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AN INTEGRATED SYSTEM FOR ANALYSIS OF METAL FLOW AND
MICROSTRUCTURAL EVOLUTION IN HOT ROLLING

DISSERTATION

Presented in Partial Fulfillment of the Requirements for
the Degree Doctor of Philosophy in the Graduate
School of the Ohio State University

By

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*****

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ABSTRACT

The microstructure and properties of hot rolled steel products are strongly influenced by the processing conditions. Designing a rolling sequence is currently a trial and error process often requiring a large number of production trials to produce parts to the required specifications. Increased competitiveness among steel companies and demand for better quality has accentuated the need to produce rolled steel products with increased consistency, improved tolerances and shorter lead times. There is therefore a strong need for the development of an off-line analysis tool that would enable production of parts to the correct specifications in a short time and at the lowest cost.

The objective of this dissertation work is to develop a computer based integrated system for analyzing metal flow and microstructural evolution during shape rolling. Development of such a system is complicated by the fact that the mechanics of metal flow are strongly influenced by the evolving microstructure which changes continuously during the process. It therefore requires an integrated approach encompassing deformation mechanics and physical metallurgy. In recent years, the finite element method (FEM) based on rigid-plastic formulation has become a powerful tool for analyzing metal forming processes including hot rolling. However, experience shows that FEM based on phenomenological flow stress models, lacks the ability to model
accurately the material flow and rolling loads. The focus of this study is on developing procedures to integrate the microstructural evolution models into FEM.

A three dimensional finite element model for analyzing deformation and heat transfer in rolling was developed and tested under industrial settings during the first phase of this research. Next, semi-empirical microstructural evolution models as well as microstructure dependent flow stress models were developed by conducting controlled laboratory experiments and integrated into the finite element program. One of the controversial topics in microstructural modeling is the application of constant strain rate and temperature models to changing strain rate and temperature conditions. In this dissertation these issues are addresses by means of systematic experiments and procedures to integrate these models into FEM are developed. This integrated approach towards modeling shape rolling has resulted in superior predictions of rolling loads and metal flow. Predictions of metal flow, rolling loads and microstructural evolution were validated by means of industrial hot rolling experiments under production settings.
Dedicated to my parents
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The success of this research has been due to the invaluable contributions of various individuals. I would like to take this opportunity to acknowledge their efforts.

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CHAPTER 1

INTRODUCTION

1.1 Need for an integrated system

New developments in the steel industry have been going along three major paths. The first path is determined by the multi-stage character of the steel production process (steel making, casting, rolling, etc.) with developments concentrated within each of the stages. The second path reflects the multidisciplinary character of the science describing the process (physics, metallurgy, mechanics, computer science, etc.) with developments occurring independently in each of these branches. The third path of development which is attracting a lot of attention involves the interfaces among the different stages of the steel-production process as well as among the different branches of science related to steel production (Ginzburg, 1989).

Of all the known methods of shaping steel, rolling is most widely employed. The process of shaping steel by rolling consists of passing the billet between two rolls revolving at the same peripheral speed in opposite directions and spaced so that the distance between them is less than the height of the incoming billet. The rolls can be either flat or contoured for the hot rolling of rods or shapes. The focus of this study is on
the rolling stage of the steel production process. With computers being increasingly used in manufacturing, various computer aided tools are being developed for applications such as expert systems for roll pass design and finite element modeling of the rolling process to analyze metal flow and temperature changes.

Increasing competitiveness among steel companies has resulted in a need to manufacture steels having tighter tolerances and consistent structure and properties, with reduced trial and error. Traditionally, the rolling process designer was concerned with ensuring appropriate metal flow during the process. In doing so, he normally made use of his experience and a few guidelines that had been established over the years. However, over the last decade, there has been a growing emphasis on rolled product structure and properties. New grades of microalloyed steel that have been developed for cold and warm forging applications call for precise control over the product microstructure. Control over microstructure calls for improved understanding of the effect of rolling mill parameters on the resulting structure. Parameters include preheating time and temperature, rolling deformation, interstand cooling.

Designing a rolling process typically involves a considerable number of preliminary trials to make a product with the desired tolerances and properties. Figure 1.1 shows a typical trial and error process involved in designing a rolling process. An initial billet with a certain geometry and structure needs to be converted to a final rolled product with desired dimensions and properties. In doing so, a preliminary roll pass sequence is designed using past experience or design rules. The designer also has the flexibility to set
some of the rolling parameters such as the initial temperature of the billet, the roll velocities, interstand cooling, etc. The roll pass design together with the process conditions determine what the dimensions and properties of the rolled product are going to be. If these are unsatisfactory, changes are made in the roll pass sequence or the process conditions to try and get the desired product characteristics. This trial and error approach during production not only adds to the cost but also increases the lead time. In today's highly competitive market, steel companies have shown tremendous interest in developing analytical tools that would aid in the design of a hot rolling process by reducing or perhaps even eliminating the trial and error approach. In other words, steel companies are looking for process modeling tools to enable them design the rolling process in an efficient and productive manner.

Figure 1.1 A typical trial and error procedure in designing a roll pass sequence
1.2 Research Objective

Over the past decade and a half, the finite element method (FEM) based on rigid-plastic formulation has become a powerful tool for analyzing metal forming processes including hot rolling. The finite element tool when used to analyze rolling processes can provide a lot of useful information such as metal flow, temperature, rolling loads, stress and strain at different locations in the workpiece. In the past, obtaining the correct dimensions was the primary concern in a rolling process. In recent years, a lot of new steel grades have been developed, many of which are manufactured (rolled) to property specifications; for example: microalloyed steels. It is a known fact that properties of rolled products are directly dependent upon the existing microstructure which in turn depends upon the processing history. This emphasis on manufacturing rolled products to property specifications has therefore resulted in researchers trying to use thermomechanical history to model microstructural evolution during the rolling process.

Most earlier studies on modeling microstructural evolution during hot working process have used the approach outlined in figure 1.2. These studies used a numerical method such as FEM or FDM to analyze the deformation and heat transfer processes during hot working to obtain the history of strain, strain rate and temperature of material points (thermomechanical history) and later used it to model the microstructural evolution using appropriate models. In essence, this approach modeled the material flow independent of the microstructural evolution. Experience shows that mechanics of metal flow are strongly influenced by the evolving microstructure, which changes continuously during the hot rolling process. During hot rolling, strain, strain rate, temperature,
microstructure, and chemistry along with associated metallurgical phenomena such as strain hardening, dynamic recovery and dynamic recrystallization are known to have a significant effect on the flow stress of the material. In other words, the approach depicted in figure 1.2 lacks the ability to predict accurately the material flow and rolling loads. A comprehensive hot rolling model should be able to make accurate predictions of the effect of the process variables on material flow, temperature distribution, microstructural evolution and their consequent effects on product properties. Accurate modeling of hot rolling therefore requires an integrated approach that models the microstructural evolution together with the deformation and heat transfer processes (figure 1.3). The overall objective of this research is to develop one such integrated process modeling tool for simulation of metal flow, heat transfer and microstructural evolution during hot rolling.

Figure 1.2. Conventional approach to modeling microstructural evolution
1.3 Research Approach

The following research procedure is employed in developing the above integrated model:

1. The first step in developing such a model is to establish a reliable tool for analyzing material flow and heat transfer during the rolling process. In this study, the first task is to develop a three dimensional FEM based program ROLPAS for analysis of metal flow and heat transfer during hot rolling.

2. To develop a model for microstructural evolution during the rolling process. The microstructural evolution model comprises relationships for the following mechanisms:

   i) Austenite grain growth in the temperature ranges commonly encountered in rolling.

   ii) Static and metadynamic recrystallization kinetics and the ensuing recrystallized grain size.
Microstructure evolution models are developed by conducting laboratory experiments (for example: hot compression tests) using temperature, strain rate, strain and grain size as the control variables. These models are developed under constant temperature and constant strain rate conditions. Hot rolling on the other hand is a process in which both temperature and strain rate change continuously during the process. Therefore, one of the objectives of this study is to address the issue of applying isothermal and iso-strain rate models under changing temperature and strain rate conditions.

3. Develop a model for flow stress as a function of the evolving microstructure and develop procedures for integrating it into flow based FEM model.

4. To integrate the material / microstructure models into the FEM program (ROLPAS) in order to model austenite evolution during hot rolling. It is expected that this integrated approach to modeling austenite evolution together with the proposed microstructure dependent flow stress model would improve the predictions of material flow and rolling loads.

5. To validate the predictions of the material flow, rolling loads and microstructural evolution by comparing the FEM results with laboratory experiments and actual mill data.

1.4 Research significance and system benefits

This research is being carried out to improve the modeling and analysis of rolled bars using an integrated approach. However, these procedures are also applicable for the
analysis of other rolled products such as plates, strips, etc. Rolled products typically end up as raw materials for hot forging processes where, in many cases, controlled thermomechanical processing is carried out to control microstructural evolution. The procedures developed can be applicable to the analysis of such processes as well.

This research is expected to contribute towards

- Helping the process designer to design a rolling process (roll pass and interstand cooling) for attaining the desired structure and properties without undergoing costly trial and error procedures
- Accurate modeling of material flow in hot rolling of shapes
- Accurate prediction of roll loads
- Prediction of structure and properties for a given roll pass sequence
- Studies on optimizing the robustness of the hot rolling process
- Quality improvements not only in rolled bars but also in end products such as forgings and extrusions.

1.5 Thesis outline

Chapter 1 outlines the motivation behind this work along with the objectives and research approach. Chapter 2 presents a brief overview of microalloyed steels along with principles of thermomechanical processing employed in rolling. A discussion on microstructural changes during the rolling process due to recrystallization and grain growth is presented in chapter 3. This includes a brief discussion on the classical theory
explaining the kinetics of the processes involved. An overview of the methods employed in empirical modeling of the microstructural changes is also presented. Chapter 4 discusses the formulation of the FEM model developed for analyzing the metal flow and heat transfer during hot rolling. Chapter 5 discusses the models developed for modeling the grain growth and recrystallization kinetics and addresses the issue of applying these models under changing temperature and strain rate conditions. Chapter 6 discusses the microstructure dependent flow stress model and its integration into FEM. Validation of the predictions of metal flow, rolling loads and microstructure are presented in chapter 7 along with preliminary results related to modeling of phase transformation. Chapter 8 summarizes the conclusions of this work and discusses the scope for future work.
CHAPTER 2

MICROALLOYED STEELS: AN OVERVIEW

2.1 Introduction

Microalloyed steel is a class of steels having chemical composition similar to those of plain carbon steels with small quantities of alloying elements such as titanium, vanadium, niobium or columbium added to overcome some of the limitations of plain carbon steels.

Typically, plain carbon steels need to be heat treated after hot working to obtain the desired properties (strength, toughness, etc.) in the product. Heat treatment in this case involves quenching from the austenite phase to form martensite which is later tempered at an appropriate temperature to form ferrite and cementite. There are some applications however where it is not possible to conduct post hot working heat treatment operations. Even in applications where heat treatment is possible, ill-effects such as distortion make it necessary for additional machining operations to correct the dimensions thereby adding to production cost and time. Another area of concern in plain carbon steels is the dependence of strength and toughness on the microstructure. In general, the yield strength can be increased and good toughness obtained by refining the
ferrite grain size. The final grain size of ferrite and other transformation products in hot working depends upon the final austenite grain size which in turn depends upon recrystallization and grain growth occurring during the hot working process. Hot working conditions in plain carbon steels typically lead to relatively coarse austenite grains and subsequently coarse ferrite grains with poor strength and toughness.

Microalloyed steels were developed to overcome the above limitations of plain carbon steels. Over the last three decades, several new grades of microalloyed steels have been developed. These are primarily low carbon steels with various combinations of small additions of vanadium, titanium, niobium and columbium. More complex steels may have elements such as Ni, Cr, Mo, Mn, B, etc. to give additional control over austenite decomposition and recrystallization. These microalloying elements are usually present in very small proportions (0.001 to 0.1 % by weight).

2.2 Effect of microalloying elements on microstructure and properties

Alloying elements have a significant effect on the structure and properties of steel and on the iron-carbon phase diagram. Some elements such as manganese and nickel stabilize the austenite phase while others such as silicon, chromium, niobium, etc. stabilize the ferrite phase. Some elements have a strong tendency to form carbides (e.g. titanium, niobium, vanadium, molybdenum, chromium), nitrides and sulphides (Meyer, et al., 1985). One way of measuring the effect of these elements is to measure the shift in $A_1$, the eutectoid transformation temperature of the iron-carbon phase diagram. Austenite stabilizers tend to lower $A_1$ making austenite stable over a wider range while ferrite
stabilizers generally tend to raise $A_1$. Figure 2.1 shows the effect of various elements on $A_1$. In addition to affecting phase transformations and reactions many alloying elements affect mechanical properties by means of precipitation of second phase particles.

![Graph showing the effect of alloying elements on the eutectoid transformation temperature ($A_1$)](image)

Figure 2.1 Effect of alloying elements on the eutectoid transformation temperature ($A_1$) (Krauss, 1990)

Most microalloying elements possess the following characteristics (Meyer, et al. 1985):

- very low content (0.001 to 0.1 % by weight)
- interaction with carbon, nitrogen and/or sulphur
- precipitation of second phases in the matrix
- strong effect on structure
- control of dissolution and precipitation reactions through processing parameters
The microalloying elements affect the structure in terms of

- grain size and shape (through retarding recrystallization and grain growth kinetics)
- precipitates (carbides, nitrides and sulphides)
- ferrite pearlite structure
- non-metallic inclusions (sulphide shape control)

Figure 2.2 shows the effects of some of the microalloying elements on the properties of steel.

<table>
<thead>
<tr>
<th>MAE</th>
<th>Effect on hot formability of CC slabs</th>
<th>Increase of rolling loads</th>
<th>Retardation of recrystallization</th>
<th>Effect on Y/ε transformation rate</th>
<th>Grain refinement</th>
<th>Precipitation strengthening</th>
<th>Texture development</th>
<th>Resistance to ageing</th>
<th>Sulphide shape control</th>
<th>Solid formability</th>
<th>Capability of IF</th>
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</table>

Figure 2.2 Effect of microalloying elements on the properties of steel (Meyer, et al., 1985)
In most microalloyed steels, the desired microstructure and properties are obtained by means of controlled cooling after controlled hot working. Three types of microalloyed steels have been developed from direct cooling processing (Deshpande, 1995)

1. Ferrite-pearlite microalloyed steels
2. Low carbon bainite steels
3. Dual phase microalloyed steels

All of the above steels result in microstructures with good strength and fatigue resistance for many automotive applications.

2.2.1 Ferrite-pearlite microalloyed steels

The most common ferrite-pearlite microalloyed steel contains about 0.1% by weight of vanadium or niobium as the microalloying element and upon cooling directly transforms to fine grained ferrite-pearlite microstructure. The major concern in manufacturing of such steels is the evolution of equiaxed ferrite structure. In thermomechanical processing, the austenite is conditioned to allow it to transform to ferrite with the finest possible grain size, to provide good strength, toughness and ductility. This is done by means of controlled hot working followed by controlled cooling. As mentioned earlier, the precipitation of second phase particles in austenite aids in producing fine grained austenite by retarding the recrystallization and grain growth.
processes. The requirement of the finest ferrite grain size necessitates a high ferrite nucleation rate during transformation. As ferrite nucleates primarily at the austenite grain boundaries, the need for a fine austenite grain size is apparent. The parameter often used to indicate nucleation frequency is the grain boundary area per unit volume which increases with refinement of the recrystallized austenite grains and with increasing elongation of the unrecrystallized grains. The dependence of transformed ferrite grain size on the hot working strain below the recrystallization temperature and on the transformation temperature is shown in figures 2.3(a) and 2.3(b) respectively. Therefore, control over hot working parameters such as reduction, temperature, and the cooling rate after finish rolling are very important in determining the final microstructure.

The precipitates formed in austenite do not contribute to precipitation strengthening in the ferrite. However, only a small portion of the available precipitate is formed in the austenite while the rest precipitates during or after transformation as either interphase precipitation (which occurs during the transformation) or precipitation in the ferrite after transformation. It is this precipitation that gives rise to precipitation strengthening. However, this also leads to a small reduction in toughness.
Figure 2.3 Dependence of ferrite grain size on the initial austenite grain size and strain below the recrystallization temperature

(a) (Pickering, 1983)

(b) (Hansen, 1986)
Another mode of strengthening ferrite is to increase its dislocation density. In the above discussion, it is to be noted that equiaxed ferrite is formed in the slowly cooled austenite. This morphology means that ferrite forms at a relatively high temperature. The dislocation density within the grains is of the order of $10^8 / \text{cm}^3$ which is not high from the point of view of strengthening. Instead, if ferrite is formed at a lower temperature, its morphology becomes more acicular and its recovery does not occur as readily at lower temperatures, the dislocation density can be as high as $10^{10} / \text{cm}^3$. This can be done by the addition of manganese and molybdenum which causes the ferrite formation to be retarded and allows the transformation of austenite at lower temperatures. Such steels are capable of developing yield strengths in the 60 to 100 ksi range (at room temperature).

Figure 2.4 Addition of boron in small amounts (0.0005 to 0.003% by weight) delays further the start of the transformation to ferrite and pearlite (Krauss, 1990)
2.2.2 Low carbon bainite steels

An extension of the acicular ferrite case is to arrange to produce low carbon bainite. In doing so, molybdenum and boron are introduced to form a bay in the TTT diagram. Addition of boron in small amounts (0.0005 to 0.003% by weight) delays further the start of the transformation to ferrite and pearlite (figure 2.4). It does not however seem to affect the time to complete these transformations. Such steels yield in the range 60 to 110 ksi.

In these steels, upper bainite forms typically above 350°C. Bainite is nothing but fine ferrite in the form of laths with a fine array of carbides within it. Carbon rich austenite often gets trapped between ferrite laths and can be either retained or can form high carbon martensite or bainite or can be transformed to produce fine particles of cementite at the boundaries between the ferrite laths. In lower bainite the initial ferrite is supersaturated with carbon due to the lower transformation temperature and leads to the precipitation of small cementite plates within the bainitic ferrite laths.

2.2.3 Dual phase microalloyed steels

In applications such as sheet metal forming, specifically stretch forming, it is desirable that the steel have:

- high ultimate tensile strength
- no yield point because that can cause non-uniform strains
- low yield stress
The ferrite-pearlite steels discussed earlier, deform via an upper and lower yield process and have a high yield strength and are therefore not suitable for sheet formed and press formed parts. The aim is to change the microstructure from ferrite-pearlite to ferrite containing islands of martensite-austenite. This results in continuous yielding as opposed to discontinuous yielding. In order to ensure martensitic transformation does occur during natural cooling, manganese (1-3%) and silicon (0.5-0.9%) are added.

