 % of the mean value at each roll stand. Taking this into account, it is seen that the model predictions agree reasonably well with measurements for the majority of the roll stands. The observed discrepancies at some stands could be due to uncertainties in the temperature and friction conditions. It is seen that the
mill load increases from the first stand to the third, and then decreases to the lowest load at F6, reflecting the changes in reduction, temperature, strain rate and grain size.

The roll force prediction is sensitive to the accuracy of the constitutive equation employed. A constitutive equation which does not take grain size variation into account will overpredict the deformation resistance from coarse grains and underpredict the deformation resistance from fine grains. Calculations show that the first pass of the CSP mill (1000 μm grain size) requires about 25 percent less force than the equivalent conventional hot strip mill (200 μm) under the same reduction, temperature and strain rate conditions. Figure 7.24 (b) compares the predicted and measured roll forces for CSP and CCR mills utilizing the rolling schedule of Table 7.5; the 4-stand CSP measurements were reported for an unspecified plain carbon steel grade in the literature⁴⁹. It is seen that the model predictions agree reasonably well with measurements for the majority of the roll stands.

The first stand of the CSP mill is essentially its only roughing operation, its main function being refinement of the as-cast grains. The as-cast austenite reduces the flow stress in this first pass, which subsequently requires a lower roll force for the same reduction. The metallurgical changes triggered in the first pass also have an implication for the deformation behavior in the second stand. High reduction and lower strain rate in the first pass enhance complete metadynamic recrystallization while the coarse, as-cast austenite retards recrystallization. When the effect of the coarse, as-cast grain size dominates, incomplete recrystallization in the interpass time increases the flow stress during the next rolling pass as a result of work hardening from retained strain. Beynon and Sellars⁶⁴ have shown that the roll force per unit width in the second pass of a roughing mill can increase by about 12 percent when the austenite grain size changes from 100 to 1400 μm due to work hardening from incomplete recrystallization in the previous pass. An indication that this might be happening in the CSP mill is provided by the roll force
measurements for the 4-stand CSP mill shown in Figure 7.24 (b). The roll force measured in the second pass (54.5 pct. reduction) was about 10 percent higher than the roll force in the first and third passes (53.5 and 54.8 pct. reduction respectively). Simulation of the rolling conditions with coarse grain size greater than 1200 μm indicated that partial static recrystallization occurred after the first pass, resulting in strain accumulation in the second pass.

7.2.3 Effect of Rolling Parameters

7.2.3.1 Effect of Coarse Grain Size

For the operating conditions at the first interstand of the CSP mill listed in Table 7.7, calculations show that complete recrystallization is achieved throughout the strip thickness for interstand times greater than 10 s. Partial recrystallization prevails at the surface (about 10 to 20 percent of the strip thickness) for interstand times less than 10 s as shown in Figure 7.25. For example, when the interstand time is 7 s, the surface of a strip with 0.6 mm as-cast austenite grain will achieve only 19 pct. recrystallization (15 pct. for 1.4 mm as-cast grain). Provided that recrystallization is completed in the first interstand, the recrystallization kinetics in subsequent CSP passes are not different from that of a conventional hot strip mill. There is complete metadynamic recrystallization at higher temperatures and higher reductions, and complete static recrystallization at lower temperatures and lower reductions in the last passes. In a typical hot strip mill, recrystallization in the last two passes is usually driven by strain rate since the strain and temperature are relatively low. Therefore, for lower exit temperatures (< 900 °C), there may be incomplete recrystallization after the last two passes, particularly for thick gauges with low strains and strain rates. In the CSP mill, the chance of partial recrystallization in the last two stands is lower due to the relatively higher strain values (≥ 10 pct. reduction compared to, as low as, 5 pct. in the last stand of a conventional hot strip mill). The predicted final austenite grain
size for CSP rolling lies in the range of 18 to 35 \( \mu m \) (ASTM 9-7.5) for AISI 1018 steel and 25 to 40 \( \mu m \) (ASTM 8-6.7) for AISI 1005 steel.

The effect of a coarse grain size on recrystallization kinetics, similar to the results obtained in the first interstand of the CSP mill, has been reported\(^ {26,64,84} \). A simulation conducted by Beynon and Sellars\(^ {64} \) showed that, whereas full static recrystallization occurred when the reheated austenite grain size is 100 \( \mu m \), coarser grain sizes of 400 and 1400 \( \mu m \) resulted in 90 and 21 percent recrystallization respectively for the same reduction and temperature. An average of 55% increase in \( t_{0.5} \) has been reported\(^ {26} \) for directly rolled Si-Mn steels when the austenite grain size increased from 100 to 900 \( \mu m \) within a deformation temperature range of 900 - 1200 °C. Priestner et al.\(^ {84} \) reported that while a reheated C-Mn steel specimen (\( \approx 150 \mu m \)) with 30 percent reduction attained complete recrystallization in one second at 1000 °C, the as-cast direct-rolled specimen of the same steel (\( \approx 850 \mu m \)) required a temperature of 1170 °C to recrystallize completely with the same amount of time and reduction.

The fact that the CSP mill operates in a wide range of austenite grain size (\( \approx 1000 \) to \( 20 \mu m \)) is envisaged to have consequences for heat transfer. The variation of deformation resistance with grain size will lead to differences in the heat of deformation. To explore the effect of this grain size variation on heat generation during CSP rolling, the finite-difference model was run for two conditions: (a) employing the earlier constitutive equation for A36 steel that does not account for grain size, and (b) employing the new constitutive equation that accounts for grain size. For a 5 mm final gauge of A36 steel, a 15 °C difference exists between the predicted F6-exit surface temperature for the two conditions as seen in Figure 7.26. In general, a constitutive equation which does not take grain size variation into account, will overpredict the heat generation from coarse grains and underpredict the heat generation from fine grains. Calculations reveal that the
first pass of the CSP mill (1000 μm grain size) generates 25 percent less heat than the equivalent conventional hot strip mill (200 μm) under the same reduction, temperature and strain rate conditions.

### 7.2.3.2 Effect of Reduction Schedule and Number of Roll Stands

As observed in Chapter 1, the number of roll stands in the CSP mill has evolved from four to five and now six stands, with one of the new mills now implementing a seven stand operation. The reduction per stand, particularly in the first and last stands has been changed considerably in an effort to reduce edge cracking, roll wear/distortion and shape problems. This trend in CSP mill evolution is illustrated in Table 7.8, noting that improper choice of the reduction schedule and the number of roll stands could lead to undesirable temperature gradients and inhomogeneous deformation, and unacceptable variations in strip width, gauge, microstructure and shape.

Table 7.8 shows that the four stand mill utilized over 50 pct. reduction in the first three stands and close to 40 pct. in the last stand for the production of thin strips (<3mm in thickness). Edge cracking in the first stand for some sensitive grades, shape and flatness problems as well as roll wear/distortion were commonly encountered. This problem continued in the five stand design where the reduction in the first and last stands remained high (schedule II in the Table 7.8) although roll wear/distortion was marginally reduced in the intermediate stands with lower reductions. Further decrease of the percent reduction in the first and last stands (schedule III in Table 7.8) was introduced to improve shape control and reduce the incidence of edge cracking. However, the high reductions at F2 and F3 imply faster rate of roll wear and distortion in the two stands. The introduction of six stand rolling (schedule IV in Table 7.8) allowed further reduction in draft at the first and last passes, and has the potential to eliminate edge cracking and ensure the desired shape and flatness. A seventh stand will obviously improve quality for very
sensitive grades by offering greater flexibility in the distribution of drafts comparable to the equivalent CCR gauge.

It is noted that roll life is closely tied to the reduction schedule and the thermal cycle, and each of the schedules in Table 7.8 demands a different optimal selection of rolls. It has been suggested that the most economic combination of rolls in the CSP mill will involve the use of chrome iron in the early stands and indefinite chill rolls at the last stands\textsuperscript{137}. However, the more expensive chrome steel has been found to perform better than chrome iron under conditions of low speeds, high loads and high heat experienced in the early stands\textsuperscript{137}. Chrome iron rolls are best known for their ability to maintain good strip surface quality and strip shape as well as wear resistance. Indefinite chill rolls are preferred where wear resistance in conjunction with high speed performance is needed.

Calculations were carried out to compare a five and six stand CSP mills in terms of heat transfer, deformation and microstructure for a given final gauge. It was found that the target exit temperature is easily achieved and the final austenite grain size is comparable for both designs. There is a slight increase in the rate of recrystallization in the five stand than the six stand mill due to relatively higher strain at each stand, which also led to higher deformation inhomogeneity. With respect to reduction schedule, simulation results for a 1000 \( \mu \)m initial grain size are compared in Table 7.9, for schedules II and III in Table 7.8. The table indicates that partial static recrystallization occurs at the end of the first interpass time \( X_{\text{rex}} = 0.65 \) and after the last pass in III compared to \( X_{\text{rex}} = 0.88 \) and full recrystallization in II after the first and last passes respectively. The final austenite grain size in II is about 4 \( \mu \)m finer than in III since the grain refinement from the last two stands in III is minimal as a result of the relatively low strains (0.4 for F4 and 0.2 for F5). A wider distribution of grain sizes is also developed in the schedule III strip due to the partial recrystallization after the first and last stands. To go from schedule II to
III, a decrease in the slab entry speed is a step in the right direction, since it ensures a low strain rate which stimulates the relatively faster dynamic and metadynamic recrystallization.

7.2.3.3 Effect of Slab Entry Speed and Temperature

The first CSP strips were rolled with a fixed slab entry speed of 0.33 m/s and tunnel furnace exit temperature of 1100 °C, for every grade and thickness. In order to meet desired target exit temperatures, simulation results indicate that the entry speed and/or temperature should increase with decreasing final thickness if the same target exit temperature is desired. The increase in entry speed and/or temperature compensates for the higher heat removal from thinner strips. Hence, entry temperatures of 1080, 1100 and 1150 °C were utilized for the 8, 5 and 2 mm coils respectively in Figure 7.12. Of course, there is a limit to how high the entry speed can get, based on the speed capacity and the process control stipulations of the runout table cooling and the coiler unit. It is known that the rate of cobble increases with decreasing strip thickness and increasing speeds\textsuperscript{132}. The tunnel furnace capacity limits the extent of entry temperature.

7.2.3.4 Scale Formation and Descaling Practice

The descaling practice listed in Table 7.1 and utilized in the previous calculations (except for HYLSA’s mill setup) was recently redesigned in most CSP mills to ensure minimal slab cooling while maintaining efficient scale removal, following incidences of rough surface, dimensional distortion and rolled-in scale attributed to inadequate descaling\textsuperscript{30,48}. The new design utilizes two spray headers, 1.2 m apart, equipped with small nozzles and operated at high pressures (up to 450 bar) with reduced water flow rate. Calculations were carried out to compare this new design with the older one for the 5 mm strip, and the results are presented in Figure 7.27. In this case the combined water flow rate from the two headers was about 25 percent of that of the single header. Figure 7.26 (a) shows that the high flow rate single header spray extracted about 50 °C
from strip surface with the temperature dropping to as low as 653 °C from 1050 °C. On the other hand, the low flow rate double header design extracted about 30 °C from the strip surface with the temperature dropping to 864 °C from 1050 °C. In both cases, the centerline temperature decreased by only 10 °C in the descaling zone. However, the difference between the strip surface and centerline temperatures for both designs at the F6 exit is only 5 °C. These calculations confirm that the low flow rate double header design satisfies the requirement of minimal slab cooling while contributing to a lower incidence of multiple scale-affected zones as reported by mill operators

Calculations were also carried out to evaluate the effect of primary and secondary scale layer on the strip thermal profile. A 1.0 mm primary scale thickness from the tunnel furnace was assumed, and secondary scale formation was computed with the parabolic scale growth equation as outlined by Chen et al.138. It was also assumed that the scale layer entering each stand undergoes exactly the same draft as the strip during deformation. The effect of scale formation on the predicted thermal profile of the 8 mm AISI 1018 strip is shown in Figure 7.28. It is seen that the predicted mill exit temperature increased by 35 °C when scale formation is included in the calculations. The secondary scale thickness varied from 28 μm at the entrance to F1, to 8 μm at F6. The insulation effect of the scale was found to be tangible from the exit of the tunnel furnace to the entrance into stand F3. It is seen that the prediction without scale provided a better match with the target temperature than the calculation with scale. This suggests that thermal insulation from scale was not tangible in this case, a situation that could arise when there is water on the strip surface from F1 to F3 or when the scale layer is porous enough that a good fraction of the strip surface is cooled by water emanating from roll cooling in conjunction with radiation/air cooling.
7.2.3.5 Effect of Interstand Cooling

Interstand cooling after the first two passes has been successfully utilized to reduce interstand scale growth in CSP rolling. The interstand cooling ensures that a significant fraction of the strip surface between two stands is covered by water, thereby reducing the surface temperature and direct steel/air contact. Calculations were carried out to investigate the effect of introducing two interstand sprays between F1 and F2, and between F2 and F3. The results of the strip temperature and austenite grain size evolution for the 5 mm AISI 1018 strip with interstand water flow rate of 4.5 l/s are presented in Figure 7.29. It is observed that the mill exit temperature decreases by 50 °C with the introduction of interstand sprays, leading to 5 μm finer austenite grain size. The reduction in temperature by the interstand sprays retards recrystallization; the surface of strip (20 pct. of the thickness) cooled with interstand sprays was partially recrystallized at F2 entry (9.5 seconds interstand time) compared to full recrystallization within 3 seconds without interstand spray. For a given target exit temperature, it is necessary to increase the rolling speed and/or the slab entry temperature when interstand sprays are utilized. For example, the calculated temperature drop in this case can easily be offset by increasing the slab entry speed by 25 percent. Raising the slab temperature at the tunnel furnace exit from 1100 to 1150 °C will also reduce the predicted temperature drop by more than half. Hence, interstand cooling is often employed as a tool for strip temperature control or to boost productivity via higher speed rolling.

7.2.4 Runout Table Cooling, Ferrite Grain Size Evolution and Mechanical Properties

Table 7.10 lists the predicted ferrite grain size and mechanical properties of a 5 mm strip of AISI 1018 and AISI 1005 steel grades for different coiling temperatures and cooling patterns. The amounts of ferrite, pearlite, non-polygonal structure and nitrogen in solution at the end of the cooling zone are also listed. Three coiling temperatures and the two most practiced cooling
patterns ((a) intense cooling from the early banks until the coiling temperature is reached and (b) moderate cooling at the beginning in conjunction with trim banks at the end) were chosen for comparison. There is a slight difference in the strip speed and austenite grain size entering the runout table for the two steels; strip speed and austenite grain size were 3.67 m/s and 25 μm for the AISI 1018 strip compared to 4.33 m/s and 32 μm for the AISI 1005 strip. The associated thermal profiles and cooling rates for the results listed in Table 7.10 are presented in Figures 7.30 through 7.33.

It is seen that a lower coiling temperature increases the yield strength and UTS while decreasing the percent elongation for both steels. The total nitrogen in solution also rises with decreasing coiling temperature. The tendency towards pearlite formation and non-polygonality in the AISI 1018 grade is seen to be higher at lower coiling temperatures. It is noted that the predicted ferrite grain size appears not to change with either the coiling temperature (except for the AISI 1018 coiled at 700 °C) or the cooling pattern. The reason for this lies in the assumption that the ferrite grain size in the final microstructure is essentially determined at the start of transformation. Hence, ferrite grain size remains the same if the conditions (temperature, cooling rate and austenite grain size) at the start of transformation are similar, as was the case in these calculations. For a given coiling temperature, the use of trim banks appear to decrease the tendency towards non-polygonality in the AISI 1018 grade while slightly increasing percent elongation at the expense of strength (YS and UTS decreases).

For the same coiling temperature, the calculations indicate that the AISI 1005 strip required 20 to 50 percent more active jetlines at the runout table than the AISI 1018 strip. This is a manifestation of the higher transformation start temperature of the AISI 1005 steel grade, the latent heat released during transformation adds to the amount of heat that must be removed by the
cooling water. The equilibrium $\gamma \rightarrow \alpha$ transformation start temperature ($T_{Ae3}$), assuming no undercooling, is 883 °C for AISI 1005 compared to 824 °C for AISI 1018 grade. This difference in transformation start temperatures between the two steels has been shown\textsuperscript{128} to remain approximately the same in the cooling rate regime of 0 to 80 °C/s, a range that covers the average cooling rate calculated at the commencement of transformation for the two steels as seen in Figures 7.30 to 7.33. It is observed that transformation is completed within the active cooling zone for the AISI 1005 strip at the three coiling temperatures when cooling is performed without trimming. This is not the case for the AISI 1018 grade where $\gamma \rightarrow \alpha$ transformation ($T_c = 700$ °C) and $\gamma \rightarrow$ pearlite transformation ($T_c = 625$ or 550 °C) continue after the active cooling zone. In fact, $\gamma \rightarrow \alpha$ transformation is not completed in the AISI 1018 steel grade when the target coiling temperature is 700 °C.

With trim bank cooling, transformation barely starts at the first stage of cooling in the AISI 1018 strip compared to the almost completed transformation in the AISI 1005 strip. Hence, trim bank cooling does not contribute significantly to transformation in the AISI 1005 strip, a situation quite different in the AISI 1018 strip where it substantially increases the rate of transformation. It is noted that the heat extraction from the strip in the ferrite regime, although at lower temperature, could be higher than in the austenite regime due to the higher heat diffusivity of ferrite as manifested in the cooling rates of Figures 7.32 and 7.33 where the ferrite fraction is close to unity.
Table 7.1 Typical CSP Mill Dimensions/Specifications (SMS)

(a) Mill Stands

<table>
<thead>
<tr>
<th>Mill Sub-unit</th>
<th>Dimension (m)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tunnel furnace exit to descaler</td>
<td>7.25</td>
</tr>
<tr>
<td>Descaler coverage length</td>
<td>0.10</td>
</tr>
<tr>
<td>Descaler to F1</td>
<td>2.7</td>
</tr>
<tr>
<td>F1 to F2</td>
<td>5.14</td>
</tr>
<tr>
<td>F2 to F3</td>
<td>5.14</td>
</tr>
<tr>
<td>F3 to F4</td>
<td>5.14</td>
</tr>
<tr>
<td>F4 to F5</td>
<td>5.14</td>
</tr>
<tr>
<td>F5 to F6 (or F5 to exit Pyrometer in the case of a 5-stand mill)</td>
<td>5.14</td>
</tr>
<tr>
<td>F6 to exit pyrometer</td>
<td>5.0</td>
</tr>
</tbody>
</table>

(b). Descaling
Water Flow Rate (litre/min): 70.97
Coverage Width (m): Maximum strip width + 100 mm

(c). Roll Cooling
Roll Type: High Chromium Iron (F1-F4), Indefinite Chill Iron (F5 and F6)
Roll Diameter (mm): 780 mm
Number of Spray Headers: 2 on entry side and 3 on exit side
Spray Coverage on Rolls (angle in degree): 25-40
Water Flow Rate per Spray Header (litre/min): 503-1700; decreases from F1 to F6
Coverage Width (m): Maximum strip width + 100 mm

(d). Run-out Table

<table>
<thead>
<tr>
<th>Run-out Table Configuration</th>
<th>Top</th>
<th>Bottom</th>
</tr>
</thead>
<tbody>
<tr>
<td>Total number of spray banks</td>
<td>7</td>
<td>7</td>
</tr>
<tr>
<td>Number of jet lines per bank</td>
<td>8</td>
<td>16</td>
</tr>
<tr>
<td>Entry Length before 1st jet (m)</td>
<td>10</td>
<td>10</td>
</tr>
<tr>
<td>Number of nozzles per header</td>
<td>60</td>
<td>30</td>
</tr>
<tr>
<td>Header length (m)</td>
<td>1.6</td>
<td>1.6</td>
</tr>
<tr>
<td>Water flow rate per header (l/s)</td>
<td>19.44</td>
<td>5.50</td>
</tr>
<tr>
<td>Distance between banks (m)</td>
<td>0.90</td>
<td>0.60</td>
</tr>
<tr>
<td>Distance between jet lines (m)</td>
<td>0.70</td>
<td>0.33</td>
</tr>
<tr>
<td>Jet width (diameter for bar) (m)</td>
<td>0.02</td>
<td>0.01</td>
</tr>
<tr>
<td>Vertical distance from nozzle to strip (m)</td>
<td>1.6</td>
<td>-0.13</td>
</tr>
<tr>
<td>Angle of jets (from vertical, degrees)</td>
<td>0</td>
<td>5</td>
</tr>
<tr>
<td>Distance between table rollers (m)</td>
<td>0.40</td>
<td></td>
</tr>
<tr>
<td>Distance from last jet to the coiler (m)</td>
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<td>21.89</td>
</tr>
<tr>
<td>Spray bank length (m)</td>
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<td>5</td>
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<tr>
<td>Spray zone length (m)</td>
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<td>40</td>
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<tr>
<td>Run-out table length (m)</td>
<td>70</td>
<td>70</td>
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</table>
## Table 7.2 Comparison of measured and predicted temperatures for ten CSP coils.