2.3 Strengthening mechanisms in microalloyed steels

Strengthening is achieved by a combination of precipitation strengthening, dislocation strengthening and modification of the matrix structure [Woodhead, 1986]. Solid solution strengthening is not a major factor in microalloyed steels. While the low carbon microalloyed steels are predominantly ferritic, some higher carbon forging steels may have as much as 80% pearlite when air cooled. Many of the forging grade microalloyed steels are ferrite-pearlite steels. The key process issue in these steels is to design the hot working process (rolling, forging, etc.) to produce the desired strength and toughness. For this it is important to have a good understanding of the strengthening mechanisms involved.

2.3.1 Ferrite refinement

The most important structural characteristic of polygonal ferrite is the grain size. Grain refinement not only increases strength but also the toughness. The relationship between the yield stress and grain size is given by the Hall-Petch relationship:
\[ \sigma_y = \sigma_i + k_v d^{-0.5} \]  

(2.1)

where \( \sigma_y \) is the lower yield stress, \( \sigma_i \) is called the internal friction stress and depends upon many factors such as temperature, strain rate, solid strengthening, precipitation strengthening, crystallographic texture, etc., \( d \) is the ferrite grain size while \( k_v \) is a material constant.

The relationship between impact transition temperature (ITT) and grain size is given by:

\[ T = T_0 - b d^{-1/2} \]  

(2.2)

where \( T_0 \) and \( b \) are material constants. The impact transition temperature is a measure of toughness. In general for a given material, lowering the impact transition temperature, raises the room temperature toughness. The fall in transition temperature resulting from grain refinement is fundamental to the development of microalloyed steels with polygonal ferrite matrices. It provides a means of compensating for the rise in transition temperature which results from strengthening by precipitation and makes possible a satisfactory balance between strength and toughness.

2.3.2 Pearlite strengthening

The major factors affecting the strength of pearlite is the inter-lamellar spacing and the pearlite colony size. Many researchers have tried to relate the yield strength of pearlite by a relationship of the following form:

\[ \sigma_y = \sigma_0 + k_s S^{-1/2} \]  

(2.3)
where $\sigma_0$ is an apparent friction stress, $k$, is a material constant and $S$ is the interlamellar spacing. Some other researchers have tried to include the effect of pearlite colony size on the yield strength of pearlite. According to these models, the yield strength increases with reduction in pearlite spacing as well as pearlite colony size.

2.3.3 Precipitation strengthening

The microalloying elements such as V, Ti and Nb all form carbides and nitrides. Typically, the carbides and nitrides that form have the same crystal structure and hence one can expect complete solid solubility between them. In fact, in steels containing Nb or V, one finds that the compounds formed are usually carbonitrides. The composition of the carbonitride depends on the nitrogen/carbon ratio in the steel.

These compounds (carbides, nitrides and carbonitrides) are fairly stable and dissolve in austenite upon heating to the appropriate temperature. Solubility in austenite depends not only on the temperature but also on the type of the compound. For compounds in equilibrium with austenite at any particular temperature, the solubility is highest for $VC_{0.75}$ followed by TiC, NbC, VN, NbN, AlN and TiN (figure 2.5). On cooling after hot working, they precipitate out leading to an increase in the yield and tensile strength of the steel.

Since the temperature for complete solubility of a compound increases as its content increases, there is a limit to increasing strength through precipitation by increasing the amount of microalloying elements. For example, a steel containing
0.45%C and 0.05%Nb will have to be heated to 1300°C for full solution of the carbides (Woodhead, 1986). Such high temperatures typically are not used in hot working of steels.

The solubility of these microalloy compounds is low near the A₁ temperature. However, the precipitation rate is sluggish in austenite and in the absence of deformation, does not occur to any appreciable extent. This is fortunate because any precipitate present in austenite coarsens rapidly and contributes little to strengthening of the steel. In the presence of deformation however, the kinetics of precipitation are accelerated mainly because of the existence of dislocations and other lattice defects which act as nucleation sites (Woodhead, 1987). Strain induced precipitation appears to follow the C-curve kinetics with the nose of the C-curve in the range of 900-950°C for Nb steels. The time to the start of precipitation is very sensitive to the amount of deformation although it is very difficult to give figures for precipitation times as the kinetics are very sensitive to the chemical composition.

Since the solubility of carbides and nitrides in ferrite is very small, any ferrite that forms is grossly supersaturated and precipitation occurs. This precipitation may occur either at the advancing austenite-ferrite boundary (interphase precipitation) or near dislocations (general precipitation). In polygonal ferrite formed at high temperatures, the precipitation is mainly of the interphase type. In acicular ferrite at low temperatures, general precipitation occurs while interphase precipitation is virtually absent. The
strengthening achieved is inversely proportional to the dispersion of precipitates. In other words, the finer the dispersion, the greater is the strengthening.

Figure 2.5 Solubility of carbides and nitrides in austenite as a function of temperature (Korchynsky, 1987)

2.4 Thermomechanical Processing of Microalloyed Steels

Thermomechanical processing involves controlling the microstructural evolution and subsequently properties of the rolled or forged product by exercising control over
process parameters such as temperature, amount of deformation and rate of deformation. Factors influencing evolution of austenite microstructure are recrystallization, grain growth and precipitation of carbides and nitrides. The main objective of thermomechanical treatment is to obtain a fine transformed structure, because both strength and toughness are enhanced. The degree of grain refinement achieved starting with coarse grains at an initial high temperature depends upon the processing history which directly influences recrystallization, grain growth kinetics and precipitation kinetics.

In order to obtain the best combination of strength and toughness in ferrite-pearlite microalloyed steels, rolling is to be carried out under controlled conditions so as to obtain a fine and uniform ferrite grain size. The ferrite nucleation rate is maximized by increasing the number of potential nucleation sites. This can be achieved by austenite grain refinement. Recrystallization under hot working conditions is the primary means of refining austenite grains. The presence of precipitates inhibits the growth of recrystallized grains under rolling conditions and hence aids in obtaining a fine recrystallized austenite microstructure. However, it is generally not possible to refine the austenite grain size to below 15-20 \( \mu m \) diameter by recrystallization which limits the degree of refinement of ferrite grains achieved by nucleation at the austenite grain boundaries. However, deforming the austenite grains below the recrystallization temperature increases nucleation not only at the austenite grain boundaries but also within the austenite grains (Amin and Pickering, 1981).
In general, four important phenomena are involved in thermomechanical processing of microalloyed steels (Leber et al. 1987)

- Dissolution of the microalloy compounds into autenite (reheating stage)
- Deformation and recrystallization in the austenite phase (deformation stage)
- Decomposition of the austenite into either ferrite and pearlite or bainite (transformation stage)
- Reprecipitation of the microalloy compounds

All of the above phenomena need to be considered in designing a rolling process.

2.4.1 Reheating

The evolution of austenite microstructure during thermomechanical processing depends upon the dissolution of certain complex carbonitrides during reheating prior to hot rolling. In general vanadium is more soluble than niobium in typical HSLA steels and can usually be dissolved at the reheating temperatures used in hot rolling. In addition to controlling the amount of microalloying elements dissolved prior to rolling, the reheating temperature also influences the austenite grain size which must be refined during subsequent rolling. On one hand it is advantageous to maximize the dissolution of carbonitrides on reheating to ensure that the desired austenitic microstructure is developed during rolling, on the other hand, it is also essential to avoid austenite grain growth as much as possible. Therefore in slab / billet reheating of microalloyed steels, striking a balance between carbonitride coarsening or dissolution and austenite grain growth is important.
2.4.2 Rolling

Typically, the initial rolling is carried out at a high temperature just below the reheating temperature. At these temperatures, each deformation stage is usually followed by recrystallization and rapid grain growth. The precipitation of carbides and nitrides at lower temperatures restrict grain growth after recrystallization and are used as an effective rolling strategy to obviate the need for low temperature controlled rolling. The addition of microalloying elements also serve to substantially retard the recrystallization kinetics in the latter passes and thereby result in pancaked austenite grain morphology which, as discussed earlier transform to fine grained ferrite. For example, figure 2.6 taken from the work of Cuddy (1981) shows the relationship between austenite grain thickness and ferrite grain size in a microalloyed steel. In this figure equiaxed grain size prior to pancaking was approximately 30 $\mu m$, and the austenitic grain thickness was varied by the amount of deformation below the recrystallization temperature ($T_R$). From this figure, it is easy to see that the finest ferrite grain size are obtained by first obtaining a very small recrystallized austenite grain size at the end of rough rolling and then applying the maximum deformation below the recrystallization temperature. By this approach, ferrite grain sizes as fine as 2-3 microns can be achieved. There are several options available to reduce the finish rolling temperature in bar mills:

- reduce billet soaking temperature
- reduce rolling speed
- introduce delays into the rolling schedule
• introduce interstand cooling

![Graph showing relationship between austenite grain thickness and ferrite grain size.]

Figure 2.6 Relationship between austenite grain thickness and ferrite grain size in a microalloyed steel (Cuddy, 1981).

2.4.3 Transformation to ferrite

The transformation of the "conditioned austenite" determines the final grain size and associated mechanical properties. Even after the minimum austenite grain size and thickness has been produced, controlling the temperature range of the austenite to ferrite transformation is critical. By increasing the ferrite nucleation rate and decreasing the ferrite growth rate, it is possible to attain a finer ferrite grain size. These effects are generally achieved by alloying or controlled cooling. For example, figure 2.7 shows the effect of increasing manganese content (and hence decreasing the Ar3 temperature) on the ferrite grain size generated from austenite rolled 75% below the recrystallization temperature. There is an obvious benefit to increasing the manganese content to about
1.5% for this steel. Molybdenum like manganese is another element that is commonly used to effectively control the $\text{Ar}_3$ temperature for maximum grain refinement.

![Diagram showing the effect of increasing manganese on ferrite grain size](image)

Figure 2.7 Effect of increasing manganese on the ferrite grain size generated from austenite rolled 75% below the recrystallization temperature (Hansen, 1986).

2.5 Conclusion

A primary research area in microalloyed steels is to optimize controlled rolling schedules so as to produce the finest and most uniform ferrite grain structures for maximum strength and toughness. One of the requirements for producing fine ferrite is fine austenite grains or closely spaced austenite grain boundaries. To achieve this structure requires control of microstructural changes in three areas of thermomechanical treatment of austenite: grain coarsening during reheat, recrystallized grain size and grain
shape changes during deformation below the recrystallization temperature. In addition, the control of the transformation conditions also affects the fineness of the ferrite structure that results from the refined austenite.
3.1 Metallurgical changes occurring in the rolling mill

Microstructural changes occurring at different stages in the rolling process affect the final microstructure and properties of the rolled product. A schematic illustration of a typical thermal cycle and microstructural evolution at each stage of the rolling process is shown in figure 3.1. Some of these changes have been described below.

Figure 3.1 Microstructural changes in a hot deformation process (Yoshie et al.)
Microstructural changes during preheating before rolling

A billet or a slab is heated to a certain temperature before rolling. This is referred to as reheating. During the initial stages of reheating, grain growth of ferrite takes place followed by grain size changes during the $\alpha - \gamma$ transformation. Finally, grain growth of austenite occurs. In practice however, grain size is mostly controlled by the grain growth in austenite.

Microstructural changes during rolling

After preheating, the material undergoes deformation in the rolls. Deformation leads to an increase in dislocation density, which causes the material to work harden. Additionally, new defects both point and line, are generated during the deformation process. These act to increase the free energy of the material. At the same time, softening processes such as recovery and recrystallization act to reduce the free energy of the material. These two competing processes, (hardening and softening) together determine the value of the flow stress.

The onset of dynamic recrystallization occurs once a critical strain ($\varepsilon_c$) has been exceeded. The onset of dynamic recrystallization at $\varepsilon_c$ marks the beginning of the strain regime in which new grains are being continually created. In bar rolling, it is believed that the influence of dynamic recrystallization on the final microstructure and properties is relatively small and is often neglected by many researchers.
Microstructural changes in the interstand region

The dislocation structures introduced in a rolling pass drive the microstructural changes in the time interval after the pass. Static microstructural changes typically consist of recovery, recrystallization and subsequent grain growth. Complete recrystallization is usually followed by grain growth if the existing temperature favors it. The extent of these static microstructural changes depends on their kinetics in relation to the time interval before the next pass. Recrystallization, both static and dynamic, results in finer grains as compared to those existing before.

Transformation after rolling

Rolling is mostly carried out with steel in the austenite phase. After rolling, steel transforms to ferrite-pearlite, bainite or martensite depending on the cooling rate.

All of these microstructural changes taking place in the material decisively affect the properties of the final product. These also affect the forming process itself, since the flow stress of the material alters with changes in the microstructure. For precise mathematical modeling of the hot rolling process, it is therefore of great importance to model accurately and take into account the changes taking place in the material during forming and during the interstand period.

3.2 Significance of modeling microstructure evolution

Some of the potential uses of modeling microstructure in rolling are described below.
Final Property Prediction

The mechanical properties of the product are of interest since they determine its service performance. These properties depend on the microstructure. The microstructure of the rolled product is determined by the metallurgical changes that occur during hot rolling. Thus, the microstructure that develops during rolling is the starting point for the final product microstructure and its properties.

Design of the Rolling Process

The major variables which affect the as-rolled structure include furnace (pre-heating) temperature, initial grain size, reductions and rolling speeds at different stands, and cooling conditions between stands. Modeling of microstructure in rolling would help improve the design of the process by identifying how various process parameters affect the structure. In addition to this type of off-line quality improvement, it could also aid in on-line process control. The information of upstream processes can be fed-forward to succeeding controllable processes such as control-cooling and the variations in an earlier stage can be compensated in later stages. This type of on-line process control is however still at the research stage.

Load Prediction During Rolling

The modeling of microstructural evolution could help in modeling metal flow more accurately. Since flow stress is sensitive to the microstructure, modeling the microstructure would help predict the changes in flow stress during deformation. The
accuracy of force estimation, which is important for gauge control in sheet steel, is influenced by whether the effect of changes in microstructure has been considered or not (Sellars, 1990). This application, however, has not been widely investigated yet.

**Better flow prediction**

Material flow is strongly influenced by the flow stress of the material which, as explained above, has a strong dependence on the microstructure, in addition to thermomechanical variables such as strain, strain rate and temperature. Consideration of microstructure in the flow stress model would therefore help in the accurate prediction of material flow and spread in the rolling process.

### 3.3 Modeling Grain Growth

Crystals within a grain have the same orientation. However, crystals in adjacent grains have different orientations. The grain boundary arises as a result of the mismatch between crystal orientations in adjacent grains. This interface between two adjacent grains of the same phase is referred to as the grain boundary. The atomic positions in the grain boundary layer differ from the site positions in the perfect lattice. This leads to an increased energy in the grain boundary layer with respect to the energy of the perfect lattice. The difference in the energy is the grain boundary free energy and is expressed as energy per unit area. Thermodynamic driving force is a decrease in the free energy per unit volume, which causes spontaneous changes in the system under consideration which in this case represent the migration of the individual grain boundaries. It is generally recognized that the driving force for grain growth lies in the decrease in surface energy
brought about by migration of grain boundaries. As the grains grow in size and the numbers decrease, the grain boundary area diminishes and the total surface area is lowered accordingly. In some cases there exists a force inhibiting the grain boundary migration and is called the drag force. Drag forces can arise due to solute segregation and grain boundary pinning effect of precipitates (Novikov, 1997).

The control of grain coarsening behavior of steels is an important step in the design of a thermomechanical process striving to achieve fine grained products (Spiech, 1984). In the process of grain coarsening, the larger grains get larger and the smaller grains tend to disappear. The grain growth may proceed with increase in temperature in one of the following two ways (figure 3.2).

- normal growth that is typical for plain carbon steels.
- discontinuous growth that is typical for microalloyed steels.

![Figure 3.2 Continuous and discontinuous grain growth](image)

Figure 3.2 Continuous and discontinuous grain growth

35
During continuous growth, the grain coarsening of C-Mn steels between 900°C and 1200°C follows the well known grain growth law. In case of discontinuous grain growth, the grain growth process is suppressed up to a certain temperature after which a sudden increase in grain size takes place. The discontinuous grain growth may also start after holding material at the same temperature for longer time periods. The main reason for discontinuous grain growth is suppression of grain boundary migration due to the pinning effect of the undissolved precipitates. At higher temperatures when the precipitates dissolve into the austenite matrix, grain growth occurs in a manner similar to C-Mn steels. Vanadium bearing steels generally resist coarsening only up to 1000 to 1050°C. Niobium bearing steels resist grain coarsening to temperatures up to 1100-1150°C. Steels containing very fine dispersion of titanium nitride (TiN) can resist grain coarsening to temperature above 1200°C.

For microalloyed steels, the reheating temperature should be high enough to provide solubility of the stable precipitates. If these precipitates remain undissolved, the beneficial precipitation hardening effects cannot be obtained. Researchers have reported that the initial reheated grain size has hardly any influence on the final austenite grain size after deformation (Senuma, 1992 and Sellars, 1990).

The migration of grain boundary under the influence of a driving force is a result of atomic displacements from one of the grains to another across their grain boundary and is likened to a diffusion process. In most cases investigated, the atoms pass the grain boundary individually under thermal activation. That is why the grain boundary mobility
$M$ is generally considered to depend upon temperature according to the Arrhenius equation:

$$M = M_0 \exp \left( \frac{Q_g}{RT} \right)$$  \hspace{1cm} (3.1)

where, $M_0$ is a frequency factor and $Q_g$ is the activation energy of grain boundary migration.

### 3.4 Kinetics of grain growth

The “grain size” refers in general to the mean diameter of an aggregate of grains. It therefore follows that the growth kinetics refers to the growth of the average diameter of the aggregate. In modeling grain growth kinetics, it is generally assumed that the growth is proportional to the curvature of the grain boundary of the average cell at any given instant.

$$\frac{dD}{dt} = k\rho$$  \hspace{1cm} (3.2)

where $D$ is the instantaneous grain diameter, $\rho$ is the radius of curvature and $k$ is a temperature dependent constant of proportionality. If it is assumed that the grain boundary curvature is inversely proportional to the grain diameter, the above equation becomes:

$$\frac{dD}{dt} = \frac{K}{D}$$  \hspace{1cm} (3.3)
where the grain growth constant $K$ can be expressed as a function of grain boundary energy ($\gamma_{gb}$) and grain boundary mobility ($M$) as follows (Militzer et al, 1996):

$$K = 3\gamma_{gb}M$$

(3.4)

The grain boundary energy increases with decreasing carbon content according to Gjostein et al. (1966), whereas, at least for low-carbon steels, the grain boundary mobility can be assumed to be that of pure iron. Komatsubara et al. (1990) reported that Mn segregation decreases the grain boundary mobility. However, it is difficult to quantify this effect reliably.