<table>
<thead>
<tr>
<th>Coil ID</th>
<th>Grade</th>
<th>Thickness (mm)</th>
<th>Measured F6 Exit Temp. (°C)</th>
<th>Predicted F6 Exit Temp. (°C)</th>
<th>Measured Coiling Temp. (°C)</th>
<th>Predicted Coiling Temp. (°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>141120</td>
<td>2061</td>
<td>2.69</td>
<td>897</td>
<td>888</td>
<td>615</td>
<td>609</td>
</tr>
<tr>
<td>140804</td>
<td>2061</td>
<td>3</td>
<td>891</td>
<td>893</td>
<td>715</td>
<td>723</td>
</tr>
<tr>
<td>141063</td>
<td>7061</td>
<td>1.524</td>
<td>910</td>
<td>905</td>
<td>722</td>
<td>734</td>
</tr>
<tr>
<td>141337</td>
<td>7061</td>
<td>2.16</td>
<td>914</td>
<td>918</td>
<td>726</td>
<td>727</td>
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<td>140611</td>
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<td>2.54</td>
<td>880</td>
<td>885</td>
<td>615</td>
<td>613</td>
</tr>
<tr>
<td>140758</td>
<td>7061</td>
<td>3.15</td>
<td>900</td>
<td>890</td>
<td>722</td>
<td>725</td>
</tr>
<tr>
<td>140783</td>
<td>7061</td>
<td>3.15</td>
<td>920</td>
<td>908</td>
<td>560</td>
<td>557</td>
</tr>
<tr>
<td>141323</td>
<td>7092</td>
<td>1.57</td>
<td>883</td>
<td>880</td>
<td>665</td>
<td>666</td>
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<td>140905</td>
<td>7092</td>
<td>3.23</td>
<td>889</td>
<td>881</td>
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<td>4.762</td>
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<td>894</td>
<td>665</td>
<td>667</td>
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</table>

7061 = 0.06C-0.18Mn, 2061 = 0.06C-0.30Mn, 7092 = 0.08C-0.35Mn
Table 7.3  Comparison of measured and predicted ferrite grain size and mechanical properties for ten CSP coils.

(a) Effect of coiling temperature

<table>
<thead>
<tr>
<th>Thickness mm [Coil ID]</th>
<th>Measured F6 Exit Temp., °C [dₜ, μm]</th>
<th>Predicted Tₛ (°C)</th>
<th>Predicted Cooling Rate at Tₛ (°C/s)</th>
<th>Predicted Coiling Temp., °C [Nₛ, %]</th>
<th>Measured dₕ (μm)</th>
<th>Predicted dₕ (μm)</th>
<th>Measured YS (MPa)</th>
<th>Predicted YS (MPa)</th>
<th>Measured UTS (MPa)</th>
<th>Predicted UTS (MPa)</th>
<th>Measured % Elong.</th>
<th>Predicted % Elong.</th>
</tr>
</thead>
<tbody>
<tr>
<td>I. 0.06C-0.18Mn (7061) Steel Grade</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>3.15 [140783]</td>
<td>920 [31.7]</td>
<td>811.9</td>
<td>86.1</td>
<td>557 [83.2]</td>
<td>9</td>
<td>15.3</td>
<td>318</td>
<td>307</td>
<td>397</td>
<td>378</td>
<td>30.8</td>
<td>36.1</td>
</tr>
<tr>
<td>2.54 [140611]</td>
<td>880 [26.7]</td>
<td>798.8</td>
<td>227.8</td>
<td>613 [61.0]</td>
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<td>297</td>
<td>379</td>
<td>382</td>
<td>34.3</td>
<td>35.9</td>
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<tr>
<td>3.15 [140758]</td>
<td>900 [29.9]</td>
<td>813.1</td>
<td>82.6</td>
<td>725 [0.0]</td>
<td>16.6</td>
<td>15.1</td>
<td>299</td>
<td>287</td>
<td>350</td>
<td>360</td>
<td>35.7</td>
<td>37.0</td>
</tr>
<tr>
<td>II. 0.06C-0.3Mn (2061) Steel Grade</td>
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<td></td>
<td></td>
<td></td>
<td></td>
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<td></td>
<td></td>
</tr>
<tr>
<td>2.69 [141120]</td>
<td>897 [27.7]</td>
<td>809.4</td>
<td>154.2</td>
<td>609 [62.9]</td>
<td>12.0</td>
<td>13.9</td>
<td>309</td>
<td>326</td>
<td>377</td>
<td>387</td>
<td>32.3</td>
<td>35.6</td>
</tr>
<tr>
<td>3.0 [140804]</td>
<td>891 [28.8]</td>
<td>805.5</td>
<td>131.4</td>
<td>723 [0.0]</td>
<td>13.1</td>
<td>13.6</td>
<td>306</td>
<td>307</td>
<td>371</td>
<td>378</td>
<td>33.4</td>
<td>36.1</td>
</tr>
</tbody>
</table>

(b) Effect of strip thickness (0.06C-0.18Mn)

<table>
<thead>
<tr>
<th>Thickness mm [Coil ID]</th>
<th>Measured F6 Exit Temp., °C [dₜ, μm]</th>
<th>Predicted Tₛ (°C)</th>
<th>Predicted Cooling Rate at Tₛ (°C/s)</th>
<th>Predicted Coiling Temp., °C [Nₛ, %]</th>
<th>Measured dₕ (μm)</th>
<th>Predicted dₕ (μm)</th>
<th>Measured YS (MPa)</th>
<th>Predicted YS (MPa)</th>
<th>Measured UTS (MPa)</th>
<th>Predicted UTS (MPa)</th>
<th>Measured % Elong.</th>
<th>Predicted % Elong.</th>
</tr>
</thead>
<tbody>
<tr>
<td>I. 0.06C-0.18Mn (7061) Steel Grade</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>1.524 [141063]</td>
<td>910 [20.9]</td>
<td>794.2</td>
<td>424.7</td>
<td>734 [0.0]</td>
<td>12.6</td>
<td>10.1</td>
<td>354</td>
<td>298</td>
<td>395</td>
<td>378</td>
<td>30.0</td>
<td>36.1</td>
</tr>
<tr>
<td>2.16 [141337]</td>
<td>914 [28.41]</td>
<td>810.4</td>
<td>104.8</td>
<td>727 [0.0]</td>
<td>14.9</td>
<td>14.3</td>
<td>328</td>
<td>306</td>
<td>384</td>
<td>365</td>
<td>35.6</td>
<td>36.7</td>
</tr>
<tr>
<td>3.15 [140758]</td>
<td>900 [29.9]</td>
<td>813.1</td>
<td>82.6</td>
<td>725 [0.0]</td>
<td>16.6</td>
<td>15.1</td>
<td>299</td>
<td>287</td>
<td>350</td>
<td>360</td>
<td>35.7</td>
<td>37.0</td>
</tr>
<tr>
<td>II. 0.08C-0.35Mn (7092) Steel Grade</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

* measured in the small polygonal section of the microstructure, Tₛ = transformation start temperature, Nₛ = Nitrogen in solution
Table 7.3 Contd. Comparison of measured and predicted ferrite grain size and mechanical properties for ten CSP coils.

(c) Effect of steel composition

<table>
<thead>
<tr>
<th>Thickness mm</th>
<th>Measured F6 Exit Temp., °C</th>
<th>Predicted ( T_s ) (°C)</th>
<th>Predicted Cooling Rate at ( T_s ) (°C/s)</th>
<th>Predicted Coiling Temp., °C</th>
<th>Measured ( d_a ) (µm)</th>
<th>Predicted ( d_a ) (µm)</th>
<th>Measured ( \bar{Y}S ) (MPa)</th>
<th>Predicted ( \bar{Y}S ) (MPa)</th>
<th>Measured ( UTS ) (MPa)</th>
<th>Predicted ( UTS ) (MPa)</th>
<th>Measured % Elong.</th>
<th>Predicted % Elong.</th>
</tr>
</thead>
<tbody>
<tr>
<td>2.54 [140611]</td>
<td>880 [26.7]</td>
<td>798.8</td>
<td>227.8</td>
<td>613 [61.0]</td>
<td>12.7</td>
<td>11.9</td>
<td>316</td>
<td>297</td>
<td>379</td>
<td>382</td>
<td>34.3</td>
<td>35.9</td>
</tr>
<tr>
<td>2.69 [141120]</td>
<td>897 [27.7]</td>
<td>809.4</td>
<td>154.2</td>
<td>609 [62.9]</td>
<td>12.0</td>
<td>13.9</td>
<td>309</td>
<td>326</td>
<td>377</td>
<td>387</td>
<td>32.3</td>
<td>35.6</td>
</tr>
<tr>
<td>3.23 [140905]</td>
<td>889 [27.6]</td>
<td>813.9</td>
<td>65.1</td>
<td>602 [66.3]</td>
<td>14.3</td>
<td>14.8</td>
<td>357</td>
<td>329</td>
<td>413</td>
<td>393</td>
<td>28.4</td>
<td>35.1</td>
</tr>
</tbody>
</table>

140611 = 0.06C-0.18Mn, 141120 = 0.06C-0.30Mn, 140905 = 0.08C-0.35Mn
Table 7.4 Typical CSP rolling schedule for AISI 1018 steel grade.

<table>
<thead>
<tr>
<th>Coil No. 1</th>
<th>[Width=1.0 m, Entry Speed=0.28 m/s, Target Exit Temperature=870 °C]</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Entry</td>
</tr>
<tr>
<td>Gauge (mm)</td>
<td>50</td>
</tr>
<tr>
<td>% Reduction</td>
<td>32.00</td>
</tr>
<tr>
<td>Target Roll Force (MN)</td>
<td>16.43</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Coil No. 2</th>
<th>[Width=1.0 m, Entry Speed=0.367 m/s, Target Exit Temperature=870 °C]</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Entry</td>
</tr>
<tr>
<td>Gauge (mm)</td>
<td>50.00</td>
</tr>
<tr>
<td>% Reduction</td>
<td>32.00</td>
</tr>
<tr>
<td>Target Roll Force (MN)</td>
<td>16.11</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Coil No. 3</th>
<th>[Width=1.0 m, Entry Speed=0.39 m/s, Target Exit Temperature=870 °C]</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Entry</td>
</tr>
<tr>
<td>Gauge (mm)</td>
<td>50.00</td>
</tr>
<tr>
<td>% Reduction</td>
<td>32.00</td>
</tr>
<tr>
<td>Target Roll Force (MN)</td>
<td>16.06</td>
</tr>
</tbody>
</table>

Table 7.5 CSP and CCR rolling schedule for thin AISI 1018 steel strip.

<table>
<thead>
<tr>
<th>I. 2.87 mm CSP Coil</th>
<th>[Width=1.26 m, Entry Speed=0.33 m/s, Target Exit Temperature=903°C]</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Entry</td>
</tr>
<tr>
<td>Gauge (mm)</td>
<td>50</td>
</tr>
<tr>
<td>% Reduction</td>
<td>50.0</td>
</tr>
<tr>
<td>Measured Roll Force (MN)</td>
<td>N/A</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>II. 2.62 mm CCR Coil</th>
<th>[Width=0.87 m, Entry Speed=0.86 m/s, Target Exit Temperature=886 °C]</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Entry</td>
</tr>
<tr>
<td>Gauge (mm)</td>
<td>30.33</td>
</tr>
<tr>
<td>% Reduction</td>
<td>44.73</td>
</tr>
<tr>
<td>Measured Roll Force (MN)</td>
<td>14.75</td>
</tr>
</tbody>
</table>

Table 7.6 Predicted heat transfer coefficient and temperature increment from plastic deformation at each stand for the CSP and CCR rolling schedules listed in Table 7.5.

<table>
<thead>
<tr>
<th></th>
<th>F1</th>
<th>F2</th>
<th>F3</th>
<th>F4</th>
<th>F5</th>
<th>F6</th>
<th>F7</th>
</tr>
</thead>
<tbody>
<tr>
<td>2.87 mm CSP Coil; h (KW/m².K) [ΔT from deformation (°C)]</td>
<td>163.3 [15]</td>
<td>186.7 [18]</td>
<td>307.7 [21]</td>
<td>434.5 [23]</td>
<td>400.4 [27]</td>
<td>- [22]</td>
<td>- [22]</td>
</tr>
</tbody>
</table>

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Table 7.7  Typical conditions at the first interstand of CSP mill

<table>
<thead>
<tr>
<th>Interpass Time (s)</th>
<th>Nominal Strain Rate (1/s)</th>
<th>Nominal Strain Rate</th>
<th>Surface Temp. (°C)</th>
<th>Center Temp. (°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>7 - 11</td>
<td>0.4 - 0.9</td>
<td>2 - 7</td>
<td>900 - 1040</td>
<td>1100 - 1050</td>
</tr>
</tbody>
</table>

Table 7.8  Stages in the development of reduction schedule for thin gauge CSP strips (AISI 1018)

<table>
<thead>
<tr>
<th>Schedule</th>
<th>Percent Reduction</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Gauge (mm)</td>
</tr>
<tr>
<td>I</td>
<td>2.87</td>
</tr>
<tr>
<td>II</td>
<td>2.87</td>
</tr>
<tr>
<td>III</td>
<td>2.87</td>
</tr>
<tr>
<td>IV</td>
<td>2.0</td>
</tr>
<tr>
<td>CCR</td>
<td>2.62</td>
</tr>
</tbody>
</table>

Table 7.9  Comparison of the recrystallization kinetics of two CSP rolling schedules (II and III in Table 7.8)

<table>
<thead>
<tr>
<th>Schedule</th>
<th>Parameter</th>
<th>F1</th>
<th>F2</th>
<th>F3</th>
<th>F4</th>
<th>F5</th>
<th>Exit</th>
</tr>
</thead>
<tbody>
<tr>
<td>II</td>
<td>$\varepsilon$ (s$^{-1}$)</td>
<td>0.81</td>
<td>0.689</td>
<td>0.648</td>
<td>0.608</td>
<td>0.608</td>
<td>-</td>
</tr>
<tr>
<td></td>
<td>$\dot{\varepsilon}$ (s$^{-1}$)</td>
<td>4.13</td>
<td>9.96</td>
<td>22.34</td>
<td>49.03</td>
<td>93.4</td>
<td>-</td>
</tr>
<tr>
<td></td>
<td>$X_{\text{rex}}$</td>
<td>0.88</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>$d_r$ (µm)</td>
<td>94.12</td>
<td>55.5</td>
<td>39.2</td>
<td>25.8</td>
<td>19.2</td>
<td></td>
</tr>
<tr>
<td>III</td>
<td>$\varepsilon$</td>
<td>0.608</td>
<td>1.09</td>
<td>1.09</td>
<td>0.405</td>
<td>0.203</td>
<td>-</td>
</tr>
<tr>
<td></td>
<td>$\dot{\varepsilon}$ (s$^{-1}$)</td>
<td>3.15</td>
<td>12.12</td>
<td>47.81</td>
<td>98.78</td>
<td>75.7</td>
<td>-</td>
</tr>
<tr>
<td></td>
<td>$X_{\text{rex}}$</td>
<td>0.65</td>
<td>1</td>
<td>1</td>
<td>1</td>
<td>0.2</td>
<td></td>
</tr>
<tr>
<td></td>
<td>$d_r$ (µm)</td>
<td>1000</td>
<td>104.37</td>
<td>59.13</td>
<td>28.16</td>
<td>23.6</td>
<td>23.6</td>
</tr>
</tbody>
</table>

$X_{\text{rex}}$ = fraction recrystallized computed at the entrance to each stand
$d_r$ = austenite grain size (recrystallized + grain growth) at the entrance to each stand
Table 7.10 Predicted ferrite grain size and mechanical properties of a 5 mm coil.

<table>
<thead>
<tr>
<th>Grade</th>
<th>$T_{en}$ ($^\circ$C)</th>
<th>$T_c$ ($^\circ$C)</th>
<th>Trim Bank</th>
<th>Ferrite Fraction</th>
<th>$\bar{d}_f$ (μm)</th>
<th>Pearlite Fraction</th>
<th>% Non-Polygonal</th>
<th>% N in Solution</th>
<th>YS (MPa)</th>
<th>UTS (MPa)</th>
<th>% Elongation</th>
</tr>
</thead>
<tbody>
<tr>
<td>AISI 1018</td>
<td>870</td>
<td>700</td>
<td>No*</td>
<td>0.7</td>
<td>4.9</td>
<td>0.0</td>
<td>11</td>
<td>0.0</td>
<td>328</td>
<td>454</td>
<td>33</td>
</tr>
<tr>
<td>AISI 1018</td>
<td>870</td>
<td>700</td>
<td>Yes*</td>
<td>0.56</td>
<td>7.0</td>
<td>0.0</td>
<td>15</td>
<td>0.0</td>
<td>305</td>
<td>449</td>
<td>34</td>
</tr>
<tr>
<td>AISI 1018</td>
<td>870</td>
<td>625</td>
<td>No</td>
<td>0.67</td>
<td>4.7</td>
<td>0.33</td>
<td>30</td>
<td>59.3</td>
<td>372</td>
<td>527</td>
<td>31</td>
</tr>
<tr>
<td>AISI 1018</td>
<td>870</td>
<td>625</td>
<td>Yes</td>
<td>0.73</td>
<td>4.7</td>
<td>0.27</td>
<td>15</td>
<td>57.5</td>
<td>364</td>
<td>505</td>
<td>31</td>
</tr>
<tr>
<td>AISI 1018</td>
<td>870</td>
<td>550</td>
<td>No</td>
<td>0.62</td>
<td>4.7</td>
<td>0.38</td>
<td>30</td>
<td>94.1</td>
<td>399</td>
<td>544</td>
<td>28</td>
</tr>
<tr>
<td>AISI 1018</td>
<td>870</td>
<td>550</td>
<td>Yes</td>
<td>0.83</td>
<td>4.7</td>
<td>0.17</td>
<td>29</td>
<td>93.2</td>
<td>355</td>
<td>492</td>
<td>32</td>
</tr>
<tr>
<td>AISI 1005</td>
<td>900</td>
<td>700</td>
<td>No</td>
<td>1.0</td>
<td>9.5</td>
<td>0.0</td>
<td>0.0</td>
<td>0.0</td>
<td>271</td>
<td>338</td>
<td>38</td>
</tr>
<tr>
<td>AISI 1005</td>
<td>900</td>
<td>700</td>
<td>Yes</td>
<td>1.0</td>
<td>9.5</td>
<td>0.0</td>
<td>0.0</td>
<td>0.0</td>
<td>271</td>
<td>338</td>
<td>38</td>
</tr>
<tr>
<td>AISI 1005</td>
<td>900</td>
<td>625</td>
<td>No</td>
<td>1.0</td>
<td>9.5</td>
<td>0.0</td>
<td>0.0</td>
<td>58</td>
<td>284</td>
<td>349</td>
<td>38</td>
</tr>
<tr>
<td>AISI 1005</td>
<td>900</td>
<td>625</td>
<td>Yes</td>
<td>1.0</td>
<td>9.5</td>
<td>0.0</td>
<td>0.0</td>
<td>58.9</td>
<td>284</td>
<td>351</td>
<td>38</td>
</tr>
<tr>
<td>AISI 1005</td>
<td>900</td>
<td>550</td>
<td>No</td>
<td>1.0</td>
<td>9.5</td>
<td>0.0</td>
<td>0.0</td>
<td>93</td>
<td>292</td>
<td>358</td>
<td>37</td>
</tr>
<tr>
<td>AISI 1005</td>
<td>900</td>
<td>550</td>
<td>Yes</td>
<td>1.0</td>
<td>9.5</td>
<td>0.0</td>
<td>0.0</td>
<td>92.7</td>
<td>293</td>
<td>361</td>
<td>37</td>
</tr>
</tbody>
</table>

$T_{en}$ = runout table entry temperature (F6 exit temperature), $T_c$ = target coiling temperature
* Transformation not completed before coiling.
Figure 7.1 Comparison of measured and model predicted (a) F6 exit and (b) coiling temperatures for ten CSP coils.
Figure 7.2 Comparison of predicted and measured temperatures during the rolling of two CSP strips, indicating that the heat of deformation from inhomogeneous deformation has to be included for adequate temperature prediction of thin strips.
Figure 7.3  Comparison of predicted and measured temperatures during runout table cooling in the CSP mill, showing the common cooling patterns for (a) high and (b) low coiling temperatures (0.06C-0.18Mn).
Figure 7.4  Comparison of predicted and measured temperatures at different locations during rolling and runout table cooling of 2.54 mm strips in the CSP mill (0.07C-0.4Mn steel grade).
Radiation
48-51 %

Temperature extracted per roll: 17 - 23 °C
Temperature extracted per descale header: 6 - 9 °C
Temperature extracted per interstand spray: 4 - 5 °C

Figure 7.5 The components of heat extraction in the CSP rolling mill.
Figure 7.6  Comparison of predicted and measured roll forces during the rolling of three CSP strips.
Figure 7.7  Comparison of predicted and measured ferrite grain sizes of ten CSP strips.
Figure 7.8  The computed cooling rate and ferrite fraction for two 0.06C-0.18Mn strips. The microstructures and temperature profiles for both strips were presented in Figures 6.15 and 7.3 respectively.
Steel Grade

Comparison of predicted and measured mechanical properties (YS and UTS) of ten CSP strips.