Equation 3.3 on integration under isothermal conditions yields

$$D^2 = D_0^2 + K't$$

(3.5)

where, $D_0$ is the initial grain size, and $K'$ is a temperature and material dependent parameter. Experimental measurements on various materials have shown that the grain size exponent is usually greater than 2. This is attributed to the presence of a drag force in the material which stops the growth process when the grain size reaches a certain critical size ($D_c$). Drag forces arise due to solute segregation and grain boundary pinning effect due to precipitates. The equation describing such a behavior is of the form

$$\frac{dD}{dt} = k(T)\left[\frac{1}{D} - \frac{1}{D_c}\right]$$

(3.6)

This form of equation was first suggested by Burke and can be integrated to obtain
The implicit assumption in deriving both equations 3.5 and 3.7 is that the energy and mobility of all boundaries are identical and that they do not change with time. This assumption is generally not true in the case of textured polycrystalline materials (Novikov, 1997).

Militzer et al (1996) used the statistical grain growth model of Abbruzzese and Lucke to predict grain growth kinetics. In this approach, a pinning parameter accounts for the characteristic inhibition of normal grain growth due to precipitates and solute drag. The actual grain structure is replaced by a spherical grain structure with equal grain volume. The grain size distribution $f(D)$ is subdivided logarithmically into $n$ grain-size classes. The growth or shrinkage of a grain of size class $i$ (grain diameter $D_i$) is considered under the assumption that it is surrounded statistically by grains of all other size classes. If size $i$ is sufficiently greater than size $j$, a driving force $F_i > 0$ exists and grain $i$ tends to consume neighboring grain $j$, where

$$F_i = \gamma_{\text{gb}} \left( \frac{2}{D_j} - \frac{2}{D_i} - P \right)$$

(3.8)

Here, $P$ is related to the pinning force $F_p$ by $P = F_p / \gamma_{\text{gb}}$ and can also be related to the volume fraction and size of pinning particles (Zener, 1948).

The growth (or shrinkage) rate of a grain with size $D_i$ is
\[
\frac{dD_v}{dt} = M \sum_j w_j F_j \tag{3.9}
\]

where \( w_j \) is the probability that grains \( i \) and \( j \) are neighbors.

It is generally supposed that isothermal unpinned grain growth kinetics obeys the parabolic kinetics law (equation 3.5). There is an enormous amount of data on grain growth kinetics in various materials in which the above parabolic law does not fit very well. The difficulty of modeling accurately complex phenomena such as grain boundary pinning has forced researchers to analyze grain growth data using an empirical relationship. Many researchers have used an equation of the following form to model the grain growth kinetics.

\[
D^m = D_0^m + k t \cdot \exp\left(\frac{Q}{RT}\right) \tag{3.10}
\]

Here, \( m, Q \) and \( k \) are material constants and can be obtained by means of a regression analysis on experimental data. A simpler equation of the following form was suggested by Beck.

\[
D = K(T) \cdot t^{1/m} \tag{3.11}
\]

This is equivalent to equation 3.10 when \( D_0 \ll D \).

Failure of experimental data to conform to the parabolic law has been attributed to various factors such as pinning forces due to second-phase precipitates, intrinsic drag forces on grain boundaries, specimen thickness effect, solute drag effect, etc. (Vandameer
and Hu, 1994). While regression analysis can give a fairly good fit with the experimental data, it often tends to mask the metallurgical phenomena that underlie the observations. In other words, parameters like $Q$ (activation energy) in equation 3.10 cease to have any physical meaning if $m$ is different from 2.

3.5 Recovery and recrystallization during hot rolling

Prior to the start of hot rolling, the steel microstructure consists of coarse grains of austenite. While passing through the rolls, the grains get flattened and elongated on the average.

Three kinds of grain restoration processes are associated with hot rolling:

- Dynamic restoration process - This process starts and completes during deformation
- Metadynamic restoration process - This process starts during deformation and completes after deformation
- Static restoration process - This process starts and completes after deformation

3.5.1 Dynamic restoration processes

Dynamic restoration process includes dynamic recovery and dynamic recrystallization. Dynamic recovery is a reduction of work-hardening effects without motion of large-angle grain boundaries.

When steel in the austenitic phase is deformed at high temperatures, the flow stress rises to a maximum and then falls to a steady state as shown in figure 3.3. The
strain corresponding to the peak of the flow curve is an important measure for the onset of dynamic recrystallization. For C-Mn steels, the critical strain for dynamic recrystallization $\varepsilon_c$, is close to the strain at which the stress reaches a peak ($\varepsilon_p$). The critical strain increases as the Zener-Holloman parameter $Z$, given by the following equation increases.

$$ Z = \left( \frac{de}{dt} \right) \exp\left( \frac{Q}{RT} \right) $$  \hspace{1cm} (3.12)

where $de/dt$ is the strain rate, $T$ is the temperature (K), $Q$ is the activation energy, and $R$ is the universal gas constant.

Figure 3.3 A typical flow stress curve under hot working conditions

When the critical strain is exceeded, dynamic recrystallization sets in. The occurrence of dynamic recrystallization provides a softening mechanism thereby
reducing the flow stress to a value between the yield stress and the peak stress. For plate and bar rolling, since the individual pass reductions are generally less than 20%, dynamic recrystallization is unlikely to occur. For strip rolling, dynamic recrystallization may occur since the reductions are higher. Most models, however, treat dynamic recrystallization less rigorously than static recrystallization. Depending on whether or not dynamic recrystallization occurs in steels, the flow stress may differ quite significantly. Recrystallization may cause the flow stress to decrease by as much as 21% (Sellars, 1979).

3.5.2 Metadynamic recrystallization

If the strain is less than the steady state strain, dynamic recrystallization is not complete. Recrystallization continues after the deformation is over and is called "metadynamic recrystallization". In hot rolling of bars, it is the static and metadynamic recrystallization processes that dominate over dynamic recrystallization.

3.5.3 Static Recrystallization in Hot Rolling

Softening that occurs on holding after deformation is the result of static recovery and recrystallization. Static recovery reduces dislocation density by a small amount but does not lead to any movement in grain boundaries and is generally not accounted for by most researchers. On the other hand, static recrystallization results in the formation of new strain free grains by a nucleation and growth mechanism. There is a critical strain
only beyond which static recrystallization will take place. In plain carbon steels, this strain is of the order of 7% (McQueen and Jonas, 1975).

3.6 Modeling static, dynamic and metadynamic recrystallization

Recrystallization like grain growth directly affects the characteristics of the austenite grains. These changes are spontaneous reactions. During plastic deformation, dislocations undergo glide to produce the resulting shape change. During deformation, dislocations interact with one another and form dense tangles which restrict further dislocation glide. In addition, new dislocations are generated during deformation. All of these barriers act to increase the free energy of the material. The processes of recovery and recrystallization occur due to the tendency of the material to reduce its free energy.

The free energy of a deformed metal is greater than that of an annealed metal by an amount approximately equal to the stored strain energy (Cotterill and Mould, 1976). This stored strain energy acts as the driving force for recrystallization. Usually a number of different reactions occur, the total effect of which is the regaining of a condition possessed by the metal before it was deformed. As these processes rely on solid-state diffusion to carry out the transformation, they become most active at elevated temperatures where there is sufficient activation energy and diffusion occurs relatively rapidly.

Recovery and recrystallization both involve dislocation annihilation. In recovery, pairs of dislocations are annihilated to form a stable structure. Recovery does not result in
any change in the grain configuration. In recrystallization, large number of dislocations are annihilated simultaneously and new strain-free grains are formed.

3.6.1 Characteristics of recrystallization

Recrystallization always originates from nuclei which form preferentially in regions where local degree of deformation is highest; such sites include grain boundaries, deformation bands, large inclusions and twin intersections. High angle grain boundaries are the most common sites. The nucleation rate increases sharply with increasing strain above a minimum critical strain. Grain boundary nucleation has been well documented by many researchers for cold worked materials (Cahn, 1965). In hot working of metals, generally nucleation is restricted to grain boundaries. Nucleation at large inclusions is not very well documented. There exists indirect evidence however. Leslie, et al. observed copious nucleation at individual oxide inclusions in cold worked Fe-O alloys. English and Backofen observed enhanced nucleation at inclusions in hot worked silicon-iron. Large inclusions tend to enhance nucleation rate while on the other hand, finely spaced inclusions retard it. While grain-growth control by finely dispersed particles is well established, the role of grain boundary drag in retarding recrystallization in metals containing finely dispersed particles is less well established. Leslie et al (1963) showed that coarse oxide inclusions in an Fe-O alloy enhanced nucleation. Isolated inclusions, 1 micron or more in diameter in an alloy deformed by 60% caused an acceleration of the recrystallization process. Below this diameter, precipitates tended to slow down recrystallization. The enhancement of nucleation for large inclusions is apparently due to
the local enhancement of the dislocation population (Cahn, 1965). Detert and Ziebs (1965) have reported that fine precipitates at the cell walls in Ni-Ta-O alloys prevent the activation of nuclei and thus inhibit recrystallization.

3.6.2 Thermodynamics of Recrystallization

Recrystallization like phase transformation is a nucleation and growth process. In the case of recrystallization, the transforming phase is represented by the deformed metal (α) and the newly transformed phase by the strain-free (recrystallized) regions (β). The new phase is nucleated at points of high strain energy such as slip-line intersections and areas close to the grain boundaries. The free energy of small regions of the new phase β are made up of the sum of free energy of their surface and that of their volume. Such initial β phases are referred to as embryos and are assumed to be spherical in shape. This assumption implies that the embryo forms by a process based on random fluctuation and does not involve any type of direction-dependent formation. This assumption is not necessarily true for the formation of recrystallization nuclei (Cotterill and Mould, 1996). This assumption because of its simplicity and conformance to many nucleation and growth processes is widely used nonetheless. As the embryo increases in size, the decrease in volume free energy favors its expansion while the increase in surface energy tends to counteract this. The change in free energy on forming the embryo is therefore given by:

\[ \Delta F_r = 4\pi r^2 \gamma - \frac{4}{3}\pi r^3 \Delta F_i. \]  

(3.13)
where, $\Delta F_T$ is the total free energy change, $r$ is the radius of the embryo, $\gamma$ the specific interfacial energy and $\Delta F_r$ the bulk free energy change ($= E$, the internal energy change). The volume free energy change in the above treatment is therefore represented by the difference in strain energy between the worked and the fully recrystallized states. Since the two free energy terms have opposite effects on the overall free energy change, there is a critical embryo radius below which the increase in surface energy outweighs the decrease in volume free energy. The critical embryo radius is given by

$$r_c = -\frac{2\gamma}{E} \quad (3.14)$$

Hence the free energy change accompanying the formation of a viable recrystallization (stable embryo) is given by

$$\Delta F_{r\text{ (c)}} = \frac{16\pi\gamma^3}{3E^2} \quad (3.15)$$

The incubation period usually observed before recrystallization is the time required for a series of thermally activated fluctuations to form stable nuclei.

3.6.3 Kinetics of recrystallization

Kinetic theory in contrast to thermodynamics attempts to derive relationships starting with atomic and molecular processes. Recrystallization being a typical nucleation and growth process, begins slowly and builds up to a maximum reaction rate after which it finishes slowly. The fraction recrystallized plotted against time takes a characteristic S-shaped curve as shown in figure 3.4. Classical work on modeling such a nucleation and
growth process was made by Johnson and Mehl (1939). Avrami later extended this work and the model came to be known as the JMA equation. Using an assumption of constant nucleation rate, Johnson and Mehl derived the following equation for fraction recrystallization (or transformation) in a three dimensional nucleation and growth process at some time $t$.

\[ X = 1 - \exp \left( -\frac{fG^3 \tilde{N}t^4}{4} \right) \]  

(3.16)

Here, $f$ is the shape factor. $G$ is the linear growth rate and $\tilde{N}$ is the nucleation rate.

Raising serious objections to the assumption of constant nucleation rate Avrami proposed a variation of the above Johnson and Mehl model based on a rigorous analysis and developed equations for various cases. In the JMA model, it is assumed that the
nucleation rate is a function of temperature and has a fixed value at a certain temperature and is given by the usual Arrhenius equation, i.e.

\[ \dot{N} = A \exp \left( \frac{Q_n}{RT} \right) \]  

(3.17)

where, \( Q_n \) is the activation energy for nucleation. The equations developed by Avrami for different cases were of the following form.

\[ X = 1 - \exp\left(-Bt^k\right) \]

(3.18)

where, \( B \) and \( k \) are constants.

Interestingly, the same formula was developed experimentally by Austin and Rickitt for isothermal transformation of austenite to bainite earlier in 1938. However, Avrami was the first to provide a theoretical basis for it.

The nucleation rate is not actually a constant even at a particular temperature. Its true form is such that it is more accurate to make experimental measurements than to assume any particular mathematical form. The conventional technique for measuring the nucleation rate is to plot a graph of the number of new grains which appear as a function of time. The nucleation rate is then taken as the slope of that graph (Cotterill and Mould, 1976).

In modeling recrystallization kinetics, it is usual to assume that the transforming duplex structure consists only of the strained and recrystallized regions. However, studies have shown that the transforming duplex consists of a mixture of recovering and
recrystallized regions indicating that at any stage of the recrystallization process, pre-recrystallization softening continues to occur in those regions of the material which have not been consumed by the advancement of new regions. The extent of this continued recovery softening is slight and is often neglected. Vandameer and Gordon (1961) have shown that in cases where recovery and recrystallization occur simultaneously, the driving force for recrystallization decreases continuously because the stored energy of the unrecrystallized regions ahead of the recrystallizing regions is now lower due to the continued occurrence of recovery.

3.6.4 Empirical models for recrystallization

The theory of recrystallization, which describes the nucleation and growth processes of recrystallized grains in a deformed matrix is applicable only to metals with extreme purity. In these metals, experimental results agree quantitatively with the predictions of the theory. In alloys, however, the rate of recrystallization, the temperature dependence and structure changes are drastically different than in high-purity metals and cannot be explained by existing theories.

Due to the difficulty in modeling accurately the recrystallization processes theoretically, researchers have resorted to empirical techniques. Many researchers have developed semi-empirical relationships for modeling the microstructural changes occurring in steels due to recrystallization. These quantitative microstructural relationships correlate the austenite grain size to process variables, such as temperature, strain, strain-rate, and the initial grain size. The important equations in these models are:
• Strain for 50% dynamic recrystallization ($\varepsilon_{0.5}$)

• Time for 50% static or metadynamic recrystallization ($t_{0.5}$)

• Recrystallized grain size ($d_{rec}$)

The Avrami type equation is generally used to describe the sigmoidal curve for the fraction of dynamic as well as the static recrystallization. The volumetric fraction of dynamic recrystallization at a given instant is a function of the strain for 50% recrystallization, and is given by

$$X = 1 - \exp\left[-0.693\left(\frac{\varepsilon - \varepsilon_c}{\varepsilon_{0.5} - \varepsilon_c}\right)^k\right]$$  \hspace{1cm} (3.19)

where $\varepsilon_c$ is the strain for the start of dynamic recrystallization, $\varepsilon_{0.5}$ is the strain at which 50% dynamic recrystallization occurs.

Similarly, the static recrystallization at a given instant is a function of the time for 50% recrystallization, and is given by the following Avrami equation:

$$X = 1 - \exp\left[-0.693\left(t / t_{0.5}\right)^k\right]$$  \hspace{1cm} (3.20)

where $X =$ fraction recrystallized at any instant, $t$ is the time elapsed after deformation, $t_{0.5}$ is the time for 50% static recrystallization, and $k =$ Avrami coefficient (different for dynamic and static recrystallization). These types of equations (equation 3.19 and 3.20) have been used extensively by many researchers since the pioneering work by Sellars and coworkers (1979). These models are characterized by the time for 50% recrystallization and $k$ which is assumed to be constant for a material. The time for 50% recrystallization...
is modeled as a function of strain ($\varepsilon$), strain rate ($\dot{\varepsilon}$), temperature $T$ and initial austenite grain size ($d_0$).

\[ t_{0.5} = A \cdot d_0 \varepsilon^n \varepsilon \exp\left(\frac{Q}{RT}\right) \]  

(3.21)

where, $Q$ is the activation energy for recrystallization, $A$, $l$, $m$ and $n$ are material constants. Many researchers have used an equation of similar form to model the recrystallized grain size ($d_{\text{rec}}$):

\[ d_{\text{rec}} = B d_0 \varepsilon^{y \varepsilon^z} \exp\left(\frac{Q_{\text{app}}}{RT}\right) \]  

(3.22)

where, $x$, $y$, $z$ and $Q_{\text{app}}$ are material constants.

Some researchers have used slightly different approaches. Saito et al (1992) use the generalized Kolmogorov-Johnson & Mehl-Avrami (KJMA) equation for modeling recrystallization. The generalized KJMA equation is given as;

\[ X(t) = 1 - \exp\left[\int_0^t J(t') V(t, t') dt'\right] \]  

(3.23)

where $X(t)$ is the fraction recrystallized, $J(t')$ is the nucleation rate at time $t'$ and $V(t, t')$ is the volume at time $t$ of the stable nucleus at time $t'$. This approach is complex and requires determination of a large amounts of constants in associated equations.
Yoshie et al (1992) have developed equations for elongation of grains, refining of grains due to recrystallization and increase in lattice defects when rolling at non recrystallization temperatures. The equations used are:

1. **Grain elongation:**

   \[ S_g = \left( \frac{24}{\pi D_0} \right) \{0.491 \cdot \exp(\varepsilon) + 0.155 \cdot \exp(-\varepsilon) + 0.143 \cdot \exp(-3\varepsilon)\} \]  
   \hspace{1cm} (3.24)

   where, \( S_g \) is the area per unit volume of grain boundary, and \( D_0 \) is the grain size before rolling.

2. **Grain refinement**

   The recrystallized grain size and fraction recrystallized at some instant \( t \) are given by

   \[ d_{rec} = \frac{5.0}{(S_g \cdot \varepsilon)^{0.7}} \]  
   \hspace{1cm} (3.25)

   \[ X_{rec} = 1 - \exp\{-e(t - \tau)^n\} \]  
   \hspace{1cm} (3.26)

   where \( e \) is a coefficient, \( n \) a constant and \( \tau \) the incubation period. This is an Avrami type equation similar to the equation used by Sellars and other researchers. In this study, only static recrystallization was considered.

Siwecki et al (1992) used equations similar to those of Sellars to model static recrystallization in a microalloyed steel. The static recrystallization for a Ti-V microalloyed steel was modeled using the following equations:
They also observed that, Nb and Ti-V-Nb microalloyed steels showed a sluggish recrystallization behavior compared to Ti-V and C-Mn steels especially at temperatures below 1000°C. This is expected due to the pinning effect of undissolved precipitates.