Figure 7.9
Figure 7.10  Comparison of predicted and measured percent elongation of ten CSP strips.
Figure 7.11 Model Predictions of (a) temperature and (b) austenite grain size evolution during CSP rolling of a 5 mm strip (AISI 1018 steel grade).
Figure 7.12 Model Predictions of (a) temperature and (b) austenite grain size evolution for various strip thickness in a 6-stand CSP mill (AISI 1018 steel grade).
Figure 7.13  Comparison of predicted temperature profiles in (a) 5-stand CSP and (b) 7-stand CCR thin strip rolling (AISI 1018 steel grade).
Figure 7.14: Austenite grain size evolution at the center of the strip rolled in the CSP and CCR mill (AISI 1018 steel grade).
Figure 7.15 Predicted strain contours in the last stand for two CSP strips (AISI 1018 steel grade).
Figure 7.16  Predicted strain rate contours in the last stand for two CSP strips (AISI 1018 steel grade).
Figure 7.17  Through-thickness effective strain profiles during CSP rolling predicted with the Eulerian finite-element model (AISI 1018 steel grade).
Figure 7.18  Temperature increment from plastic deformation during CSP rolling predicted with the Eulerian finite-element model (AISI 1018 steel grade).
Figure 7.19  Comparison of heat generation from plastic deformation for the uniform strain model and FEM predictions (AISI 1018 steel grade).
Figure 7.20 Predicted recrystallization kinetics between stands five and six for CSP rolling of two AISI 1018 strips.
Figure 7.21 Effect of through-thickness strain on the final austenite grain size for two CSP AISI 1018 strips.
Figure 7.22 Through-thickness effective strain profiles in (a) a 2.87 mm CSP strip and (b) a 2.62 mm CCR strip (AISI 1018 steel grade).
Figure 7.23  Effect of through-thickness strain on the final austenite grain size in the (a) CSP and (b) CCR strips (AISI 1018 steel grade).
Figure 7.24  Comparison of predicted and measured/target roll forces for (a) three CSP strips and (b) thin CSP and CCR strips (AISI 1018 steel grade).
Figure 7.25  Predicted recrystallization kinetics in the first interstand of the CSP mill for two initial grain sizes (AISI 1018 steel grade).
Figure 7.26 Comparison of thermal predictions for a grain size dependent constitutive equation with a constitutive equation that does not account for grain size.
Figure 7.27  Comparison of the new low flow rate double-header descaling with the earlier high flow rate single-header descaling during CSP rolling of a 5 mm AISI 1018 strip.
Figure 7.28  Effect of scale formation on the thermal predictions of the uniform strain model for 8 mm AISI 1018 strip.
Figure 7.29 Effect of interstand cooling on the thermal and grain size predictions of the uniform strain model for a 5 mm AISI 1018 strip.
Figure 7.30  Predicted (a) temperature and (b) cooling rate and ferrite fraction for 5 mm AISI 1018 strip showing the effect of coiling temperature.
Figure 7.31 Predicted (a) temperature and (b) cooling rate and ferrite fraction for 5 mm AISI 1018 strip showing the effect of trim bank cooling at three coiling temperatures.
Figure 7.32 Predicted (a) temperature and (b) cooling rate and ferrite fraction for 5 mm AISI 1005 strip showing the effect of coiling temperature.
Figure 7.33  Predicted (a) temperature and (b) cooling rate and ferrite fraction for 5 mm AISI 1005 strip showing the effect of trim bank cooling at three coiling temperatures.
Chapter 8

Summary, Conclusions and Recommendations

8.1 Summary and Conclusions

The present research was directed at adequate prediction of the temperature, deformation behavior (roll force, flow stress, strain and strain rate) and microstructural evolution (recovery, recrystallization, grain growth, austenite and ferrite grain sizes) during Compact Strip Production (CSP) rolling, as well as the final mechanical properties of the hot rolled strips. This was accomplished with the aid of integrated process modeling, involving mathematical simulation, laboratory experiments and industrial campaigns. The study covered two conventional plain carbon steel grades, the A36 (AISI 1018, 0.17C-0.74Mn) and DQSK (AISI 1005, 0.038C-0.3Mn), and a range of steel grades (0.06-0.09 C, 0.16-0.9 Mn) produced at HYLSA's CSP mill at Monterrey, Mexico.

8.1.1 Laboratory Experiments

In the laboratory, compression tests (both single and double-hits) were carried out on the Gleeble 1500 thermomechanical simulator in order to quantify the flow stress and recrystallization behavior of the steels studied under typical CSP mill conditions. Metallographic examinations were conducted on test specimens and grain sizes measured with the image analyzer. Continuous cooling transformation (CCT) tests were also conducted.

The important results of this aspect of the work are summarized below:

(i) Coarse grain sizes of A36 and DQSK steel grades, greater than 500µm and with reasonable homogeneous microstructure, have been developed with the aid of a multi-stage heat treatment in the Gleeble 1500 thermomechanical simulator. These coarse grains were utilized to
elucidate the effect of coarse grain size on the flow stress and recrystallization behavior of the two plain carbon steels.

(ii) Single-hit tests were carried out to measure the flow stress of these coarse grained austenitic steels. It was found that the flow stress at a given strain decreases with increasing grain size provided that the softening mechanism remains the same. An increase in grain size from 244 to 1110 μm which is typical of the first stands of a conventional finishing mill and CSP hot-strip mill respectively, can result in up to a 30 MPa decrease in the flow stress of both A36 and DQSK steel grades. Furthermore, coarse grain size was found to substantially reduce the tendency toward dynamic recrystallization, ensuring that work hardening predominates with a consequent increase in flow stress, a situation that can lead to a higher stress in a coarse grain size compared to a finer one at relatively higher temperatures and low strain rates.

(iii) In combination with flow stress curves for finer grain sizes, it was found that a distinctive boundary exists between flow curves with peak and those without peaks, a very important finding that allowed for a novel quantitative delineation of the occurrence of peaks in flow curves for any given set of deformation conditions.

(iv) It was found that the CSP steel grade with carbon equivalent in the range of 0.12-0.15 exhibited similar hot strengths as DQSK at 1100 °C and 1 s⁻¹ but higher strengths at 1050 °C and 10 s⁻¹. Furthermore, the CSP steels exhibited lower peak strain than DQSK, indicating faster dynamic recrystallization rate. At higher carbon equivalent (> 0.15), the CSP steels showed higher hot strength than DQSK at similar test conditions. The higher hot strength of CSP steels was attributed to the higher alloying and residual contents of the steels.

(v) Double-hit compression tests were performed to evaluate the effect of coarse grain size on the softening rate at the interstands during rolling. It was found that for the range of grain size and strain measured, complete recrystallization (Fₓ ≥ 0.95) occurred between 2 and 4 seconds at
1100 °C and strain rate of about 5 s⁻¹. This finding was further confirmed with metallographic examination of the recrystallized grain size. It was found that the recrystallization kinetics obtained for fine austenite grains cannot be extrapolated to the coarse grain range of A36 while this is not a problem for DQSK. In carbon equivalent range of 0.12 to 0.15, the CSP steels displayed faster recrystallization rate than DQSK due to the shift from static to metadynamic recrystallization at the interstand.

(vi) For the A36 steel, the average recrystallized austenite grain size was measured to be 67 and 142 μm for the 727 and 1110 μm initial coarse austenite grains respectively. The recrystallized austenite grain size of DQSK could not be measured due to the very fast transformation kinetics of the steel.

(vii) With respect to transformation, it was found that transformation starts and finishes at relatively lower temperatures in the CSP steels than in DQSK, the difference increasing as the carbon equivalent increases. For example, at carbon equivalent of 0.12, the transformation start temperature was only 5 °C lower than DQSK while transformation finished at 12 °C lower than in DQSK. When the carbon equivalent increased to 0.24, the difference in transformation start and finish temperatures increased to 76 °C and 92 °C respectively.

8.1.2 Industrial Trial

In order to validate the model and laboratory results with mill measurements from an operating CSP plant, an industrial trial was carried out at HYLSA’s CSP mill in Monterrey, Mexico. During the industrial campaign, intermediate temperature measurements were made, CSP slab and coil samples were acquired, and all measured and recorded mill data and practices were obtained. Room temperature mechanical property tests were performed on coil specimens utilizing an Instrón machine and adhering to the standard ASTM tension test procedure.
Metallographic examinations were also conducted on test specimens and grain sizes measured with the image analyzer. The major results of the industrial trial are summarized below:

(i) It was observed that a variability of 4 to 12 pct. was tolerated in the final strip thickness, this variability decreasing with increasing thickness. For the width, a maximum of 2 mm was allowed below the target width while a variability of 2.5 to 3.6 pct. was tolerated above the target width, this variability decreasing with increasing width.

(ii) Starting from the tunnel furnace exit, the mill entry surface temperature was found to range from 1060 to 1150 °C, and decreases with increasing final strip thickness. The entry speed which range from 0.3 to 0.47 m/s is seen to decrease with decreasing strip thickness. The combination of higher entry temperature and high reduction per pass for thinner gauges ensure that the desired FM exit temperature is attained. The low entry speed of thin strips compensates for the higher speeds emanating from high reductions such that the desired runout table speed is not exceeded.

(iii) Cobbling was found to be the biggest operational problem associated with production of thin strips in the CSP mill while the deterioration of strip shape constituted the major product quality challenge. It was observed that almost all the cobbling occurred during the production of strips in the thickness range of less than 2 mm. In fact, during the plant trail, the cobbling rate during the production of 1.24 mm strips was found to range from 20 to 40 pct. Strip buckling between stands, set up problems, looper control, coiler issues, mill tracking and automatic gauge control were the major causes of cobbling.

(iv) With respect to strip shape, it is more difficult to maintain tight tolerances in the profile and flatness as the strip thickness decreases due to increased deformation inhomogeneity from higher reductions. In fact, the surface quality of the thinnest gauge produced at HYLSA’s CSP mill (a 0.91 mm thick strip) which was meant to replace equivalent cold rolled sheet was found to
be inadequate for direct application that it has to be subjected to skin pass rolling before use. However, there is a microstructural and mechanical property advantage in producing very thin strips, the ferrite grain size can be refined to such an extent that the strength becomes closer to cold rolled strength. It was found that the final grain size of the 0.91 mm strip is about 32 percent finer than that of 2.69 mm strip of the same grade and coiling temperature, a refinement that amounted to 21 percent increase in both YS and UTS.