3.6.5 Double hit compression test for modeling recrystallization kinetics

The critical strain for the onset of static recrystallization is generally less than 10%. If the critical strain is exceeded, an incubation period must elapse before recrystallization occurs. The length of this period depends upon the strain, strain rate, temperature and grain size. If the critical strain for recrystallization is not reached, the recovery process does not lead to full softening but saturates at a softening level much below 100%. In other words, full softening cannot be achieved by recovery processes alone. The existence of critical strain for full softening can be studied using a technique of interrupted compression testing (double hit compression tests). In this technique, samples are deformed at a constant strain rate and then at a selected strain rate, they are unloaded, held for increasing delay times and then reloaded at the original strain rate (Jonas, et al, 1976). Figure 3.5 shows an example of stress-strain curve obtained from such compression tests. In these tests, the fractional softening that takes place during the inter-hit time can be evaluated as

\[ X = 1 - \exp(-0.693(t / t_{0.5})^{1.7}) \]  \hspace{1cm} (3.27)

\[ t_{0.5} = 5.0 \cdot 10^{-18} \cdot (e - 0.058)^{-1.5} d_0^2 \exp(280000 / RT) \]  \hspace{1cm} (3.28)

\[ d_{\text{neu}} = -1.25 = 24.4(V + Nb)^{-0.2} \cdot N^{-0.04} \cdot d_0^{0.25} \cdot e^{-0.55} \left[ \exp(350 / RT) \right]^{-0.07} \]  \hspace{1cm} (3.29)
\[ FS = \frac{\text{flow stress on interruption} - \text{reloading yield stress}}{\text{flow stress on interruption} - \text{original yield stress}} \]

Figure 3.5 Stress-strain data in a two hit interrupted compression test

Figure 3.6 Interaction between static recovery, static and metadynamic recrystallization
When strains are high, metadynamic recrystallization is observed instead of classical static recrystallization. As this type of recrystallization does not require any incubation period, it proceeds very rapidly upon the termination of deformation. The interrelation between static recovery, static recrystallization and metadynamic recrystallization depends on the interruption strain and is depicted schematically in figure 3.6. This figure shows that for deformations below the critical strain, the only restoration process is static recovery. For strains which are larger but below the peak strain, static recovery is followed after a suitable interval by static recrystallization. For strains greater than the peak but less than the steady state strain, there is a period of concurrent static recovery and metadynamic recrystallization followed by static recrystallization. Finally, once the steady state strain is exceeded, only metadynamic recrystallization takes place with some concurrent static recovery. To determine fraction static recrystallization in a two hit compression test, as a rule of thumb, it is generally assumed that about 20% of the initial softening observed is due to recovery. Therefore the fraction of static recrystallization is calculated as

$$X = \frac{FS - 0.2}{0.8}$$ (3.30)

### 3.7 Effect of alloying elements on recrystallization

In most metals, the rate of recrystallization can be reduced by the addition of solute elements. An increase in solute concentration can also lead to finer recrystallized grain sizes. One of the reasons for this behavior is the drag effect of impurity atoms on
moving grain boundaries. Another reason is the decrease in the density of possible nucleation sites caused by impurity addition. Figure 3.7 illustrates the effect of alloying elements by comparing two carbon steels containing Nb with three unmodified plain carbon steels. It can be seen that substantial softening by recrystallization takes at least an order of magnitude longer in the niobium steels than in the plain carbon steels.

![Figure 3.7 Effect of Nb addition on the recrystallization rates of plain carbon steels at about 0.68 Tm (McQueen and Jonas, 1976)](image)

**3.8 Application of microstructural modeling to hot rolling**

Sellars (1990) has compared some of the different available microstructure models for multi-pass rolling. While the general trends pertaining to $t_{0.5}$, $d_{rel}$, and $d$ predicted by each of these models are similar, they exhibit considerable differences for
specific conditions. Sellars found that assuming complete static recrystallization between passes (sufficient dwell time between passes), all the models predicted that:

- Overall, the austenite grain size refines progressively with successive passes as a result of successive recrystallization.
- Grain growth following recrystallization is an important mechanism controlling the grain size entering the next pass.
- The grain sizes calculated with the different relationships converged with increasing number of passes, so that the variation in the final grain size was relatively low (35-50 μm); Significant deviations existed for the early passes, but this may not be of much practical relevance since the final as-rolled grain sizes were similar.

If recrystallization is complete between passes, the modeling is relatively simpler and the structures are more homogenous (Sellars et al, 1979). This is the case for slower mills like reversing mills, with long interpass times. Static or metadynamic recrystallization is followed by grain growth.

However, if recrystallization is not complete, some of the strain is retained in the structure. The structure then contains regions of different grain sizes and different retained strains. When recrystallization is incomplete, different treatments for modeling of partial recrystallization exist. For example, Yada (1990) computes a mean grain size and computes a residual strain based on changes to the dislocation density which is added to the subsequent nominal strain. Researchers at Kawasaki Steel (1985), Karhausen and Kopp (1993), etc. consider the recrystallized and unrecrystallized portions separately. For
the unrecrystallized portion, they add retained strain to the strain in the subsequent pass. This treatment could however become complicated as it leads to a large number of fractions soon.

**Influence of Experimental Uncertainties on Results**

When using these semi-empirical models for prediction purposes, it is necessary to have some estimate of how much error will be introduced in the predictions due to uncertainties in the values of constants in these models. Differences in the values of recrystallization constants to be used in the equations did not seem to affect the end results significantly (Sellars et al, 1979). When using these models, changing the plate rolling schedule parameters such as pre-heating temperature, or reduction in the final passes, did lead to very significant differences in the end result. Delays in pre-heating did not produce large differences in the final structure. These results are consistent with actual mill observations.

**3.9 Computer modeling of microstructural evolution**

Sellars (1979) pioneered the modeling of microstructural evolution during rolling. In one of his studies, he modeled the recrystallization and grain growth for a C-Mn steel in a multipass rolling schedule using a computer model. The rolling schedule consisted of rolling a plate of 20mm thickness starting with a 250 mm slab in equal passes of 15% reduction with 20 s between all passes. He found that the grain size is refined by very rapid recrystallization after each roughing pass and when grain size is below 100 mm, some subsequent grain growth occurred before entry to the next pass. The
recrystallization behavior is shown in figure 3.8. By changing the recrystallization rate (decreasing by a factor of 5), he saw that the overall pattern of microstructural changes remained similar although incomplete recrystallization occurred between all finishing passes. This leads to believe that different formulations developed by different research groups and some differences in details of the computational procedures are all found to lead to good quantitative agreement between predicted and observed structures and rolling loads. Modeling has therefore become an accepted and a reliable tool for understanding and optimizing the thermomechanical process of steels.

Figure 3.8 Grain refinement modeled in a plate rolling process (Sellars 1979)
Several researchers have attempted at computer modeling of microstructure in rolling. Notable early works were by Sellars (1979), Yada (1983) and Roberts (1984). While these early studies focused on the microstructural changes during the rolling process, recent efforts have been made at developing an integrated model for modeling microstructural evolution, phase transformation and correlating microstructure and ensuing material properties. Some notable models have been reported by Suehiro et al (1987), Yoshie et al (1992), Watanabe et al (1992), Saito et al (1992), and Anelli et al (1992).

Watanabe et al used empirical equations for computing strain and strain rates during the rolling process and a finite difference method for computing the temperature at multiple points in the billet. While this approach is faster than FEM, its accuracy is limited and does not allow for the modeling of static and dynamic recrystallization separately. Anelli et al and Saito et al modeled the rolling and microstructural evolution process separately. As a result, the model fails to consider the effects of microstructural changes in the interstand region on the subsequent pass. This approach will not give accurate prediction of load and metal flow. The models by Nanba et al (1992) and Kern et al (1992) use a similar approach and therefore suffer from the same drawback.
3.10 Application of FEM to model microstructure evolution

Pietrzyk (1990) has coupled the recrystallization and grain growth relationships available in literature to a finite element model. This allows the use of local values at each node for temperature, strain, and strain-rate, rather than average values in the microstructural models, helping predict the microstructure distribution more accurately. The mechanical component of his model was based on the rigid-plastic FE method. An algorithm was used that calculates at the exit of each pass, whether static or dynamic recrystallization occurs, the time for grain growth, and the grain size after growth.

Pietrzyk et al (1993) conducted a number of rolling experiments with different reductions per pass, number of passes, and with different C-Mn steel grades. The strips were instrumented with thermocouples. The temperature at the entry to each pass, interpass times, roll separating force, and roll torque were recorded. There was reported to be good agreement between measured and experimental data for roll forces and torques. All the microstructural models from literature mentioned earlier, used in conjunction with the same FE code, gave proper qualitative assessment. Calculated grain sizes decreased towards the surface (due to lower temperature at surface) and became finer after the second and third passes, as was seen in experiments. Overall, among the equations of Sellars, Choquet, Roberts, Jonas, and Yada, Roberts’ relationships compared the best with experimental results. But in general, there were wide variations among the predictions by the different models. One reason is that each model was arrived at based on experiments on a particular steel grade. Hence, using the model to predict microstructure in a different C-Mn steel grade could lead to some error. Also, most of
these models assume grain growth and recrystallization kinetics at a constant temperature and a constant strain rate. In reality, heat is lost to the surroundings continuously during rolling and in the interpass intervals. The strain rate also changes continuously during rolling. Therefore isothermal and iso-strain rate conditions do not apply in the case of rolling.

Rantanen et al (1990) coupled the elastic-plastic FE code ABAQUS with a microstructure modeling program called MICROPLA. Data from plate rolling of a low carbon steel, with a schedule consisting of 8 passes, was used for validation. Their model assumed that the steel recrystallizes fully after each pass. The estimated rolling force differed from the actual load by less than 4%. The strain and temperature distributions calculated using ABAQUS were used in the microstructure analysis. The predicted average ferrite grain size and yield strength near the surface and center were close to the experimental values (Korhonen et al, 1991).

3.11 Interaction between Flow Stress and Microstructure

As emphasized earlier, flow stress of the material at any instant during a hot working process must be considered as dependent upon its thermomechanical history. Models with evolving internal structure variables are gaining in popularity. These "unified" models characterize the mechanical behavior of a given material under arbitrary loading conditions by defining its instantaneous inelastic strain-rate response in terms of state variables which include temperature, stress, and microstructure.
\[ \dot{\varepsilon}_p = \dot{\varepsilon}_p(\sigma, T, \alpha_1, \alpha_2, \ldots, \alpha_n) \]  
(3.31)

\[ \dot{\alpha}_i = \dot{\alpha}_i(\dot{\varepsilon}_p, \ldots) \]  
(3.32)

These equations are capable of modeling accurately the complete range of physically observed behavior, since the structure parameters in the model, \( \alpha_n \), can evolve in the same way as their associated microstructural properties that characterize the material's resistance to plastic flow. Many of these models with evolving structure variables are difficult to integrate numerically and implement into existing FEM programs.

Kozlowski et al (1992), investigate four different forms of constitutive relations; constant structure, time hardening, strain hardening and simultaneous time and strain hardening models; for use in finite element stress analysis of processes such as continuous casting. Karhausen et al (1990), modeled the flow stress according to the following relations:

\[ \sigma = f(S, \frac{de}{dt}, T) \]  
(3.33)

\[ \frac{dS}{de} = g(S, \frac{de}{dt}, T) \]  
(3.34)

where \( \sigma \) is the flow stress, \( \varepsilon \) is the strain, \( T \) is the temperature, and \( S \) represents the structure. In their approach, the initial structure is input to their program. FEM provides forming parameters \( (\varepsilon, \frac{de}{dt}, T) \) incrementally, which are fed into the microstructure simulation model, consisting of Sellars' semi-empirical equations. Microstructural states at any instant during forming are evaluated successively on the basis of these equations. For steels, recovery is very slow. Steels work-harden till they start softening due to
dynamic recrystallization, which begins at the critical strain ($\varepsilon_c$). The competing processes of hardening and recrystallization lead to a steady-state stress at a certain strain ($\varepsilon_S$). Different regions in the rolled section experience different levels of strain and soften to different degrees. In Karhausen's work, an algorithm is set up to calculate for each time increment, the volume fractions of these various substructures. The flow stress formulation used is structure-sensitive. Based on the microstructural state, the corresponding flow stress for each substructure is inferred. Thus, microstructure and FEM are coupled in their approach.

The model was validated by simulation of flow stress curves obtained during upsetting of specimens of a chromium-vanadium steel. The characteristic variables for the steel (kinetics constants, flow stress related constants, and initial austenite grain sizes at different temperatures) were determined by upset tests. The flow stress curves predicted by their approach were found to be close to the actual flow stress curves. While this approach simulated the upsetting process very closely, it is inherently complex and computationally intensive. This is primarily due to the multiplication effect of evolving substructures. The number of substructures existing at the $n^{th}$ time-step is $2^n$. Storage of thermomechanical history of these many structures results in an enormous amount of required memory. This makes its application to multi-pass rolling difficult.

Laasraoui and Jonas (1990) have also formulated constitutive equations which describe the high temperature-high strain rate flow behavior of steels. They investigated the behavior of four different grades of steels and developed relations for the flow stress that
take into account the softening effect due to dynamic recrystallization. Their model making simultaneous use of the Bergstrom model and fractional softening by dynamic recrystallization was found to give accurate prediction of the flow stress after the peak. A similar model was used by Beynon and Sellars (1992) in their rolling simulation model. Another approach involving the use of a dislocation model was used by Yada (1990) for predicting the austenite grain size.

The grain structure of dynamically recrystallized material affects the high temperature flow stress in two distinct ways. In the first place, the volume fraction that has recently recrystallized determines the degree to which the flow stress approaches that of statically recrystallized material on one hand and that of work-hardened material on the other. Secondly, the grain size itself can influence the high temperature flow stress when the grain boundaries act so as to impede the propagation of slip throughout the material. With regard to the effect of grain size on the flow stress, the situation is not very clear (McQueen and Jonas). Under most conditions, a relationship between steady state flow stress \( \sigma \), and grain size \( d \), of the following form is observed

\[
\sigma = M d^{-q}
\]

(3.35)

where, \( M \) is an empirical constant and \( q \) has been observed to be about 1.0 for various materials. There is clear evidence that for grain sizes above 20 microns, the high temperature yield stress increases with decreasing grain size (McQueen and Jonas). The strengthening component due to grain sizes has been reported to follow the Hall-Petch relation. Jonas further goes on to show that the steady state flow stress can be written as
\[ \sigma_e \approx \sigma^* + K_1 d_g^{-1} + K_2 d_g^{-1/2} \]  \hspace{1cm} (3.36)

Here \( \sigma^* \) is the effective stress and can be approximated to the yield stress.

### 3.12 Conclusions

This chapter discussed different approaches to modeling microstructure evolution in thermomechanical processing. Although the details of the models differ from one research group to another, they all represent a similar approach that involves characterizing microstructural changes due to recrystallization and grain growth. The models presented in this chapter were semi-empirical models. Researchers have used other techniques such as the Monte Carlo method (Anderson et al., 1983; Rollett et al., 1990) and pure regression (Burnett, 1986; Vedhanayagam, 1993). The true essence of the physical phenomena is best grasped by the physical metallurgy based empirical models and is considered most suited for modeling microstructural changes using FEM under a wide range of processing conditions (Shen, 1994).
4.1 Introduction

In spite of the long history of development in rolling, difficulties in theoretically understanding and describing the rolling process still exist. The reason for this is that the metal flow or the deformation mechanics in rolling is of a very complex nature and is affected by many variables simultaneously. Many of these variables such as friction, heat transfer coefficients, etc. are very hard to determine accurately and do not stay constant during the process. As a result, simple analytical tools or methods often find limited success when applied to hot rolling.

Many empirical formulas and theories have been developed to explain the physical phenomena occurring during rolling. Some of these methods are:

- spread formulas
- slip line field analysis
- slab method
The first four methods have until now been applied successfully only to simple processes. However, for a complex process like shape rolling, the applicability of these processes becomes severely restricted. In recent years, the development and application of the finite element method (FEM) has provided a new powerful tool for engineering analysis. FEM programs use incremental and numerical techniques for solving non-linear equations and hence require large amounts of computations. With significant advances being made in the development of fast computers, computation time is no longer a major constraint. FEM programs have the advantage over other analytical methods such as slip-line, upper bound, etc. in their ability to generate much more detailed information. The main advantage of FEM is that it can solve a wide range of problems with little restriction on workpiece geometry and boundary conditions. This makes it a very useful tool for analyzing metal forming processes.

4.2 Overview of past studies using FEM

Over the last decade, FEM has become a widely accepted tool for modeling many metal forming processes including rolling. A great deal of effort has been dedicated to developing FEM based tools for analyzing various metal forming processes. The deformation involved in a shape rolling process is complex and three dimensional. Spread is an important issue in shape rolling affecting dimensions of the rolled product.
A tool for the accurate prediction of spread would go a long way in helping design a roll pass to obtain the correct dimensions. A plane strain approach (2D FEM) is not capable of predicting spread in rolling. That makes 3D FEM the only choice for modeling the shape rolling process.

The rigid-plastic FEM is generally considered as an effective method for the analysis of metal forming processes. Many researchers in the past have used rigid-viscoplastic formulation for modeling hot shape rolling (Chenot (1989), Mori and Osakada (1989), Kiuchi (1988, 1990), Park (1990), etc.). Compared to the elastic-viscoplastic material, the rigid-viscoplastic material simplifies the solution process and requires considerably less computation time (Park 1990). Several different formulations based on the finite element method have been developed for modeling rolling. The principal difference between them stems from the Lagrangian or Eulerian frame of reference employed. The Lagrangian formulation is based on material coordinates while the Eulerian formulation (also referred to as steady state formulation) is based on a coordinate system fixed in space. In a steady state approach, solution is obtained only for the control volume encompassing the deformation zone. Such an approach has also been applied to modeling of extrusion with streamlined dies (Kiuchi 1995). A limitation of the steady state approach is that it cannot simulate the transient deformation at the head and tail ends of the rolled products. The simulation of non steady state rolling is desirable for improving rolling yield by minimizing these undesirable transients (Park, 1990). In the steady state scheme, the stream lines and the flow stress distributions are iteratively updated until a convergence is obtained. The computation time for the steady state
scheme is shorter than that for the non-steady state scheme (Lagrangian formulation) in which the simulation is repeated from an initial state for the front end until a steady state is attained. Non steady state simulation follows the deformation path incrementally over several time steps. Kiuchi and Yanagimoto (1988, 1990) introduced a new approach to modeling the rolling. Their method called the complex element method is a combination of the rigid-plastic FEM and the slab method. A variation of this approach was adopted by Kim et al (1991). An overview of the application of the finite element analysis to hot rolling was published recently by Montmitonnet et al (1991).

4.3 Development of ROLPAS

The objective of this study is to model microstructural evolution using the thermomechanical history of the process. FEM modeling of the deformation process provides a convenient and accurate way of obtaining the thermomechanical history of different material points in the workpiece at different instants in time. Development of the FEM program for simulating the rolling process was therefore the first step in this research. In this study an Eulerian formulation was used to develop the finite element model. Early work on developing the finite element model was carried out by Kim (1992). The model was first developed as a 2 ½-D FEM model using an approach similar to that of Kiuchi (1990). During the course of this study, it was changed to a full 3-D FEM model and renamed to ROLPAS.
4.3.1 Features of ROLPAS

The program ROLPAS (developed from TASKS) is a finite element code, capable of analyzing three-dimensional deformation of multi-pass rolling process. Some of the features that make it a very versatile tool for simulating metal flow and microstructural evolution in shape rolling are;

- Capability of performing heat transfer analysis in the stand as well as in the interstand region.
- 3D-FEM formulation enables it to predict spread which is an important issue in shape rolling
- Less computation time: Multi-stand roll pass sequences can be analyzed in a much shorter time as compared to commercially available programs such as DEFORM, ABAQUS, FORGE 3D, etc.
- Ability to handle arbitrary shaped roll profiles.

4.3.2 Theory of Deformation Analysis

ROLPAS is based on a rigid-viscoplastic formulation. Compared to the elastic-plastic material, the rigid-viscoplastic material simplifies the solution process requiring less computational time by neglecting the elastic behavior of the workpiece. Since deformation encountered in metal forming operations such as rolling is much larger than the yield point of the material, the error expected due to this approximation is very small. In the formulation of ROLPAS, the deformation condition within the roll-bite is assumed to be ‘kinematically steady’. This means that only the steady state condition is solved for.
ROLPAS uses a control volume approach in modeling the hot rolling process. Figure 4.1 illustrates the control volumes at different stages in the rolling process. Deformation and heat transfer is analyzed in the roll bites while only heat transfer is analyzed in the interstand region. For modeling the material flow (deformation), an initial guess shape is assumed and iterations are carried out to arrive at a final shape that satisfies the steady state conditions.

Figure 4.1 Control volume approach

Figure 4.2 Boundary value problem
Any metal forming problem can be considered as a boundary value problem in which there is a body of volume $V$, bounded by a surface $S$ as shown in figure 4.2. Assume the boundary surface of the body defined by $S$, in a part of which $S_v$, a velocity boundary condition $v_i^*$ is imposed and in the rest of which $S_f$ traction loads $t^*$ are specified.

Compatibility and equilibrium conditions within the volume need to be satisfied in addition to boundary conditions in accordance with flow rules and yield criteria. The equilibrium equations are written as:

$$\sigma_{ij,j} = 0 \text{ in } V$$  \hspace{1cm} (4.1)

where $\sigma_{ij}$ is the stress tensor.

The boundary conditions specified are:

$$v_i = v_i^* \text{ in } S_v$$  \hspace{1cm} (4.2)

$$t_i = t_i^* \text{ in } S_f$$  \hspace{1cm} (4.3)

Equations (4.1), (4.2) and (4.3) together define a boundary value problem.

The compatibility equations relate the strain rate to velocities

$$\dot{\varepsilon}_v = \frac{1}{2}(v_{i,j} + v_{j,i})$$  \hspace{1cm} (4.4)

Multiplying equation 4.1 by an arbitrary variation of the velocity field $\delta v_i$ and integrating over the volume this becomes equivalent to:

$$\int_v \sigma_{ij,j} \delta v_i dV = 0$$  \hspace{1cm} (4.5)

Integrating by parts and using the divergence theorem we get
\[
\int \sigma_{ij} \delta v_{i,j} \, dV - \int \sigma_{ij} n_j \delta v_i \, dS = 0
\]  \hspace{1cm} (4.6)

It may be noted that \( \sigma_{ij} n_j \) is the same as traction \( \tau^* \), with \( n_j \) being the components of the surface normal unit vector.

According to the Von Mises yield criterion, yielding will occur when

\[
\bar{\sigma} = \sqrt{\frac{3}{2} \sigma_{ij} \sigma_{ij}'}
\]  \hspace{1cm} (4.7)

where \( \bar{\sigma} \) is the flow stress of the material and \( \sigma_{ij}' \)'s are the deviatoric stresses. Using plasticity theory with the usual assumptions of isotropy, incompressibility along with the Von Mises yield criterion, the following flow rule is obtained:

\[
\dot{\varepsilon}_y = \frac{3}{2} \frac{\dot{\varepsilon}}{\bar{\sigma}} \sigma_{ij}'
\]  \hspace{1cm} (4.8)

where 
\[
\dot{\varepsilon} = \left( \frac{2}{3} \dot{\varepsilon}_y \dot{\varepsilon}_y \right)^{1/2}, \quad \bar{\sigma} = \left( \frac{3}{2} \sigma_{ij}' \sigma_{ij}' \right)^{1/2}
\]  \hspace{1cm} and \( \sigma_{ij}' \) represents the deviatoric stress and can be written as

\[
\sigma_{ij}' = \sigma_{ij} - \delta_{ij} \sigma_m
\]  \hspace{1cm} (4.9)

Here, \( \delta_{ij} \) is the Kronecker delta, and \( \sigma_m \) is the mean stress represented as

\[
\sigma_m = \frac{1}{3} \sigma_{kk}
\]  \hspace{1cm} (4.10)

It may be noted that equation 4.8 relates the strain rates to deviatoric stresses and therefore implicitly states the incompressibility, i.e.;

\[
\dot{\varepsilon}_{ii} = 0 \]

\hspace{1cm} (4.11)
However, for it to substitute into equation (4.6), the kinematically admissible fields must be restricted to ones that satisfy incompressibility. Such restrictions can be imposed mathematically by means of a Lagrange multiplier or by means of a penalty term in equation 4.6. In ROLPAS, the penalty term approach is used. Penalty formulation changes a constrained minimization problem to an unconstrained minimization problem. In this case, equation 4.6 becomes

\[ \int \frac{2}{3} \sigma_{ij} \dot{\varepsilon}_{ij} dV + k \int \dot{\varepsilon}_{ii} \delta \dot{\varepsilon}_{ii} dV - \int \delta v_i dS = 0 \tag{4.12} \]

where \( k \) is an arbitrarily large constant such that \( k \dot{\varepsilon}_{ii} = \sigma_m \). In other words, for a given \( \sigma_m \), the volumetric strain rate \( \dot{\varepsilon}_{ii} \) is inversely proportional to the penalty constant \( k \). Therefore, by making \( k \) large, the incompressibility condition can be satisfied. This formulation assumes that hydrostatic component of stress does not cause plastic deformation. This result has been verified experimentally by Bridgeman and other researchers.

### 4.3.3 Finite Element Discretization Procedure

Velocity is usually the field variable in rigid viscoplastic boundary value problems. Velocity \( \mathbf{v} \) at any location in an 8-node isoparametric hexahedral element can be obtained from the nodal velocities \( \mathbf{v} \) and the corresponding matrix of shape functions \( N \) as follows:
\[
\mathbf{v} = \begin{bmatrix} v_1 \\ v_2 \\ v_3 \end{bmatrix} = N\mathbf{\tilde{v}}
\]

(4.13)

Since the highest derivatives in equation (4.12) are of order one, continuity and compatibility conditions require \( \mathbf{v} \) to be locally of class \( C^1 \) and globally of class \( C^0 \). By substituting the interpolated approximate velocity given by equation (4.13) into the equilibrium equation or integral form of the boundary value problem in equation (4.13), an error is introduced. This error can be minimized by the use of Galerkin's method in which the shape functions in the matrix \( N \) are used as test functions.

The variation of the velocity vector can be written as

\[
\delta \mathbf{v} = \begin{bmatrix} \delta v_1 \\ \delta v_2 \\ \delta v_3 \end{bmatrix} = N\delta \mathbf{\tilde{v}}
\]

(4.14)

Noting that the strain rate tensor is symmetric, the number of components can be reduced by representing strain as a \( 6 \times 1 \) vector instead of a \( 3 \times 3 \) matrix as shown below.
Here, $A$ is a matrix of the differential operator and the matrix $B$ contains appropriate spatial derivatives of the shape functions. To incorporate the incompressibility condition, a constant vector $c$ is defined which gives the volumetric strain rate when multiplied with the strain rate vector.

$$\dot{\epsilon} = \begin{bmatrix} \dot{\epsilon}_{11} \\ \dot{\epsilon}_{22} \\ \dot{\epsilon}_{33} \\ \dot{\epsilon}_{12} \\ \dot{\epsilon}_{23} \\ \dot{\epsilon}_{31} \end{bmatrix} = \begin{bmatrix} \frac{\partial v_1}{\partial x_1} \\ \frac{\partial v_2}{\partial x_2} \\ \frac{\partial v_3}{\partial x_3} \\ \frac{1}{2} \left( \frac{\partial v_1}{\partial x_2} + \frac{\partial v_2}{\partial x_1} \right) \\ \frac{1}{2} \left( \frac{\partial v_2}{\partial x_3} + \frac{\partial v_3}{\partial x_2} \right) \\ \frac{1}{2} \left( \frac{\partial v_3}{\partial x_1} + \frac{\partial v_1}{\partial x_3} \right) \end{bmatrix} = A\nu = AN\ddot{v} = B\ddot{v}$$ (4.15)

$$\dot{\epsilon}_{kk} = \dot{\epsilon}_x + \dot{\epsilon}_y + \dot{\epsilon}_z = c^T \dot{\epsilon}$$ (4.16)

$$c^T = (1,1,1,0,0,0)$$ (4.17)

To compensate for the difference between the multiplication of strain rate matrices or strain rate tensors and cross product of strain rate vectors a constant matrix $D$ is introduced as shown below.

$$\dot{\epsilon}_{ij} \dot{\epsilon}_{ij} = \dot{\epsilon}^T D \dot{\epsilon}$$ (4.18)

where,
Substitution of the above equations (4.10) through (4.14) into (4.9) yields an integral equation in the matrix-vector form instead of tensor notation, i.e.,

\[ Y(v) = D B v \text{d}V + k (B^T c c^T B \text{d}V - \int_{s} N^T t^* dS) = 0 \]  

(4.20)

In equation 4.20, the first and second integrals represent the energy dissipation by shear and dilational deformation respectively, and the third integral represents the traction boundary condition. Equation 4.21 represents a set of nonlinear scalar elemental equations in discretized form. The continuum (control volume) is discretized using 8 node hexahedral elements.

\[ \Psi(v) = \sum_{\xi} \left\{ \int_{\xi} \frac{2}{3} B^T D B \text{d}V + k \int_{\xi} B^T c c^T B \text{d}V - \int_{s_{\xi}} N^T t^* dS \right\} = 0 \]  

(4.21)

In the FEM model, volume integrations are carried out using Gaussian quadrature formulas. For evaluation of the deformation energy term, 2x2x2 integration points are used while only one point is used for the dilational term so as to avoid the over-constraint caused by high order integration. Likewise, for evaluating the surface integral (last terms in equation 4.21), 5x5 integration points are used. Equation 4.21 can be expressed in a more concise form as follows:
\[ \psi(\ddot{\upsilon}) = \sum \left( F^e \ddot{v}^e + G^e \ddot{v}^e - h^e \right) = 0 \quad (4.22) \]

where

\[ F^e = \int_\Omega \frac{2}{3} \frac{\overline{\sigma}}{\overline{e}} B^T DBdV \]

\[ G^e = \int_\Omega k B^T c c^T BdV \]

\[ h^e = \int_{S_T} N^T r^T dS \]

### 4.3.4 Handling of Interface friction

The modeling of frictional forces is traditionally made in two ways: by means of a friction coefficient and by means of a friction factor. A directly applied friction coefficient leads to nonsymmetric matrices and is not commonly used in metal forming. A friction factor defined by the equation

\[ \tau = -mk \frac{v_g}{|v_h|} \quad (4.23) \]

where, \( \tau \) = shear friction stress

\( k_y \) = shear yield stress of the material

\( m \) = friction factor

\( v_h \) is the relative velocity between workpiece and die

A major problem faced with the above model is when a point of reverse relative velocity between the workpiece and die exists and its position is not known beforehand. This happens in rolling. During the initial part of deformation under the rolls, the roll
moves faster than the material in contact with it. However, after the neutral point, the workpiece material begins to move faster than the roll surface causing the orientation of the friction force to change at that point. So an abrupt jump (like a step function) in the value shows up rendering equation 4.21 very unstable. Chen and Kobayashi (1978) have suggested the following model which uses a tangent function to substitute for the step function.

\[ f = -mk_y \left( \frac{2}{\pi} \tan^{-1} \left( \frac{|v_R|}{a} \right) \right) \frac{v_R}{|v_R|} \]  

(4.24)

where, \( m \) is the friction factor

\( k_y \) is the shear yield stress

\( v_R \) is the magnitude of relative velocity between die and workpiece

\( a \) is a constant several orders of magnitude less than the die velocity (~ 10^{-5})

The relative velocity is approximated by

\[ v_R = N(\tilde{v} - \tilde{v}_{de}) \]  

(4.25)

where, \( \tilde{v}_{de} \) is the roll velocity at the material point and \( N \) is the matrix of shape functions.

The last term in equation 4.21 therefore becomes

\[ h(\tilde{v}) = \int_{S'}^{S'} N^T \tilde{t} \cdot dS = \int_{S'}^{S'} -mk_y \left( \frac{2}{\pi} \tan^{-1} \left( \frac{|v_R|}{a} \right) \right) \frac{N^T}{v_R} v_R dS \]  

(4.26)
4.3.5 Solution techniques

Equation 4.21 represents a set of equations that need to be solved for the nodal velocities. Because of the non-linear nature of the equations, an iterative approach has to be resorted to. Two iteration techniques are used in solving this set of equations:

- The direct iterative method
- Newton-Raphson iterative method

Direct iterative method

Direct iterative method is used to find an initial guess for subsequent iterations using the Newton-Raphson iterations. From equation 4.22 we have

\[
\left\{ F(\tilde{v}) + G \right\} \tilde{v} = h
\]  

(4.27)

This can be written as

\[
K_{dir} \tilde{v} = L_{dir}
\]  

(4.28)

where, \( K_{dir} \) is the stiffness matrix and \( L_{dir} \) is the load matrix. Since, \( K_{dir} \) is a nonlinear function of \( \tilde{v} \), an iterative approach is used to solve the set of equations as follows:

\[
K'_{dir} \tilde{v}'^{i+1} = L_{dir}
\]

(4.29)

where the superscript ‘\( i \)’ denotes the iteration number. This method gives a reasonable initial guess for the Newton-Raphson iterations. The iterations are terminated if the solution of the linearized equation satisfies

\[
\frac{\|v'^{i+1} - v'\|}{\|v'\|} \leq \varepsilon_{tolerance}
\]

(4.30)

where \( \varepsilon_{tolerance} \) is the error tolerance for fractional norm.
Newton Raphson iterative method

After an approximate initial guess is obtained, the Newton Raphson technique is used to obtain convergence. If \( v_{\text{guess}} \) is the initial guess solution and \( v_{\text{exact}} \) is the exact solution, then

\[
\psi(\hat{v}_{\text{exact}}) = \psi(\hat{v}_{\text{guess}} + \Delta \hat{v})
\]

\[
= \psi(\hat{v}_{\text{guess}}) + \frac{\partial \psi}{\partial \hat{v}} \bigg|_{\hat{v}_{\text{guess}}} \Delta \hat{v} + O(\Delta \hat{v}^2)
\]

\[
= \psi(\hat{v}_{\text{guess}}) + \frac{\partial \psi}{\partial \hat{v}} \bigg|_{\hat{v}_{\text{guess}}} \Delta \hat{v} = 0
\]  

(4.31)

The above linearized equation is solved iteratively as follows:

\[
\psi(\hat{v}^i) = -\frac{\partial \psi}{\partial \hat{v}} \bigg|_{\hat{v}_{\text{guess}}} \Delta \hat{v} = 0
\]  

(4.32)

This set of equations is solved for \( \Delta \hat{v} \) which is then used to update the velocity vector for the next iteration, i.e.:

\[
\hat{v}^{i+1} = \hat{v}^i + \alpha \Delta \hat{v}
\]  

(4.33)

where \( \alpha \) is called the acceleration coefficient and has a small value (~0.1) at the beginning of the iteration process and is gradually increased to unity as the iteration progresses. Equation 4.31 can be written in the following form:

\[
K'_{N-R} \Delta \hat{v} = L'_{N-R}
\]  

(4.34)

where,

\[
K'_{N-R} = \frac{\partial \psi}{\partial \hat{v}} \bigg| = F(\hat{v}) + G + \frac{\partial F(\hat{v})}{\partial \hat{v}} \hat{v} - \frac{\partial \mathcal{H}(\hat{v})}{\partial \hat{v}} \bigg|'
\]  

(4.35)
and

\[ L'_{N-R} = -\psi(\hat{v}') - F(\hat{v}')\hat{v}' - G\hat{v}' + h(\hat{v}') \]  \hspace{1cm} (4.36)

Here,

\[ \frac{\partial F(\hat{v})}{\partial \hat{v}} \hat{v} = \sum_{e} \int_{\varepsilon} \left( \frac{4}{\varepsilon^2} \left( \frac{\sigma}{\varepsilon} - \frac{\overline{\sigma}}{\varepsilon} \right) \right) \left( B^T DB\hat{v} \right) \left( B^T DB\hat{v} \right)^T dV \]  \hspace{1cm} (4.37)

The iterations are stopped if the following convergence criteria are satisfied:

\[ \left\| \nu'^{n+1} - \nu' \right\| \leq \varepsilon_{tolerance} \]  \hspace{1cm} (4.38)

and

\[ \left\| \psi(\hat{v}') \right\| \leq \psi_{tolerance} \]  \hspace{1cm} (4.39)

where \( \psi_{tolerance} \) is the error tolerance for functional norm.

4.3.5 Deformation Analysis Procedure

For a given workpiece cross section geometry (2-D mesh in the cross-section of incoming bar), the program ROLPAS automatically generates the 3-D finite element control volume in the roll gap using eight-node hexahedral isoparametric elements. The concept of control volume is illustrated in figure 4.3. After the initial control volume is established, traction and velocity boundary conditions are assigned to the nodes after determining the geometrical relationship between workpiece and rolls. The program then solves a set of simultaneous nonlinear equations under the assigned boundary conditions. In order to find the kinematically steady state geometry, ROLPAS updates the geometry
of workpiece after every iteration in the solution process based on the velocity field obtained. Iterations are continued until convergence of solution occurs.