(v) It was observed that the runout table cooling is very dynamic in nature, continuously changing from the head end of the strip until the coiling temperature is within ± 15 °C of the target temperature. Furthermore, the strip surface is covered with water beyond the impingement point, until the nearest cross spray. The cross sprays located 300 mm from the last jetline in each bank, removes water from the strip surface, the efficiency of water removal is controlled by the water volume at the cross spray. The cross sprays are most efficient when all the headers before it are active, and least efficient when the nearest active header is further away. For coils less than 2 mm, it is common practice to turn on the first two spray headers and one or two headers of the trim bank. For these coils, water covers the strip surface for almost the entire cooling length of the runout table.

(v) The temperature measurements of strip surface which were carried out at various locations in the mill with portable Minolta/Land CYCLOPS pyrometers captured the progressive decrease in temperature from mill entry to exit as well as the temperature decrease due to water cooling from descale sprays.

(vi) At an average strip speed of 7 m/s and with only the first ten jetlines active, an average cooling rate of 30 to 40 °C/s was measured in active cooling zone (that is, between the run out table entry and the portable pyrometer) and 18 to 25 °C/s in the radiation/air cooling zone (that is, from the portable pyrometer to the coiler).
(vii) To gain some understanding of the increase in water temperature as a function of distance from the impingement zone, water temperatures at the runout table were measured for 6 coils with type-K chromel-alumel thermocouples inserted into the water boxes opposite the cross sprays. It was found that the cooling water heats up from its initial temperature of 36 °C/s to about 60 °C/s in a distance of 300 mm and up to 76 °C/s after 2700 mm. The importance of this lies in the fact that heat removal from strip by the cooling water decreases as the water temperature increases.

(viii) The standard deviation in the roll force measurement was found to be higher for the thinner strips than for thicker ones. Detailed analysis of the measured roll force for most of the CSP strips revealed that a maximum of 5 to 10 pct. deviation from the mean measured value for a given coil is common, although the standard deviation of measurement was only about 2 pct. of the mean value at each roll stand.

(ix) Attempt was made to reveal the prior as-cast austenite grain size utilizing various macroetching procedures. A satisfactory result was realized only in the case of the 0.071C-0.899Mn (7094) slab etched in 4 pct. picric acid, where an estimate of 990 μm was obtained. Based on known characterization of ferritic microstructures, the slab micrographs indicated that the solidified dendritic structure transformed into a mixture of Widmanstatten (lath or plate like) and quasi-polygonal ferrite (ferrite grains with irregular grain boundaries) at room temperature.

(x) Thickness measurements revealed that most CSP strips end up with a simple convex strip profile, indicating a higher deformation at strip edges than at the center. The distribution of deformation across the width manifested in the strip microstructure where it was observed that the strip center exhibited coarser grain size than the left and right edges. The variation of microstructure across the strip thickness was also investigated. No tangible through-thickness variation was observed in the final microstructures of the strips.
(xi) The micrographs of a cobbled strip was utilized to study the evolution of grain size from mill entry to exit. It was discovered that the Widmanstatten/quasi-polygonal ferrite structure was preserved until F2 exit, beyond which equiaxed polygonal ferrite is attained. The fact that the Widmanstatten/quasi-polygonal ferrite structure is associated with transformation from the original solidified dendritic structure seems to suggest that it takes the first two stands to break down the cast structure and subsequent stands to refine the resulting equiaxed microstructure through recrystallization as in the conventional rolling.

(xii) The preservation of the Widmanstatten/quasi-polygonal ferrite structure until F2 exit is a very important finding and has ramifications for CSP rolling practice. Firstly, it indicates that in sensitive grades where edge cracking occurs, the rolling conditions at F1 and F2 should be closely monitored to establish safe windows of operation. This might imply adding an additional edger between F1 and F2 in conjunction to the edger that is already being recommended ahead of F1 for these sensitive grades since it is known that a cast microstructure has a greater tendency towards edge cracking. Secondly, it sheds more light on the improvement of quality associated with increased number of stands in CSP rolling. The initial use of four stands implies that equiaxed grain refinement is only occurring at two stands, stands F3 and F4, which might not be enough for structural uniformity since it takes about twelve stands to achieve structural uniformity from reheated austenite in a conventional mill. This finding also has some implications for direct strip casting, a reasonable amount of deformation is necessary to breakdown the solidified microstructure to attain the desired equiaaxed grains of hot rolled strip or cold rolled sheet.

(xiii) Another striking feature of the micrographs of the cobbled strip was the extensive microstructural inhomogeneity at each stand exit, compared to a more homogeneous microstructure just before entry into the next stand, a situation that was so clearly reflected in the surface microstructures. This was thought to be a manifestation of the extent of recrystallization
at each location. The fraction recrystallized at stand exit is minimal compared to the full recrystallization and some grain growth before entrance to the next stand.

(xiv) The final ferrite grain size decreased as the coiling temperature and strip thickness were reduced. Low coiling temperature sometimes lead to non-polygonal structure as was the case in a 3.15 mm thick strip coiled at 560 °C.

(xv) The mechanical properties of the CSP strips was found to be dependent on the coiling temperature, strip thickness and steel composition. The yield and ultimate tensile strengths (YS and UTS) decreased with increasing coiling temperature and strip thickness as well as with reduced carbon equivalent and residual content, while the percent elongation increased. When the coiling temperature decreased from 722 to 560 °C, the YS and UTS increased by 20 and 47 MPa respectively while the percent elongation decreased by 5 pct. A decrease in thickness from 3.15 to 1.52 mm, raised the yield and ultimate tensile strengths by 54.5 and 47 MPa respectively, while the percent elongation decreased by 5.7 pct. YS and UTS increased by 55 and 34 MPa respectively while the percent elongation decreased by 6 percent as the carbon and manganese contents increased from 0.06C-0.18Mn to 0.08C-0.35Mn.

(xvi) To gain some knowledge about the change in mechanical properties during the cooling of coil bundles, tests were carried out at HYLSA to evaluate the change in mechanical properties for 2.54 to 3.15 mm strips of the 0.06C-0.18 Mn (7061 grade) coiled at 715 ± 15 °C. In these tests, samples were removed from the tail end immediately after coiling and just before pickling for mechanical property evaluation. The results for 27 coils indicated a substantial difference in the mechanical properties, the samples taken at the pickling line exhibited on the average, a 25 percent decrease in yield strength, a 10 percent decrease in ultimate tensile strength and a 2 percent increase in percent elongation compared to the samples obtained immediately after coiling. This was attributed to the annealing effect of slow cooling of the coil bundles and the
continuation of transformation at a slow pace in the coil bundle when it is not completed on the runout table.

8.1.3 Model Validation and Results of the Computer Simulation of CSP Rolling

The CSP mill measurements were utilized to validate model predictions of temperature, roll force, grain size and mechanical properties. Comprehensive mathematical modeling of the rolling process was carried out employing finite difference and finite element analysis. An integrated finite-difference model (developed earlier at UBC\textsuperscript{19,120}) that assumes uniform through-thickness strain to predict the thermal and microstructural evolution in the finishing mill was modified and adapted to this study. Sims’s equation was added to the original code for the prediction of roll forces\textsuperscript{121}. To gain a better understanding of the deformation process, another model (also developed earlier at UBC\textsuperscript{63,122}) which utilizes an Eulerian finite-element deformation analysis based on the flow formulation method to predict the through-thickness strain and strain rates as well as the mill load during deformation at each roll stand was adapted. For runout table cooling and austenite decomposition, the newly developed finite-difference runout table model (another UBC model\textsuperscript{45,47}) which utilizes a semi-theoretical analysis of jet impingement and parallel flow boiling to compute strip temperature and the consequent $\gamma \rightarrow \alpha$ phase transformation was also adapted. The model also computes the mechanical properties of the strips with some empirical structure/property equations.

8.1.3.1 Effect of Coarse Grain on Flow Stress and Recrystallization

To ensure that model calculations are relevant to CSP rolling, some contributions were made towards the quantitative description of flow stress of steels and the effect of coarse grain size on the flow stress and recrystallization. A summary of these contributions follows.

(i) A new constitutive equation was developed based on the measurement for the coarse grains and a review of earlier attempts to model the finer grain sizes. The equation is of the form:
\[ \sigma = \sigma_0 + (\sigma_{sr1} - \sigma_0)(1 - \exp(-\beta e))^{n_0} - \Delta \sigma \]

where \( \sigma_0 \), \( \sigma_{sr1} \), \( \beta \), \( \Delta \sigma \) are all functions of temperature, strain rate and grain size. This equation overcame the pitfalls of earlier attempts and were shown to be applicable to all the known conditions operating in the CSP mill as well as the roughing and finishing stands of a conventional mill.

(ii) A quantitative description of peak occurrence in a flow stress curve was developed in terms of the boundary strain rate \( (\dot{\varepsilon}_b) \) or temperature corrected strain rate \( (Z_b) \), below which peak occurs in the flow stress and above which peak will not occur in the strain range of 0 to 1.

\[ \dot{\varepsilon}_b = \exp(a_2 - a_3 d_y - a_4 c / T) \quad \text{or} \quad Z_b = \dot{\varepsilon}_b \exp(Q_{def} / RT) = a_5 \exp(-a_6 d_y) \]

This is the first attempt known to the author to quantitatively delineate the occurrence of peak in a flow curve for a given set of deformation conditions, and has important practical implications. In terms of deformation resistance during rolling, it is now only necessary to compute the reduction in flow stress due to dynamic recrystallization when the applied strain rate is lower than the boundary value \( (\dot{\varepsilon}_b) \), otherwise the recovery equations are sufficient for the determination of the deformation resistance. With respect to the recrystallization kinetics, the equations predict the presence or absence of dynamic/metadynamic recrystallization in conjunction with the critical strain. Dynamic or metadynamic recrystallization will occur when the applied strain rate is lower than the boundary value \( (\dot{\varepsilon}_b) \) and the corresponding applied strain is greater than the critical strain; otherwise static recrystallization takes place at interstand locations.

(iii) A new grain size exponent was proposed for the \( t_{0.5} \) (time for 50 pct. recrystallization) equation of coarse A36 grain size to enable the extrapolation to coarse grain size.

\[ t_{0.5} = Ad_0^p \dot{\varepsilon}^q \dot{\varepsilon} \exp(Q / RT), \quad p = 3.021d_y^{-0.1289} \]
8.1.3.2 Model Validations

With respect to model validation, the following summarizes the result obtained by comparing the mill measurements with model predictions.

(i) It was found that the uniform strain model consistently predicts lower temperatures than the target exit temperature for thin gauges, the discrepancy between prediction and target increasing as the thickness decreases. This was attributed to inadequate estimation of deformation heat when uniform strain distribution is assumed, particularly for thin coils where the contribution of deformation heat could be significant as a result of high reductions. For example, the F6 exit temperature predicted by the uniform strain model matches quite well with measurement for a 6.35 mm strip while the prediction for a 1.524 mm strip is 60 °C lower than the measured value. A good agreement is obtained between prediction and measurement for the 1.524 mm strip only when the heat of deformation from inhomogeneous deformation was included in the calculations.

(ii) By accounting for heat of deformation from inhomogeneous deformation for thin strips, good agreement was observed in all cases between model prediction and pyrometer measurements at F6 exit and coiler entrance, the difference between prediction and measurement was within the tolerance limit of 15 °C allowed in the CSP mill. It is noted that the sum of the standard deviation and pyrometer measurement error (~1 pct. of reading) is approximately 15 and 25 °C for the F6 exit and coiling temperatures respectively. Good agreement was also obtained between prediction and pyrometer measurements at intermediate locations in the mill.

(iii) An estimate of the heat extraction from the various mill sub-units was conducted from the validated calculations. It was found that the descaling unit accounts for 4-6 percent of the total heat extraction, each spray header effectively extracting 6-9 °C from the strip despite the huge temperature drop at the surface. The two interstand sprays accounts for 3-4 percent of the
total heat extraction, each spray header effectively removing 4-5 °C from the strip. The heat loss
to the six rolls was 41-44 percent of the total, each roll effectively removing 17-23 °C from the
strip despite the extensive chilling at the surface. Radiation constitutes the dominant heat loss by
the strips, accounting for 48-51 percent of the total.

(iv) The roll force per unit width, predicted by the finite-element and Sims models were
found to compare favorably with measurement, considering that a maximum deviation of about
10 pct. was associated with the measurements. The finite-element predictions were only
marginally better than the simpler Sims model in this case. The fact that the calculated values
were lower than the measurements in the first stand, was an indication of a higher deformation
resistance than the model assumed.

(v) Reasonable agreement also existed between predicted and measured ferrite grain size,
considering that the prediction was based on the assumption that the transformation behavior of
the CSP steels is exactly the same as the DQSK steel grade. In reality, transformation starts and
completes at lower temperatures in the CSP steels than in DQSK under the same conditions.
Furthermore, a 100 percent polygonal structure was predicted for all cooling conditions, a
situation that was found to be true in laboratory measurements of DQSK steel but does not apply
to the CSP steels.

(vi) A good agreement between predicted and measured mechanical properties was obtained,
particularly when the strengthening effect of residuals are taken into account. The CSP steels
contained 3 to 12 times more copper, 2 to 7 times more tin, and 1.5 to 3 times more nickel than
the DQSK steel grade.

8.1.3.3 Heat Transfer during CSP Rolling

With respect to heat transfer during rolling, the major results of the finite-difference uniform
strain model are summarized below:
(i) It was found that the desired CSP mill exit target temperature of 870 to 920 °C is easily attained by a careful combination of mill entry speed and temperature in both five-stand and six-stand CSP rolling. This is particularly important for thin strips where a slow entry speed (the entry speed is limited by the desired strip speed at the runout table) and low entry temperature could lead to undesirable low exit temperature due to the higher through-thickness heat extraction from thinner strips.

(ii) The major difference between the thermal histories of CSP and conventional cold charge rolling (CCR) lies in the entry temperature, roll-bite heat-transfer coefficient and heat of deformation. The surface to center temperature gradient is higher at the entrance to the finishing mill during CCR rolling than in CSP rolling. The simulations showed that the roll-bite heat-transfer coefficient for the CSP can be up to 25 pct. more than the equivalent CCR gauge while the temperature rise due to deformation can be 2 to 3 times higher, due to the relatively large reduction per stand. The high heat-transfer coefficient and the slower entry speed of the CSP process allows for enough heat extraction within the fewer number of stands to meet the target mill exit temperature which is comparable to that of CCR rolling.

(iii) The fact that the CSP mill operates in a wide range of austenite grain size (~1000 to 20 μm) was found to have consequences for heat transfer. Calculations revealed that the first pass of the CSP mill (1000 μm grain size) generates 25 percent less heat than the equivalent conventional hot strip mill (200 μm) under the same reduction, temperature and strain rate conditions. For a 5 mm final gauge of A36 steel, a 15 °C reduction in the F6-exit surface temperature was predicted when the effect of grain size was not included in the constitutive equation. In general, a constitutive equation which does not take grain size variation into account, will overpredict the heat generation from coarse grains and underpredict the heat generation from fine grains.
(iv) Calculations were carried out to compare the use of the newly designed double-header descaler (with low water flow rate) to the older single-header descaler (with high water flow rate). For the rolling of a 5 mm strip, it was found that the low flow rate double-header design extracted about 30 °C from the strip surface compared to the 50 °C extracted by high flow rate single-header descaler. In both cases, the centerline temperature decreased by only 10 °C in the descaling zone. However, the difference between the strip surface and centerline temperatures for both designs at the F6 exit is only 5 °C. These calculations confirmed that the low flow rate double-header design satisfied the requirement of minimal slab cooling while contributing to a lower incidence of multiple scale-affected zone as reported by mill operators.

(v) Calculations were also carried out to evaluate the effect of primary and secondary scale layer on the strip thermal profile. It was found that the predicted mill exit temperature increased by 35 °C when scale formation was included in the thermal calculations for a 5 mm strip. The secondary scale thickness varied from 28 μm at the entrance to F1, to 8 μm at F6. The insulation effect of the scale was found to be tangible from the exit of the tunnel furnace to the entrance into the third stand (F3).

(vi) Calculations were carried out to investigate the effect of introducing interstand cooling after the first two passes, a practice that some mill operators have utilized to reduce interstand scale growth in CSP rolling. It was found that the mill exit temperature decreased by 50 °C with the introduction of interstand sprays during CSP rolling of a 8 mm strip. This temperature drop can easily be offset by increasing the slab entry speed by 25 percent. Raising the slab temperature at the tunnel furnace exit from 1100 to 1150 °C will also reduce the predicted temperature drop by more than half. Therefore, for a given target exit temperature, an increase in the rolling speed and/or the slab entry temperature is recommended when interstand sprays are utilized such that
interstand cooling becomes a tool for both strip temperature control and increased productivity via higher speed rolling.

8.1.3.4 Recrystallization and Austenite Grain Size Evolution

With respect to recrystallization and austenite grain size evolution, the finite-difference model results are summarized as follows:

(i) The first stand of the CSP mill is essentially its only roughing operation, its main function being the refinement of the as-cast grains. Full recrystallization is desirable in the first interstand of CSP mill to avoid the inhomogeneity that will result from the deformation of a mixture of lower strength (coarse unrecrystallized grains) and higher strength (fine recrystallized grains) material. Full dynamic and metadynamic recrystallization is promoted by high reduction at low strain rate and high temperature. Full static recrystallization is favored by high reduction, high strain rate and temperature with enough time to offset the retarding effect of coarse grain size.

(ii) For the operating conditions at the first interstand of the CSP mill, calculations showed that complete recrystallization is achieved throughout the strip thickness for interstand times greater than 10 s in the AISI 1018 strip. Partial recrystallization prevails at the surface (about 10 to 20 percent of the strip thickness) for interstand times less than 10 s. For example, when the interstand time is 7 s, the surface of a strip with 0.6 mm as-cast austenite grain will achieve only 19 pct. recrystallization (15 pct. for 1.4 mm as-cast grain). The recrystallization rate of AISI 1005 steel is faster than in AISI 1018.

(iii) Grain refinement in the CSP mill was found to be dominated by metadynamic recrystallization in the first three stands and by static recrystallization in the last three stands. CSP rolling is characterized by relatively longer interstand times and lower strain rate when compared to equivalent CCR rolling. From the present analysis, the interstand time was approximately twice that of CCR while the strain rate was about a third of the CCR value. Although the lower
strain rate of the CSP process is expected to encourage more dynamic restoration, the coarse, as-cast austenite grain size offsets this expected effect.

(iv) The predicted final austenite grain size for CSP rolling lies in the range of 18 to 35 μm (ASTM 9-7.5) for AISI 1018 steel and 25 to 40 μm (ASTM 8-6.7) for AISI 1005 steel. It was found that the final austenite grain size resulting from CSP rolling is about 6 μm smaller than that of equivalent CCR rolling despite the initial retardation of recrystallization from the coarse, as-cast grain size. The finer grain size in the CSP process results primarily from the high strains in the final stands, which gives rise to complete recrystallization that consequently refines the microstructure at lower temperatures.

8.1.3.5 Through-thickness Strain Variation and Roll Force

The through-thickness strain and strain rates as well as the mill load during deformation were calculated with the finite-element model. The results can be summarized as follows:

(i) Model predictions of roll forces agreed reasonably well with mill targets and measured roll forces reported in literature, a few discrepancies that occurred at some stands could be due to uncertainties in the temperature and friction conditions. In all cases, the mill load increased from the first stand to the second or third, and then decreased to the lowest load at F6, reflecting the changes in reduction, temperature, strain rate and grain size. The roll force prediction was found to be sensitive to the accuracy of the constitutive equation employed. A constitutive equation which does not take grain size variation into account will overpredict the deformation resistance from coarse grains and underpredict the deformation resistance from fine grains. Calculations showed that the first pass of the CSP mill (1000 μm grain size) requires about 25 percent less force than the equivalent conventional hot strip mill (200 μm) under the same reduction, temperature and strain rate conditions.
(ii) It was found that for both CSP and CCR strip rolling, the strain and strain rate were distributed non-uniformly through the strip thickness and from roll-gap entry to exit, both attaining their highest values close to the strip surface. However, the large reductions employed in CSP rolling allow for through-thickness strains greater than unity for a good percentage of the strip thickness. In the case of strain rate, two peaks were observed near the surface, at the roll-gap entrance and exit, owing to the high redundant shear associated with constrained metal flow in and out of the roll bite. Three major factors control the through-thickness strain inhomogeneity - reduction, strip speed and temperature gradient. The through-thickness strain inhomogeneity increases with increasing reduction, strip speed and temperature gradient from surface to center.

(iii) The through-thickness strain inhomogeneity has implications on the heat transfer through its contribution to heat generation from plastic deformation. It was found that the uniform strain assumption largely underpredicts the heat generation at the surface, from a maximum of 15 °C for 8 mm strip to 100 °C for a 2 mm strip. Therefore, a model based on uniform strain assumption will not accurately predict the thermal profiles, particularly for thin coils where the contribution of deformation heat could be significant as a result of high reductions.