![Initial guess and converged solution](image)

Figure 4.3 Initial guess and converged solution

### 4.3.6 Heat Transfer Analysis Procedure

During rolling, the bar at an elevated temperature undergoes temperature changes continuously during the process. The heat loss at the workpiece surface is due to convection, radiation and conduction. On the other hand, heat is generated in the workpiece due to the plastic deformation and due to friction at the roll-workpiece interface.

In steady state hot rolling, the longitudinal temperature gradients in the workpiece are small compared to transverse thermal gradients. Therefore, the heat flow along the rolling direction is neglected. This assumption allows us to model heat transfer using 2D FEM instead of 3D FEM formulation. In ROLPAS, the temperature and heat transfer rate
are found in the cross-section perpendicular to the rolling direction by the use of a 2-D finite element heat transfer analysis procedure.

**Governing equation**

The transient heat transfer problem is governed by the following equation:

$$\frac{\partial}{\partial t}\left(-\frac{\partial q_x}{\partial x} - \frac{\partial q_y}{\partial y}\right) + \dot{Q} = \rho c \frac{\partial T}{\partial t} \quad (4.40)$$

where,

- $q_x$ and $q_y$ are the components of the heat flow rate vector per unit area
- $\dot{Q}$ is the internal heat generation rate per unit volume $= \bar{\sigma} \dot{\varepsilon}$
- $\rho$ is the material density
- $c$ is the specific heat of the material

**Initial condition**

$$T(x, y, 0) = T_0(x, y) \quad (4.41)$$

**Boundary conditions:**

$$q_x n_x + q_y n_y = h_{int} (T_i - T_r) + \dot{Q}_{friction} \quad \text{on } S_1 \quad (4.42)$$

$$q_x n_x + q_y n_y = h_e (T_s - T_e) \quad \text{on } S_2 \quad (4.43)$$

$$q_x n_x + q_y n_y = \sigma \varepsilon (T_s^4 - T_e^4) \quad \text{on } S_3 \quad (4.44)$$

where,

- $n_x$ and $n_y$ denote the direction cosines of the outward normal to the surface
- $h_{int}$ is the interface heat transfer coefficient
\( h_c \) is the convection heat transfer coefficient

\( \varepsilon \) is the emissivity of the material

\( \sigma \) is the Stefan-Bolzman constant

\( T_s \) is the surface temperature

\( T_r \) is the roller temperature

\( T_e \) is the temperature of the environment

\( \dot{Q}_{\text{fric}} \) is the heat generated due to friction at the tool workpiece interface

**Finite element formulation**

In the finite element formulation for heat transfer analysis, the solution domain \( V \) is divided into \( M \) elements of 4 nodes each. The elements used are isoparametric 4 node quadrilateral elements. The temperature and temperature gradients within each element are expressed as:

\[
T^e(x,y,t) = \sum_{i=1}^{4} N_i(x,y)T_i(t)
\]

or in matrix notations,

\[
\{T^e\} = [N]\{T(t)\}
\]

\[
\begin{bmatrix}
\frac{\partial T}{\partial x} \\
\frac{\partial T}{\partial y}
\end{bmatrix} = [B]\{T(t)\}
\]

where,

\( \{T(t)\} \) is the vector of elemental nodal temperatures

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\( N \) is the temperature interpolation matrix

\([B]\) is the temperature gradient interpolation matrix

using the method of weighted residuals and by further simplification after introducing boundary conditions we get,

\[
\int_{\Omega} \rho c \frac{\partial T}{\partial t} \mathbf{N} \cdot d\mathbf{V} - \int_{\Omega} \left[ \frac{\partial \mathbf{N}_x}{\partial x} \frac{\partial \mathbf{N}_y}{\partial y} \right] \begin{bmatrix} q_x \\ q_y \end{bmatrix} d\mathbf{V} - \int_{\partial \Omega} \left( T - T_r \right) \frac{\partial \mathbf{N}_x}{\partial x} d\mathbf{S} + \int_{\partial \Omega} \mathbf{h}_{\text{int}} \left( T - T_r \right) \mathbf{N} \cdot d\mathbf{S} = 0 \quad i=1,2,3,4
\]

(4.48)

For convenience we write,

\[
\begin{bmatrix} q_x \\ q_y \end{bmatrix} = \begin{bmatrix} k_{11} & k_{12} \\ k_{21} & k_{22} \end{bmatrix} \begin{bmatrix} \frac{\partial T}{\partial x} \\ \frac{\partial T}{\partial y} \end{bmatrix} = -[k \mathbf{B}]T
\]

(4.49)

After some manipulation, the following element equation is obtained

\[
\begin{bmatrix} C \end{bmatrix} \begin{bmatrix} \frac{dT}{dt} \end{bmatrix} + [K_e] + [K_{h_m}] + [K_{h_c}] + [K_r] = \{R_{h_m}\} + \{R_{h_c}\} + \{R_r\} + \{R_c\} + \{s\}
\]

(4.50)

where

\[
[C] = \int_{\Omega} \rho c \mathbf{N} \cdot \mathbf{N} d\mathbf{V} \quad \text{(heat capacity matrix)}
\]

(4.51)

\[
[K_e] = \int_{\Omega} [B]^T [k \mathbf{B}] d\mathbf{V} \quad \text{the conductivity matrix}
\]

(4.52)

\[
[K_{h_m}] = \int_{\partial \Omega} h_{\text{int}} \mathbf{N} \cdot \mathbf{N} d\mathbf{S}
\]

(4.53)

\[
[K_{h_c}] = \int_{\partial \Omega} h_c \mathbf{N} \cdot \mathbf{N} d\mathbf{S}
\]

(4.54)
\[ [K_r][T] = \int_{S_i} \sigma e T^4 \{N\} dS \quad (4.55) \]

\[ [R_{\text{in}}] = \int_{S_i} h_{\text{in}} T_e \{N\} dS \quad (4.56) \]

\[ [R_{\text{con}}] = \int_{S_i} h_{\text{con}} T_e \{N\} dS \quad (4.57) \]

\[ [R_T] = \int_{S_i} \sigma e T_e^4 \{N\} dS \quad (4.58) \]

\[ [R_f] = \int_{S_i} \dot{Q}_{\text{friction}} \{N\} dS \quad (4.59) \]

\[ [R_o] = \int_{\Omega} \dot{Q} \{N\} dV \quad (4.60) \]

Schematic drawing of steady state analysis sequence is illustrated in figure 4.3.

Time integration scheme

An implicit one-step method is used in the time integration in equation 4.50. Consider two temperature states; \( T_i \) and \( T_{i-\Delta t} \) such that:

\[ T_i = T_{i-\Delta t} + \Delta t \cdot \dot{T} \quad (4.61) \]

The integration scheme is based on the generalized trapezoidal rule;

\[ T_i = T_{i-\Delta t} + \Delta t \left\{ (1 - \beta) \dot{T}_{i-\Delta t} + \beta \dot{T}_i \right\} \quad (4.62) \]

Rearranging the terms in the above equation, the temperature change rate can be determined as:

\[ \dot{T}_i = \frac{1}{\beta \Delta t} T_i - \frac{1}{\beta \Delta t} T_{i-\Delta t} - \left( \frac{1 - \beta}{\beta} \right) \dot{T}_{i-\Delta t} \quad (4.63) \]

Substituting equation 4.61 into equation 4.50 we get:
\[
\begin{bmatrix}
[K] + \frac{1}{\beta \Delta t} [C]
\end{bmatrix} \begin{bmatrix} T_i \end{bmatrix} =
\begin{bmatrix}
\{R_{hm}\} + \{R_h\} + \{R_r\} + \{R_Q\} + \left[ C \left( \frac{1}{\beta \Delta t} \right) \right] \{T_{i-\Delta t}\} + \left[ C \left( \frac{1 - \beta}{\beta} \right) \right] \{\bar{T}_{i-\Delta t}\}
\end{bmatrix}
\] (4.64)

where; 
\[
[K] = \begin{bmatrix}
[K_c] + [K_{hm}] + [K_h]
\end{bmatrix}
\]

Equation 4.64 represents a set of non-linear equations which needs to be solved to obtain the nodal temperatures.

### 4.4 Features in ROLPAS

The following are some of the recent improvements in ROLPAS that have enhanced its capability to model microstructure evolution.

- Capability to specify temperature dependent material properties such as conductivity, specific heat and density
- Location dependent heat transfer coefficients: This feature is important for modeling processes such as interstand cooling
- Pressure dependent heat transfer coefficient for heat exchange with the rollers: Researchers have determined that the heat transfer coefficient increases almost linearly with pressure up to a certain limit [Samarasekhara, et al. 1991]. Modeling pressure dependent heat transfer coefficient enhances the accuracy of the simulation.
- Initial guess optimization routine: In steady state FEM, the final geometry of the deformed workpiece is unknown. Therefore, an initial guess shape for the final geometry is necessary. Iterations are carried out until a convergence to the final
shape is achieved. If the initial guess shape is close to the final deformed shape, less number of iterations would be required. Keeping this in mind, a routine was developed to obtain the initial guess shape close to the final shape. This has not only reduced the computation time, but has also made feasible the simulation of some severe deformations which was earlier not possible.

- Spring loaded rolls to model deflection of the rolls
- Capability to simulate quarter geometry: This feature can be used only on symmetric shapes and gives results an order of magnitude faster than the full 3-D finite element model.
- Integration of austenite evolution module (MICON): A module for modeling microstructure evolution was integrated into ROLPAS. Details on this module are presented in the chapter 7. One of the by products of modeling microstructure evolution has been an improvement in the roll force predictions and material spread predictions (chapter 7).
- Development of phase transformation module (AUSTRANS): A module for modeling post rolling phase transformation was developed. Details about this module are presented in chapter 7.
- Development of PC version of ROLPAS: In keeping with the growing popularity of PCs in the metal forming industry, a PC version of ROLPAS was developed. It was observed that, the program runs at a comparable speed on a 300 MHz Pentium as compared to an IBM RISC workstation.
4.5 Summary

This chapter presented the formulation of the 3-D finite element model for non-isothermal simulation of multi-pass rolling processes. Details of the microstructure evolution modules and the procedures for integrating them into FEM will be presented in the next chapter.
Figure 4.3 Flow chart of steady state FEM.
CHAPTER 5

DEVELOPMENT OF MICROSTRUCTURAL EVOLUTION MODELS FOR TMS80R

5.1 Introduction

The pioneering work by Sellars and Whiteman (1979) demonstrated that semi-empirical equations describing microstructural phenomena, such as grain growth and recrystallization kinetics can be used along with the thermomechanical history of the process obtained from a computer model to predict metallurgical changes during a hot rolling process. Over the past few years, several researchers have tried to use microstructural modeling to predict ensuing microstructure in the rolled product. The objective of this study is to develop microstructure evolution models for a microalloyed steel and establish procedures for integrating them into the FEM model. This chapter describes the procedure used in developing the microstructure evolution model and the details of the model developed.
5.2 Microalloyed steel under study

The microalloyed steel used in this study is a ferrite-pearlite steel designated as TMS80R. It has 0.14% vanadium as the microalloying element. Table 5.1 lists the chemical composition of TMS80R. This steel has been used in previous studies by Vedhanayagam (1993) and Deshpande (1995).

<table>
<thead>
<tr>
<th>Element</th>
<th>% weight</th>
</tr>
</thead>
<tbody>
<tr>
<td>C</td>
<td>0.33</td>
</tr>
<tr>
<td>Mn</td>
<td>1.44</td>
</tr>
<tr>
<td>P</td>
<td>0.017</td>
</tr>
<tr>
<td>S</td>
<td>0.021</td>
</tr>
<tr>
<td>Si</td>
<td>0.26</td>
</tr>
<tr>
<td>Cr</td>
<td>0.23</td>
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<tr>
<td>Ni</td>
<td>0.11</td>
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<tr>
<td>Mo</td>
<td>0.05</td>
</tr>
<tr>
<td>Cu</td>
<td>0.16</td>
</tr>
<tr>
<td>Al</td>
<td>0.024</td>
</tr>
<tr>
<td>N ppm</td>
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<td>V</td>
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</tr>
<tr>
<td>Ti</td>
<td>0.001</td>
</tr>
<tr>
<td>Sn</td>
<td>0.013</td>
</tr>
</tbody>
</table>

Table 5.1 Chemical composition of TMS80R microalloyed steel
Figure 5.1 Gleeble test setup (a) and close-up of the specimen (b) (Phadke, 1997)
5.3 The Gleeble thermomechanical test system

Physical simulation techniques attempt to replicate real-world processes on laboratory scale in such a way that the resultant data can be used to solve real-world problems. These techniques make it possible to simulate off-line, various processing schedules for a particular material and to accurately predict the microstructural and mechanical properties that will result in actual production environment.

To model microstructural evolution in a thermomechanical control process, it is necessary to develop the following models

- Model for grain growth kinetics
- Model for static recrystallization kinetics
- Model for metadynamic recrystallization kinetics
- Model for recrystallized grain size ($d_{rec}$)

To develop these models, hot compression and heating tests were conducted on a Gleeble 3500 thermomechanical testing machine. The Gleeble machine is a fully integrated digital close loop control thermomechanical testing system. Developed and manufactured by Dynamic Systems Inc. (formerly Duffers Scientific Inc.), Troy, NY, the Gleeble system has been successfully used for a variety of standard materials tests like hot/warm tensile and compression testing, dilatometry/phase transformation tests, creep and fatigue tests, etc. It is also capable of simulating a variety of processes like hot rolling, forging, extrusion, continuous casting, welding, etc.
Gleeble 3500 is a horizontal press with a load capacity of 10 tons. The Gleeble mechanical system that controls the deformation of the test specimen consists of a fully integrated hydraulic servo system. LVDT transducers, load cells, or non-contact laser extensometry provide feedback to ensure accurate execution and repeatability of the mechanical test program. The thermal system which consists of a direct resistance heating system can heat the specimen at high rates, or can hold it at steady-state equilibrium temperature. Thermocouples welded to the specimen or a non-contact infrared pyrometer provide signals for accurate feedback control of specimen temperatures.

Figure 5.1a shows a photograph of the Gleeble test setup. This entire test setup is located in a vacuum chamber to prevent oxidation of the heated specimen. A thermocouple is attached to the specimen for temperature feedback. The specimen is mounted between two jaws. The jaw assembly on the left hand side is movable and is connected to the ram by a coupler. The specimen is heated by means of resistance heating. Figure 5.1b shows a close-up of the specimen that is held between the tungsten carbide anvils. A thin foil of graphite is used between the anvils and the specimen for interface lubrication. At very high temperatures, an additional sheet of tantalum is needed to prevent diffusion of tungsten carbide from the anvils into the test specimen.

5.4 Specimen preparation

Cylindrical specimens measuring 10 mm in diameter and 15 mm in height (aspect ratio 1.5), were used for all the tests. These specimens were machined using wire-
EDM from a 2.75 inch hot rolled bar of the test material. The small size of the specimen minimizes thermal gradients in the specimen during the tests.

5.5 Grain growth tests

Modeling of grain growth is important because the initial grain size prior to deformation is one of the contributing factors to the recrystallization kinetics. It is therefore important to model accurately the austenite grain growth in the soaking pit prior to rolling as well as in the interstand region after recrystallization.

![Diagram](image)

Figure 5.2 Experimental procedure for the grain growth tests
5.5.1 Experimental procedure

The grain growth experiments were performed with austenitization temperature and hold time at that temperature as the control variables. The schematic of the experimental procedure is shown in figure 5.2. The experiment consisted of heating the specimen at the rate of 30 °C/sec. to the austenitization temperature (T °C) and holding the specimen at the temperature for a given amount of time (t sec.). The initial high heating rate was used to prevent aluminum nitride precipitation during heating. After holding the specimen at the test temperature, the specimen was cooled rapidly using compressed air to about 600 °C. The specimen was held there for a few seconds to allow formation of proeutectoid ferrite at the prior austenite grain boundaries. The specimen was then quenched to the room temperature.

The experiments were conducted for the following values of austenitization temperatures and holding times:

- **Austenitization temperature**
  - 1000, 1100, 1200 and 1250 °C

- **Holding time**
  - 1, 5, 10, 30, 180 sec.

These values were chosen to reflect mill conditions based on prior experience and preliminary experiments. A full factorial experiment was performed with three replications for each input combination. Thus a total of 60 tests were performed. The grain growth at the lower temperature (1000 °C) was found to be very small.

The grain size measurements on these test samples were performed using an image analysis system. Heyn’s linear intercept procedure was used to determine the
average grain size. ASTM guidelines (standard E-112) were followed. Table 5.2 shows the average grain sizes recorded in the experiments. The mean value shown is an average over four readings (4 fields examined per sample). The conventional grain growth law

\[ d^n = d_0^m + Kt \exp\left(\frac{Q}{RT}\right) \]  

(5.1)

was used to model the grain growth.

Here, \( d \) is the grain size at time \( t \) (in microns)

\( d_0 \) is the initial grain size (microns)

\( Q \) is the apparent activation energy (J/mole)

\( T \) is the absolute temperature (K)

\( m \) and \( K \) are material constants.

The measured exponent \( m \) is usually higher than the theoretically expected value of 2 as discussed in chapter 3. Earlier studies (Militzer, et al.(1996), Sellars (1979), etc.) have used values of \( m \) ranging from 2 to 13. In microalloyed steels, the effect of manganese segregation, presence of carbides and other precipitates cause the grain size exponent to be larger than 2.

The constants in the grain growth equation, namely; \( m \), \( Q \) and \( K \) were obtained by means of a regression analysis. An iterative approach was required because of the implicit non-linear nature of the equation. The regression analysis yielded the following values: \( m = 5 \), \( Q = 655826 \text{ J/mole} \quad 0^\circ \text{K} \) and \( K = 1.26 \times 10^{32} \). Thus the grain growth model for TMS80R can be written as:
\[ d^s = d_0^s + 1.26 \times 10^{32} t \cdot \exp \left( \frac{655826}{RT} \right) \] (5.2)

Figure 5.4 shows the experimental measurements and the best fit curve representing the grain growth equation. The correlation between the predicted and the measured grain sizes is seen to be fairly good.