(iv) Through-thickness strain and strain rate variation increased the rate of recrystallization at the surface more than at the center because of the high surface strains and strain rates, a situation that differed from the predictions made with the uniform-strain assumption where the center always recrystallized faster than the surface due to the lower surface temperatures. Through-thickness strain and strain rate variations also resulted in inhomogeneous grain size distribution through the thickness, with the finer grains located near the surface.
8.1.3.6 Runout Table Cooling, Austenite Decomposition and Mechanical Properties

In the runout table, calculations were carried out to investigate the effect of coiling temperature (530 to 730 °C) and two cooling patterns ((a) intense cooling from the early banks until the coiling temperature is reached and (b) moderate cooling at the beginning in conjunction with trim banks at the end) on the microstructure and mechanical properties. The model predictions are summarized below:

(i) With trim bank cooling, transformation barely starts at the first stage of cooling in the AISI 1018 strip compared to the almost completed transformation in the AISI 1005 strip. This is a reflection of the higher transformation start temperature and a faster rate of transformation in the AISI 1005 strips when compared to AISI 1018. For a given coiling temperature, the use of trim banks decreased the tendency towards non-polygonality in the AISI 1018 grade while slightly increasing percent elongation at the expense of strength (YS and UTS decreased).

(ii) For the same coiling temperature, the calculations indicated that the AISI 1005 strip required 20 to 50 percent more active jetlines at the runout table than the AISI 1018 strip. This is also a manifestation of the higher transformation start temperature of the AISI 1005 steel grade, the latent heat released during transformation adds to the amount of heat that must be removed by the cooling water.

(iii) The predicted ferrite grain size lies in the range of 4 to 7 μm for AISI 1018 steel and 9 to 10 μm for AISI 1005 steel. The ferrite grain size was found to decrease with decreasing coiling temperature only when the cooling conditions (cooling rate and austenite grain size) at the start of transformation are different.

(iv) A lower coiling temperature increases the yield strength and ultimate tensile strength while decreasing the percent elongation for both steels. The total nitrogen in solution also rises
with decreasing coiling temperature. The tendency towards pearlite formation and non-polygonality in the AISI 1018 grade was higher at lower coiling temperatures.

(v) The computed yield and ultimate tensile strengths of AISI 1018 steel were found to be 13 to 37 pct. and 33 to 52 pct. higher than that of AISI 1005 respectively, while the percent elongation was 11 to 30 pct. lower.

8.2 Recommendations

A novel attempt has been made in this study to understand the underlying metallurgical phenomena involved in the hot direct rolling of plain carbon strips from thin slab casting, and to suggest optimization routes. The immediate and future challenges of this new technology are envisaged to encompass the following:

(i) the optimization of hot direct rolling of the current product mix to meet more stringent specification such as external automotive application.

(ii) the production of sensitive plain carbon steels grades (e.g. peritectic grade)

(iii) the lucrative rolling of high quality end of the flat steel production (micro-alloyed and interstitial free steel grades)

(iv) the production of thin hot band coils (< 1 mm thickness) on CSP mills to compete with conventional cold rolled products. This might incorporate ferritic rolling.

To ensure that these challenges are met, the following recommendations are suggested:

(1) Comprehensive investigation of the precipitation phenomenon and its consequent effect on grain refinement and mechanical properties of CSP rolled strips. This should include studies of precipitation during continuous casting as well as the mechanism of precipitate redissolution in the tunnel furnace taking into account the relatively short reheating time and the low operating temperature.
(2) In-depth study of the operational and strip quality limitations in the production of thin strips (≤ 1 mm thickness) on CSP mills. In this study, cobbling and poor strip shape were identified.

(3) Investigations into ferritic rolling on CSP mills as a way of producing thin strips to compete with conventional cold rolled products.

(4) Study of the contribution of the unique features of as-cast structure such as segregation, to the hot strength and austenite grain refinement during CSP rolling.

(5) Adequate characterization of the as-cast structure in terms of the grain size. The relationship between dendritic arms and grain size might be useful in this regard.
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APPENDIX A

THERMOPHYSICAL PROPERTIES OF STEEL STRIP AND WORK ROLL

The following thermophysical properties were utilized in the model\textsuperscript{47,120}.

A.1 Steel Strip - Density, Thermal Conductivity and Specific Heat

\[ \rho_{\gamma} \,(kg/m^3) = 8111.44 - 0.5605T \]  
\[ \rho_{\alpha} \,(kg/m^3) = 7870 - 0.1644T - 5.7228 \times 10^{-4} T^2 + 4.5899 \times 10^{-7} T^3 \]  
\[ \rho_{\text{pearlite}} \,(kg/m^3) = 8111.44 - 0.5605T \]  

\( T \) is in °C.

The law of mixture is used when more than a single phase coexists at a given temperature. For example, in the two phase region (\( \gamma+\alpha \)), the density is calculated by:

\[ \rho_{\text{strip}} \,(kg/m^3) = F_{\alpha} \rho_{\alpha} + (1 - F_{\gamma}) \rho_{\gamma} \]  
\[ k_{\gamma} \,(W/m.K) = a + bT \]  
\[ a = 19.09314 - 17.7866 \times C_{\text{arb}} \]  
\[ b = 0.008834 + 0.014706 \times C_{\text{arb}} \]  
\[ k_{\alpha} \,(W/m.K) = 65.422 - 0.052176T + 9.7673 \times 10^{-6} T^2 \]  
\[ k_{\text{pearlite}} \,(W/m.K) = 50.742 - 0.030567T + 1.1539 \times 10^{-7} T^2 \]

\( T \) is in °C.

The law of mixture is also employed when more than a single phase coexists at a given temperature. For example, in the three phase region (\( \gamma+\alpha+\text{pearlite} \)), the thermal conductivity is calculated by:

\[ k_{\text{strip}} \,(W/m.K) = F_{\alpha} \times k_{\alpha} + F_{\text{pearlite}} \times k_{\text{pearlite}} + (1 - F_{\alpha} - F_{\text{pearlite}}) \times k_{\gamma} \]  
\[ C_{p,\gamma} \,(J/kg.K) = c + dT \]  
\[ c = 657.4553 - 414.832 \times C_{\text{arb}} \]  
\[ d = 0.005852 + 0.35783 \times C_{\text{arb}} \]
\[ T > 1060.0 \quad C_p^a(J/\text{kg.K}) = -10034.5 + 5.96681 \times 10^9 T^{-2} \]

\[ 1042.0 < T < 1060.0 \quad C_p^a(J/\text{kg.K}) = 34754.5 - 31.9196 \times 10^9 T \]

\[ 1000.0 < T < 1042.0 \quad C_p^a(J/\text{kg.K}) = -11462.6 + 12.4346 \times 10^9 T \]

\[ 800.0 < T < 1042.0 \quad C_p^a(J/\text{kg.K}) = 4704.5 + 4.5687 \times 10^9 T^{-2} \]

\[ T < 800.0 \quad C_p^a(J/\text{kg.K}) = 503.13 - 0.13068 T - 5.1702 \times 10^6 T^{-2} + 4.4712 \times 10^{-4} T^2 \]

T is in Kelvin

\[ C_{p_{\text{pearlite}}} (J/\text{kg.K}) = 449.5 - 0.4501T \]

(A.11)

T is in °C.

As in the case of density and thermal conductivity, the law of mixture is also employed when more than a single phase coexists at a given temperature. For example, in the three phase region (γ+α+pearlite), the specific heat is calculated by:

\[ C_p^{\text{strip}}(J/\text{kg.K}) = F_\alpha \times C_p^\alpha + F_{\text{pearlite}} \times C_{p_{\text{pearlite}}} + (1 - F_\alpha - F_{\text{pearlite}}) \times C_p^\gamma \]  

(A.12)

A.2 Work Rolls - Density, Thermal Conductivity and Specific Heat

\[ \rho_r = 8000.0, \text{kg/m}^3 \] in all cases.

Cr-Mo Steel

\[ C_{pr} = 497 - 0.17T_r + 0.000366T_r^2 \]  

(A.13)

\[ k_r = \rho_r C_{pr} \left(8.53 \times 10^{-6} + 1.2 \times 10^{-9} T_r - 7.0 \times 10^{-12} T_r^2\right) \]  

(A.14)

High-Cr Iron

\[ C_{pr} = 503 - 0.00781T_r + 0.000462T_r^2 \]  

(A.15)

\[ k_r = \rho_r C_{pr} \left(5.32 \times 10^{-6} + 1.68 \times 10^{-9} T_r - 5.15 \times 10^{-12} T_r^2\right) \]  

(A.16)

Ni-hard Iron

\[ C_{pr} = 541 - 0.1972T_r + 0.000557T_r^2 \]  

(A.17)
$$k_r = \rho_r C_{pr} \left( 4.82 \times 10^{-6} + 3.39 \times 10^{-9} T_r - 5.14 \times 10^{-12} T_r^2 \right) \quad (A.18)$$

$T_r$ is work roll temperature in Kelvin.

### A.3 Enthalpies of Transformation

$$H_{\gamma \rightarrow \alpha} (J/mol) = 221656.4 - 864.4T + 1.9795T^2 - 0.001478T^3; \quad (T \leq 720)$$

$$H_{\gamma \rightarrow \alpha} (J/mol) = -2.917 \times 10^7 + 114590T - 148.8T^2 + 0.063999T^3; \quad (720 < T \leq 780)$$

$$H_{\gamma \rightarrow \alpha} (J/mol) = 3277373 - 10575T + 11.545T^2 - 0.00424T^3; \quad (T > 780)$$

$$H_{\gamma \rightarrow \phi} (J/mol) = 70651 + 225.23T - 0.3469T^2 + 6.755 \times 10^{-5} T^3 \quad (A.20)$$

### A.4 Equilibrium Concentration of Carbon (0.03<C<0.2; 0.3<Mn<2.0)

$$C_\alpha = \alpha_0 \times 10^{-3} - \alpha_1 \times 10^{-6} T; \quad C_\alpha \leq 0.00177 \quad (A.21)$$

$$C_\gamma = \gamma_0 - \gamma_1 \times 10^{-3} T + \gamma_2 \times 10^{-7} T^2 \quad (A.22)$$

where $\alpha_0 = 6.4668 - 1.5852Mn + 0.934C + 1.3612CMn$

$\alpha_1 = 5.4812 - 1.2718Mn + 0.9288C + 0.8839CMn$

$\gamma_0 = 1.1417 - 0.0893Mn + 2.3999C - 4.8483C^2 + 0.5185CMn - 4.117C^2 Mn$

$\gamma_1 = 1.8764 - 0.1054Mn + 4.5524C - 9.0994C^2 + 0.9129CMn - 7.9834C^2 Mn$

$\gamma_2 = 7.7013 - 0.2571Mn + 21.5056C - 42.4328C^2 + 4.2027CMn - 39.1015C^2 Mn$