The best fit for the grain size exponent \( m = 2 \) resulted in the following values of \( Q \) and \( K \). \( Q = 21500 \text{ J/mole } ^\circ \text{K} \) and \( K = 4.12 \times 10^9 \). However, the correlation coefficient was very poor (\( \sim 50\% \)). Failure of experimental data to conform to the parabolic growth law \( (m = 2) \) can be attributed to various factors such as pinning forces due to second phase particles, drag forces on grain boundaries, etc. While equation 5.2 gives fairly good correlation between measurements and predictions as compared to the parabolic law, it too does not explicitly model the pinning effect and drag forces and other phenomena that affect grain growth. Therefore it is not a mechanism based model in the true sense. The value of activation energy can be misleading in a model like this. Higher value of the exponent \( m \) typically results in a higher value of \( Q \). Therefore in materials such as microalloyed steels that exhibit pinning effect due to precipitates, \( m \) and consequently the \( Q \) value obtained by means of regression analysis will be higher than theoretical estimates. Since this type of model does not represent the mechanism in the true sense, its use must be limited to the range of the parameters used in developing the model.
(a) Grain size after 3 minutes at 1000°C (200X)
Figure 5.3 Micrographs showing grain growth at different temperatures
<table>
<thead>
<tr>
<th>Expt. #</th>
<th>Temperature °C</th>
<th>Hold Time sec.</th>
<th>Mean Grain Size (microns)</th>
</tr>
</thead>
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<tr>
<td></td>
<td></td>
<td></td>
<td>replication 1</td>
</tr>
<tr>
<td>1</td>
<td>1100</td>
<td>1</td>
<td>26.7</td>
</tr>
<tr>
<td>2</td>
<td>1100</td>
<td>5</td>
<td>28.2</td>
</tr>
<tr>
<td>3</td>
<td>1100</td>
<td>10</td>
<td>59.0</td>
</tr>
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<td>4</td>
<td>1100</td>
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</tr>
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<td>155</td>
</tr>
<tr>
<td>16</td>
<td>1250</td>
<td>180</td>
<td>241.6</td>
</tr>
</tbody>
</table>

Table 5.2 Grain growth modeling: Test matrix and measured grain sizes
5.5.2 Grain growth under non-isothermal conditions

The grain growth model discussed above was developed under isothermal conditions. On the other hand, the temperature changes constantly in a rolling process. Semiatin, et al. (1994) have developed closed form equations for applying the isothermal grain growth model under constant heating rate conditions. During cooling however, the temperature does not usually change at a constant rate. To apply this model under non-
isothermal conditions we use incremental numerical computation. In this procedure, we divide the time-temperature cooling (or heating) curve into several small time segments. In each segment, the temperature is assumed to be constant as illustrated in figure 5.5. If the initial grain size is $d_{01}$, the grain size at the end of first step $\Delta t_1$ is

$$d_{1}^{m} = d_{01}^{m} + K\Delta t_1 \exp\left(\frac{Q}{RT_1}\right)$$  \hspace{1cm} (5.3)

$d_1$ is then used as the initial grain size for the next time step $\Delta t_2$ and the grain growth during this time interval is modeled using the grain growth equation. This procedure is repeated for every subsequent step. Consequently, the grain size at the end of step $i$ is given by

$$d_{i}^{m} = d_{i-1}^{m} + K\Delta t_i \exp\left(\frac{Q}{RT_i}\right)$$  \hspace{1cm} (5.4)

To validate this procedure, experiments were conducted in which specimens were heated from 1100 °C to 1250 °C at two different heating rates (5°C/sec and 30°C/sec). The initial heating to 1100 °C from room temperature was carried out at a high heating rate (30°C/sec) to prevent precipitation of aluminum nitride in the test specimens. Grain size measurements were made at 1100, 1200 and 1250°C. Figure 5.6 shows the comparison between measured and predicted grain sizes for the three cases. The predictions are found to match well with the measurements thereby validating the procedure.
Figure 5.5 Approximating the continuous time-temperature curve by discrete isothermal steps.
Figure 5.6 Measurements vs. Predictions in non-isothermal grain growth tests
5.6 Modeling recrystallization

Recrystallization is the primary mechanism by which austenite grains get refined during hot working. The objective of the recrystallization tests is to obtain models for static and metadynamic recrystallization kinetics as a function of the processing conditions and the existing microstructure.

As discussed in chapter 3, recrystallization that starts and finishes after deformation is complete is termed as static recrystallization while, recrystallization that occurs during deformation is called dynamic recrystallization. A third type of recrystallization that starts during deformation (formation of recrystallization nuclei) and undergoes completion after deformation is referred to as metadynamic recrystallization. Under production conditions for bar rolling, dynamic recrystallization is difficult to achieve because the strains per pass are generally lower than the critical strain for the onset of dynamic recrystallization. Most of the microstructural changes in bar rolling are due to static recrystallization and metadynamic recrystallization. In this study, the occurrence of dynamic recrystallization will be modeled as metadynamic recrystallization.

A good measure to differentiate static from metadynamic recrystallization is the strain corresponding to the peak of the flow stress curve. The stress-strain curve under the conditions of temperature and strain-rates typically seen in hot rolling initially rises to a peak stress at a certain strain ($\varepsilon_n$). Thereafter, as a result of dynamic recrystallization, the material softens and the flow stress drops and stabilizes to a value between the yield
stress and the peak stress. The critical strain for the onset of dynamic recrystallization \( \varepsilon_c \) is reported to be somewhat lower than the peak strain \( \varepsilon_p \) (Rossard, 1973; Sellars, 1990) and depends on temperature, strain rate and the grain size. While it is very difficult to accurately pinpoint the strain for the start of dynamic recrystallization, researchers have found that the following relationship between critical strain and peak strain works very well for most steels.

\[
\varepsilon_c = 0.83 \varepsilon_p
\]  

(5.5)

Flow stress tests described in chapter 6 were conducted using strain rate, temperature and grain sizes as the control variables. Consequently, the following model for peak stress as a function of these parameters was developed.

\[
\varepsilon_p = 1.20 \times 10^{-3} d_0^{0.27} \left( \dot{\varepsilon} \exp\left( \frac{280000}{RT} \right) \right)^{0.18}
\]  

(5.6)

where \( d_0 \) is the as heated austenite grain size,

\( T \) is the absolute temperature

\( \dot{\varepsilon} \) is the strain rate

The form of this equation is based on the equations used by Sellars (1990), Samarasekara et al. (1996) and other researchers for steels.

**5.6.1 Double hit compression tests to model recrystallization kinetics**

Recalling from chapter 3, the recrystallization kinetics is characterized by the time for 50% recrystallization which is a function of the strain, strain rate, temperature and
grain size. To determine the time for 50% recrystallization, single hit deformation tests under hot working conditions followed by quenching to "lock" the microstructure can be performed (Shen, 1994). However, these single hit tests need to be followed by extensive metallography and measurements to determine the recrystallized fraction in each test specimen. On the other hand, double hit deformation test can give instantaneous estimates of the fraction recrystallized based on measurements on the stress-strain curve. Figure 5.7 shows a typical stress-strain curve during a double hit compression test. $\sigma_1$ and $\sigma_2$ are the yield stresses for the first and the second hit. The fraction softening, $F_s$, occurring in the interval between the two hits is given by:

$$F_s = \frac{\sigma_m - \sigma_2}{\sigma_m - \sigma_1}$$

(5.7)

where $\sigma_m$ is the stress at the end of the first hit as shown in figure 3.4. It is assumed that initial 20% softening takes place due to static recovery. The fraction recrystallized is therefore given by:

$$F_x = \frac{F_s - 0.2}{0.8}$$

(5.8)
Figure 5.7 Typical stress-strain curve in a double hit compression test

It is often argued that examination of microstructure is a more accurate way of measuring the fraction recrystallized and that the softening due to static recovery can vary between 15 to 30% (McQueen and Jonas, 1976). Therefore, the assumption of 20% initial softening due to recovery is not always valid. Few others state that under some conditions, the recrystallized grain size is very close to the initial grain size which makes it difficult to estimate accurately the recrystallized fraction using metallographic techniques. Under these conditions, it would be more accurate to use a double hit compression test. In this study, considering the time constraint and the enormous amount of testing that was planned, it was decided to use the double hit compression test to study recrystallization kinetics.
**Figure 5.8** Experimental procedure for a double hit compression test

**Test procedure**

The test procedure for the double hit compression test is shown in figure 5.8. Using the grain growth model, the heating temperature and the corresponding hold time necessary for obtaining the desired grain size was determined. The specimen was then heated to this temperature and held there until the desired grain size was obtained. The specimen was then cooled to the test temperature (T). After homogenizing at the test temperature, the specimen was compressed by a predetermined amount and held at the test temperature for a certain time (t) before being compressed again. The stress-strain data during these tests was recorded using a high speed data acquisition system available on the Gleeble 3500. The fraction recrystallized in each test was computed using the stress-strain data.
Figure 5.9 Stress strain curves in double hit compression tests conducted at 1100°C at a strain rate of 1.0/sec with an as heated grain size of 135 microns.
Figure 5.9 shows three of the stress-strain curves recorded during the experiments. These curves are for a test temperature of 1100 °C, strain rate of 1.0/sec. and a strain of 0.35 in the first hit and a strain of 0.20 in the second hit. Clearly, longer inter hit time leads to more recrystallization and consequently results in lowering of the yield stress during the second hit.

5.6.2 Model for Static Recrystallization

Double hit compression tests were conducted for modeling static recrystallization kinetics using the following control variables:

- strain rate: 0.1 / sec, 1.0 / sec and 10.0 /sec
- temperature: between 900 °C and 1100 °C
- grain size: 200 microns and 75 microns
- strain: between 0.15 and 0.4 and depending on other conditions
- inter hit time: three different inter hit times per test condition

Table 5.3 lists the test conditions and the corresponding fraction recrystallization measured in each of these tests. The test matrix was designed to cover the range of processing conditions (strain, strain rate, temperature and grain size) normally encountered in hot rolling and within limitations of the test machine capability.
<table>
<thead>
<tr>
<th>Test</th>
<th>strain</th>
<th>stn rate (per sec)</th>
<th>(d_0) (microns)</th>
<th>temp (°K)</th>
<th>inter hit time (sec)</th>
<th>Fraction Recrystallized</th>
</tr>
</thead>
<tbody>
<tr>
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<td>1</td>
<td>200</td>
<td>1000</td>
<td>2.000</td>
<td>0.3154</td>
</tr>
<tr>
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<td>200</td>
<td>1000</td>
<td>1.000</td>
<td>0.1549</td>
</tr>
<tr>
<td>ts1_5</td>
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<td>200</td>
<td>1000</td>
<td>5.000</td>
<td>0.6333</td>
</tr>
<tr>
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<td>75</td>
<td>1000</td>
<td>2.000</td>
<td>0.5137</td>
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</tr>
<tr>
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Table 5.3 Test matrix for modeling static recrystallization kinetics

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The measured fraction recrystallized data was used to fit an Avrami type sigmoidal curve.

\[ X = 1 - \exp \left( -0.693 \left( \frac{t}{t_{0.5}} \right)^n \right) \]  

(5.9)

where \( X \) is the fraction recrystallized at time \( t \)

\( t_{0.5} \) is the time for 50\% recrystallization

\( n \) is the time exponent and is assumed to be constant.

The value of \( n \) was determined from a least squares analysis on the fraction recrystallized measurements. Figure 5.10 shows a plot of \( \ln(\ln(1/(1-X))) \) vs \( \ln(t) \) for different test conditions. The slope of the lines is the value of \( n \). The average value of \( n \) obtained for the above data set is 1.46. This value is somewhat lower than the value of 2 used by Sellars and co-workers for C-Mn steels. However it is within the range of 1.50 and 0.7 reported by Medina and Mancilla (1996) for low alloy and microalloy steels. They also report that the value of \( n \) diminishes slightly with temperature. Matja, et al (1996) have reported values between 1.0 and 2.0 to be most satisfactory for microalloyed steels.

After determining the time exponent \( n \), the Avrami type equation was used to determine the time for 50\% static recrystallization \( (t_{0.5}) \) for the tests conducted. \( t_{0.5} \) can then be related to the as heated grain size, strain, strain rate and temperature using a regression analysis. The following relationship for the time for 50\% static recrystallization was obtained.
In(time)

Figure 5.10 Determination of time exponent in the Avrami type equation

\[ t_{0.5} = 1.73 \times 10^{-10} \varepsilon^{-1.78} \dot{\varepsilon}^{-0.433} d_0^{0.66} \exp\left(\frac{197000}{RT}\right) \]  

(5.10)

where, \( \varepsilon \) is the strain
\( \dot{\varepsilon} \) is the strain rate ( /sec)
\( d_0 \) is the as heated grain size (microns)
\( T \) is the absolute temperature (°K)

**Effect of grain size**

The nucleation of new recrystallized grains preferentially takes place on the grain boundaries. For this reason, the initial grain size will have an influence on the
recrystallization kinetics; the smaller the grain the faster the recrystallization rate (figure 5.11).

**Effect of temperature**

Recrystallization is a thermally activated phenomenon. The phenomenon which governs recrystallization is grain boundary self diffusion the rate of which increases with an increase in temperature. Consequently, higher temperatures are associated with faster recrystallization rates (figure 5.12).

**Effect of strain**

Larger the strain the higher is the dislocation density (until the peak strain). Higher dislocation density at larger strains results in faster recrystallization rates (figure 5.13).

**Effect of strain rate**

The driving force for recrystallization is the increase in dislocation density that accompanies plastic deformation. As the material is deformed dislocations get piled up throughout the grains. At the same time dynamic recovery processes result in annihilation of some of these dislocations. Under high strain rates, dislocations get piled up much faster whereas the annihilation rate due to dynamic recovery remains about the same. Consequently for deformation under high strain rates, the dislocation density for the same amount of deformation is higher as compared to deformation at a lower strain rate. This results in the material recrystallizing at a faster rate (figure 5.14).

**Value of apparent activation energy**

The value of apparent activation energy 197000 J/mole is lower than the values reported by other researchers for C-Mn steels (Sellars, Siwecki, Roberts, Yada, etc.). A similar
observation was reported by Madina and Mancilla (1996). The reason could be that the temperatures under which the experiments were carried out were higher than those used by other authors. Medina and Mancilla define a temperature called static recrystallization critical temperature (SCRT) below which microalloyed steels display a plateau in the recrystallization kinetics curve due to precipitation of carbides and nitrides. For temperatures higher than the SCRT, the slope of the logarithm of \( t_{0.5} \) against the inverse of temperature multiplied by the constant R which is in effect the value of \( Q \) is lower compared to its value at temperatures lower than the SCRT (figure 5.15). Any point corresponding to a temperature lower than SCRT would raise the slope and therefore the value of \( Q \), this being the case with some authors.

The influence of the content of each alloying content on the apparent activation energy was modeled by Medina and Mancilla in the following equation:

\[
Q(J/mol) = 124714 + 28385.7[Mn] + 64716.7[Si] + 72775.4[Mo] \\
+ 76830.3[Ti]^{0.213} + 12100.4[Nb]^{0.11}
\]  

(5.11)

Using this equation, the value of activation energy obtained for TMS80R is 203700 J/mol. This agrees quite well with the value obtained experimentally. It is interesting to note that equation 5.11 does not show any effect of vanadium on the value of activation energy although many of the steels studied by the authors did contain vanadium as a microalloying element. Also of interest is the fact that the independent term of equation 5.11 with a value of 124714 J/mol is similar to other values found in literature for grain boundary self diffusion energy of pure iron (Ratochka et al. 1996). The presence of alloying elements as expected serves to increase the apparent activation energy.
Figure 5.11 Effect of grain size on static recrystallization kinetics. (Test conditions: \( \dot{\varepsilon} = 1.0/\text{sec}, \varepsilon = 0.3, \text{temperature} = 1000 \, ^\circ\text{C} \) )
Figure 5.12 Effect of temperature on static recrystallization kinetics. (Test conditions: \( \dot{\varepsilon} = 0.1 \text{/sec}, \varepsilon = 0.15 \), grain size = 75 microns)
Figure 5.13 Effect of strain on static recrystallization kinetics. (Test conditions: $\dot{\varepsilon}=10.0/s$, grain size = 200 microns, temperature = 1100 °C)
Figure 5.14 Effect of strain rate on static recrystallization kinetics. (Test conditions: $\varepsilon = 0.2$, grain size = 75 microns, temperature = $1050 \, ^\circ C$)
Figure 5.15 Plot of $t_{0.5}$ against $1/T$ for a microalloyed steel (Medina and Mancilla, 1996)
5.6.3 Model for Metadynamic Recrystallization

Double hit compression tests for metadynamic recrystallization conditions were conducted with the following factors as control variables:

- **strain rate:** between 0.1 / sec and 10.0 / sec
- **temperature:** between 950 °C and 1200 °C
- **grain size:** 200 microns and 75 microns
- **strain:** between 0.30 and 0.50 and depending on other conditions
- **inter hit time:** three different inter hit times per test condition

Table 5.4 lists the test conditions and the corresponding fraction recrystallized measured in each of these tests.

The strain rate exponent in the Avrami type equation was found to be close to 1.0 for metadynamic recrystallization using a similar procedure as in section 5.6.2. This is in agreement with the Sellar's model and models used by other researchers (Samarasekhara, Roberts, Siwecki, etc.). The time for 50% recrystallization was related to strain, strain rate, temperature and grain size using a regression model.

\[
\tau_{0.5} = 5.78 \times 10^{-6} \varepsilon^{-100} d_0^{-0.15} Z_{app}^{-0.6} \exp \left( \frac{230000}{RT} \right) \quad (5.12)
\]

where, \(Z_{app} = \dot{\varepsilon} \exp \left( \frac{197000}{RT} \right)\) is the apparent Zener Hollomon parameter.

Comparing the exponents of various terms in equations 5.12 and 5.10, it can be seen that the effect of grain size and strain is significantly less in metadynamic recrystallization as compared to static recrystallization. The effects of grain size strain,
temperature, and strain rate are displayed in figures 5.16, 5.17, 5.18 and 5.19 respectively. It appears that static and metadynamic recrystallization follow two different kinetic laws. Static recrystallization depends on strain and only slightly on strain rate. Whereas, metadynamic recrystallization depends on strain rate and only slightly on strain. These observations are consistent with previous results on other microalloyed steels by Roucoules, et al. (1995) and others. The apparent activation energy for metadynamic recrystallization is of the same order of magnitude as those reported by Roucoules, et al. for Mo, Ti and Nb steels. As with other regression models discussed earlier, the form of the model used for characterizing recrystallization, does not explicitly model the effect of mechanisms such as precipitation of carbonitrides. As a result, the activation energy appearing in equation 5.12 may be looked at as apparent activation energy without much physical significance attached to it.
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Table 5.4 Test matrix for modeling metadynamic recrystallization kinetics
Figure 5.16 Effect of grain size on metadynamic recrystallization kinetics. (Test conditions: $\dot{\varepsilon} = 0.1$/sec, $\varepsilon = 0.4$, temperature = 1100 °C)
Figure 5.17 Effect of strain on metadynamic recrystallization kinetics. (Test conditions: \( \varepsilon = 0.1 \) sec, grain size = 200 microns, temperature = 1200 °C)
Figure 5.18 Effect of temperature on metadynamic recrystallization kinetics. (Test conditions: $\dot{\varepsilon} = 1.0/$sec, $\varepsilon = 0.35$, grain size $= 135$ microns)
Figure 5.19 Effect of strain rate on metadynamic recrystallization kinetics. (Test conditions: grain size = 75 microns, $\varepsilon = 0.4$, temperature = 1100 °C)
The size of the recrystallized grains is dependent on the strain, as heated grain size, strain rate and temperature. Experiments were performed with strain, strain rate, grain size and temperature as the control variables. To develop models for recrystallized grain size, the specimens were heated to obtain the desired grain size after which they were cooled to the test temperature and deformed to a certain strain at an appropriate strain rate as per the test matrix. After deformation, the specimens were held at that temperature to allow for complete recrystallization to take place. The time for complete recrystallization (95% recrystallization) was determined from the recrystallization kinetics models developed earlier. The specimens were then cooled rapidly to 600°C and held there for a short time to allow proeutectoid ferrite to form along the prior austenite grain boundaries before being quenched to room temperature. The microstructures in the specimens were later examined in an optical microscope to measure the grain size.
The following equations were developed to model recrystallized grain size ($d_{rec}$) for static and metadynamic recrystallization:

**Static Recrystallization:**

$$d_{rec} = 36.5 \cdot \varepsilon^{-0.341} \cdot \dot{\varepsilon}^{-0.06} \cdot d_0^{0.58} \exp\left(-\frac{3586}{T}\right)$$  \hspace{1cm} (5.13)

**Metadynamic recrystallization:**

$$d_{rec} = 53.41 \cdot \varepsilon^{-0.72} \cdot \dot{\varepsilon}^{-0.113} \cdot d_0^{0.39} \exp\left(-\frac{3544}{T}\right)$$  \hspace{1cm} (5.14)

The higher the temperature of deformation and the lower the strain rate, the larger are the recrystallized grains that are formed during high temperature deformation.

### 5.8 Recrystallization in a multiple stage deformation process: Issue of handling partial recrystallization

Rolling is a multiple stage deformation process. The primary mode of grain refinement is the static and metadynamic recrystallization that occurs during the period between two successive deformations. Often during the rolling process, the time in the interstand is not sufficient for complete recrystallization to occur. Under such circumstances, only partial recrystallization occurs resulting in a part of the microstructure consisting of fine strain free recrystallized grains while the other fraction consists of deformed grains which are usually relatively coarse and elongated as compared to the recrystallized grains. In other words, some amount of strain is retained in the structure when it enters the next deformation pass while at the same time, the
structure is non-uniform. This poses a very critical issue about how to handle the two fractions.

Several approaches have been proposed to handle partial recrystallization. One of the approaches is to treat the microstructure as an aggregate. In this approach, the retained strain and the effective grain sizes are determined using the rule of mixtures as follows:

\[ \varepsilon_{\text{ret}} = \varepsilon \cdot (1 - X) \]  \hspace{1cm} (5.15)

\[ d_{\text{eff}} = \left( X \cdot d_{\text{recr}}^3 + (1 - X) \cdot d_0^3 \right)^{1/3} \]  \hspace{1cm} (5.16)

where, \( X \) is the fraction recrystallized, \( \varepsilon_{\text{ret}} \) is the retained strain, \( d_{\text{eff}} \) is the effective grain size, \( d_0 \) is the initial as heated grain size and \( d_{\text{recr}} \) is the recrystallized grain size.

This approach has been used by several researchers (Saito and Shiga (1992), Samarasekhara, et al. (1996), etc.). It is easy to implement but gives rise to inherent errors in the modeling of recrystallization kinetics. This is because the grain size and strain which is used for calculating recrystallization kinetics in the subsequent pass is not representative of either of the fractions (recrystallized and unrecrystallized fractions) of the material. It is an average of fractions that are expected to behave differently (have different recrystallization kinetics). The error is however expected to be small if the fraction recrystallized is either close to zero or close to 1.0. The latter is usually the case in bar or shape rolling with reversing mills. A variation of the rule of mixtures was used by Jin, et. al (1996) following Hodgson and Gibbs (1992) who claimed that the following
model predicted the softening kinetics more accurately than the rule of mixtures for low carbon steels.

\[ \varepsilon_{\text{rev}} = 0.5 \cdot \varepsilon \cdot (1 - X) \]  

(5.17)

The other approach is to treat the recrystallized and unrecrystallized fractions independently (Karhausen et al, 1993). This allows close representation of the true behavior of the material. However, the number of fractions to be handled increases very rapidly. If three are m number of passes, the total number of fractions can be as high as \(2^m\). Typically in shape rolling, 10 to 15 passes are fairly common. This approach therefore calls for tremendous amount of computer memory and time and is therefore not recommended.

Recently, Yanagimoto (1997) proposed a variation of Karhausen’s model. In this approach, the number of fractions increases linearly as against exponentially in Karhausen’s model and therefore requires considerably less memory. However, as with the rule of mixtures, considerable approximation is involved the true behavior of the system is not represented.

In this study, the plan is to use the rule of mixtures or a variation of it that works best for the steel under study. Three hit compression tests were conducted to determine the validity of the rule of mixtures.

In the three hit compression tests, the first inter hit time was deliberately kept short to cause partial recrystallization. The second hit was followed by a third hit with an inter-hit time between the two. The amount of recrystallization in the second inter-hit
time was measured using the same procedure as was used in the double hit compression tests. Figure 5.21 shows an example from one of the three hit tests conducted. The deformation in each hit was equivalent to an increment in strain of 0.15. The noise in the data is because of the low strain rate used in the test.

Figure 5.21 Stress-strain data from three hit compression tests
In figure (a), the low temperature and the low strain in the first hit led to very little softening in the ten seconds of the first interstand. Consequently, the retained strain at the beginning of the second was large which got added to the strain in the second hit. The net result was that the accumulated strain in the specimen was larger (~ 0.25) which led to much faster recrystallization after the second hit. This is reflected in the significant softening that is seen in the stress-strain curve for the third hit. Note that the second inter-hit time is only 3 seconds while the first inter-hit time was 10 seconds.

In figure (b), the significantly larger duration of the interstand (30 seconds) led to a large amount of recrystallization in the first interstand. As a result, the retained strain at the beginning of the second hit was very small. After, the second hit, the net accumulated strain was much smaller than that in case (a). Consequently, the recrystallization kinetics after the second hit were significantly slower. This is reflected in very little softening that is seen in the stress-strain curve for the third hit. It may be noted that the second inter-hit time in this case is the same as that in the earlier case, i.e. 3 seconds.

Such three hit compression tests were conducted under various conditions of temperature, strain, strain rate and as heated grain size. The effective grain size after partial recrystallization in the first interstand was estimated using equation 5.14. This together with the retained strains obtained using the rule of mixtures (equation 5.13) and Hodgson and Gibbs rule (equation 5.15) were used to model recrystallization kinetics in the second interstand. Figures 5.22 (a) and (b) show how the predictions for the two rules compare with the measurements. It was found that the rule of mixtures showed a better
correlation with the measurements for TMS80R and will subsequently be used in the integrated model. There is however a large variance in the predictions. This arises due to the approximations we have made and also due to the fact that it is somewhat difficult to accurately determine the yield stresses for calculating the fraction recrystallized.

![Graph of rule of mixtures](a) rule of mixtures

![Graph of Hodgson and Gibbs](b) Hodgson and Gibbs

Figure 5.22 Recrystallization kinetics under partial recrystallization conditions

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5.9 Recrystallization kinetics under changing temperature conditions

Recrystallization like grain growth was developed under isothermal test conditions. Temperature changes continuously during hot rolling. Semiatin et al. (1996) have developed closed form equations to model recrystallization kinetics under constantly increasing temperature conditions. An equation of such a form for non-constant heating / cooling rate is not easy to obtain and therefore, an incremental computation approach needs to be adopted. In this study, it is assumed that the thermal properties used in the finite element model do not change significantly with microstructure. This assumption allows us to model heat transfer and microstructure evolution separately although they are modeled on the same time scale. The cooling
curve in the interstand region is divided into several time steps. In each of these time steps, the temperature is assumed to be constant. In other words, the cooling curve is approximated by a large number of isothermal steps. Figure 5.23 illustrates the technique that is discussed below. The figure shows two sigmoidal curves for representing recrystallization kinetics for two successive iteration steps $i$ and $i+1$ during which the temperatures are assumed to be constant at $T_i$ and $T_{i+1}$ respectively. The two curves have different start and finish times for recrystallization because of the change in temperature. The shift in the recrystallization kinetics curve due to the change in temperature between steps $i$ and $i+1$ is represented by:

$$\Delta t_{0,s,1} = \frac{\partial t_{0,s}}{\partial T} \Delta T_i = -Ae^{\frac{Q}{RT_i}} \exp \left( \frac{Q}{RT_i} \right) \frac{Q}{RT_i^2} \Delta T_{i+1} \quad (5.18)$$

At the end of iteration step $i$, the recrystallized fraction is $X_i$. For the next iteration step $i+1$, the fraction transformed $X_{i+1}$ needs to be computed. The procedure for computing $X_{i+1}$ is described below.

- First a fictitious time $t_{i+1, fict}$ is computed as the time needed at the new temperature $T_{i+1}$ for transformation of the fraction that was completely transformed at the end of the previous iteration $X_i$.
- Then the new time step is added to $t_{i+1, fict}$ and at this new time $t_{i+1, fict} + \Delta t$, the new fraction transformed $X_{i+1}$ is determined.
To validate this procedure, recrystallization experiments were conducted under changing temperature conditions. The experimental procedure consisted of initially heating the specimen and homogenization to obtain the required grain size. This was followed by cooling to the test temperature where it was deformed to a predetermined strain at a constant predetermined strain rate. After deformation, the specimen was cooled at a constant rate and the recrystallized fraction was measured at different temperatures along the cooling curve. To measure recrystallized fraction at some temperature during cooling, the specimen was deformed immediately upon reaching the temperature during cooling. The stress-strain curve recorded was compared with that of a fully recrystallized material to determine the fraction recrystallized. Figures 5.24(a) and (b) show a comparison between measured and predicted fraction recrystallized for two of the tests. A fairly good agreement between the predictions and measurements was found in all the tests conducted thereby validating the procedure.

5.10 Recrystallization kinetics under changing strain rate conditions

It is known that strain rate has a significant effect on static recrystallization kinetics. Increasing the strain rate increases the rate of recrystallization. The driving force for recrystallization is the strain energy stored in the material. The higher strain rate (keeping other conditions same) gives rise to higher dislocation density and higher stored energy and consequently faster recrystallization.
(a) The specimen with an initial grain size of 75 microns was deformed to a strain of 0.2 at a strain rate of 0.1/sec at 1050 °C before being cooled to 975 °C in 10 seconds.

(b) The specimen with an initial grain size of 200 microns was deformed to a strain of 0.2 at a strain rate of 1.0/sec at 1050 °C before being cooled to 950 °C in 10 seconds.

Figure 5.24 Comparison between measured and predicted fraction recrystallized at different instants under changing temperature conditions.
Like temperature, strain rate changes continuously in a rolling process. The strain rate increases from zero at the entry to the rolls to a certain maximum value between the entry and exit and falls back to zero at the roll exit. The austenite evolution models have been developed under conditions of constant strain rate. To use these models for modeling microstructural evolution in a rolling process, a procedure has to be adopted to determine recrystallization kinetics in a changing strain rate situation. Some researchers have used an average strain rate to model recrystallization kinetics while others have used the strain rate near the exit. In this study, an attempt was made to develop a scientific procedure to adapt the microstructure evolution models to changing strain rate conditions.

Consider for example a case in which deformation up to a certain strain ($\varepsilon_1 + \varepsilon_2$) is carried out in one deformation step but using two strain rates as shown in figure 5.25. The initial strain $\varepsilon_1$ is carried out at a strain rate of $\dot{\varepsilon}_1$ and the final strain $\varepsilon_2$ is carried out at a strain rate of $\dot{\varepsilon}_2$. The temperature is held constant at some temperature $T$ during this deformation. Consider two points $A$ and $B$ as shown in figure 5.24. $A$ corresponds to the end of the deformation at the strain rate of $\dot{\varepsilon}_1$ while $B$ corresponds to the beginning of subsequent deformation at $\dot{\varepsilon}_2$. Between points $A$ and $B$, technically there is no deformation and no additional strain energy gets added to the material. Consequently, the recrystallization kinetics at points $A$ and $B$ should be the same, i.e. $t_{0.5}^A = t_{0.5}^B$. Using this argument, the equivalent initial strain ($\varepsilon_2^*$) corresponding to strain rate $\dot{\varepsilon}_2$ that satisfies the above condition is given by:
\[
\left( \varepsilon_{2t} = \varepsilon_1 \left( \frac{\dot{\varepsilon}_1}{\dot{\varepsilon}_2} \right)^m \right)^{1/m} \text{ at constant temperature}
\]

where, \( l \) and \( m \) are constants from equations 5.10 or 5.12.

(a) The initial deformation is at a strain rate of \( \dot{\varepsilon}_1 \) and the final deformation is at a strain rate of \( \dot{\varepsilon}_2 \).

(b) Flow stress curves at different strain rates

Figure 5.25 Procedure for adapting the recrystallization model under changing strain rate conditions
To determine the recrystallization rate after the additional strain at \( \dot{\varepsilon}_2 \) (point \( C \)), the strain \( \varepsilon_1 \) is added to the initial strain \( \varepsilon_3 \), to obtain the new equivalent strain \( \varepsilon_2 \) and used in equation 5.10 (or 5.12) to compute the new time for 50% recrystallization (figure 5.25). This procedure is repeated for any additional strain rate steps in the deformation process. In using this procedure for decreasing strain rates, the equivalent strain \( \varepsilon_2 \) corresponding to the lower strain rate is computed. If this strain is larger than the strain for 95% dynamic recrystallization (from equation 6.17), the strain, strain rate and time for 50% recrystallization in the previous step is preserved and used for modeling microstructural evolution in the interstand region. This follows from the fact that strain does not have any influence on recrystallization kinetics after the flow curve reaches the steady state regime.

To validate this procedure, isothermal compression tests were conducted under changing strain rate conditions. The experimental procedure consisted of heating the specimen to obtain an initial grain size of 200 microns. The specimen was deformed to a strain of 0.5 at a temperature of 975°C under different conditions of strain rate as shown in table 5.5. In these tests, the total deformation (0.5 strain) was divided into equal strain increments. During each increment, the strain rate was held constant as listed in the table. The measured fraction recrystallization within 3 seconds after deformation was used to determine the time for 50% recrystallization. Table 5.5 also lists the time for 50% recrystallization computed using different techniques. It is clearly seen that the procedure described above gives the best correlation between measurements and predictions.
Table 5.5 Compression tests under changing strain rate conditions.

( $t_{0.5}$ is the predicted time for 50% recrystallization using the procedure described. $t_{0.5 \_2}$ and $t_{0.5 \_3}$ are the predicted 50% recrystallization using average strain rate and exit strain rate respectively)

This approach assumes that the grain structure is the same at points $A$ and $B$. This is not necessarily true because the subgrain size during deformation depends upon the stress state (McQueen and Jonas, 1976) which could be different at the two points.

Immarigeon and Jonas (1971) studied the evolution of subgrain size during transient deformation in Armco iron at 675°C. The substructural changes initiated by a rapid change in strain rate at the strains are indicated in figure 5.26. The upper figure shows the change in subgrain diameter with strain when the strain rate was rapidly decreased from $10^{-1}$ to $10^{-3}$ /sec. The lower figure similarly shows the subgrain size change when the strain rate was increased from $10^{-3}$ to $10^{-1}$ /sec. It was shown that the substructural modifications associated with hardening take place over greater strains than those associated with softening between the same limits.
Figure 5.26 Substructural changes initiated by a rapid change in strain rate (Immarigeon et al, 1971)
5.11 Physical Simulation Experiments

Physical simulation experiments were conducted on the Gleeble 3500 to validate the microstructure evolution models and procedures developed for using them. Each experiment consisted of a five stage compression test as depicted in figure 5.27. Initially the specimen was heated to a high temperature $T_0$ to obtain an initial grain size of 200 microns. The specimen was then cooled to temperature $T_1$ and compressed to a strain of $\varepsilon_1$ at a certain strain rate. It was then cooled to a temperature $T_2$ at a predetermined cooling rate before being hit again. The specimen was thus successively compressed and cooled as illustrated in figure 5.27. The grain sizes in the specimens were examined after each hit.

Table 5.6 shows the conditions under which these physical experiments were run. The conditions of temperature, strains in each pass and inter-hit times were chosen to reflect conditions encountered in a bar rolling process. Figures 5.28, 5.29 and 5.30 compare the austenite grain size predicted using the microstructure evolution models with measurements at different instants. It was seen that the predictions were within ±10 microns of the measured austenite grain sizes in most cases.
Figure 5.27 Thermomechanical procedure in a physical simulation experiment
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Table 5.6 Physical simulation experiments
Figure 5.28 Physical simulation experiment 1: Comparison between measured and predicted austenite grain sizes.
Figure 5.29 Physical simulation experiment 2: Comparison between measured and predicted austenite grain sizes.
Figure 5.30 Physical simulation experiment 3: Comparison between measured and predicted austenite grain sizes.
5.12 Summary

In this chapter, the recrystallization and grain growth kinetics for TMS80R were studied and quantitative models were developed to model these phenomena during hot rolling. This chapter also presented procedures for applying these models under changing strain rate and temperature conditions in an FEM model. These models and procedures were validated by means of additional validation experiments. Integration of these models into ROLPAS will be presented in chapter 7.
CHAPTER 6

A MICROSTRUCTURE DEPENDENT FLOW STRESS MODEL FOR STEEL ROLLING

6.1. Introduction

In recent years, finite element techniques have been widely used to model and analyze the hot rolling process. FEM analysis provides useful information including metal flow, roll forces and thermomechanical history (strain, strain rate and temperature at different instants in time). The success of this technique depends upon the availability of accurate material property data. Flow stress has a significant effect on the roll forces and the flow of material. Accurate determination of flow stress is therefore important in finite element modeling of hot rolling.

Flow stress of steels is known to be a function of strain, strain rate, temperature, chemistry and microstructure. Modeling flow stress as a function of microstructure is a complex problem since the microstructure constantly changes during a multiple pass hot working process. As a result, flow stress was previously modeled as a function of strain, strain rate and temperature. Several researchers have reported the effects of these parameters on the flow behavior of steels (Rao, et al.). Many mathematical models have
been developed and applied to process control. However, these models ignore the effects of microstructural changes taking place during deformation. With significant progress being made in recent years in modeling of microstructural evolution as discussed in the earlier chapters, it has become possible to model flow stress as a function of the evolving microstructure.

Accurate modeling of flow stress involves taking into consideration different phenomena such as work hardening, dynamic recovery and dynamic recrystallization. Significant progress was made by Laasraoui and Jonas (1992) in developing a microstructure based flow stress model. While this model was able to predict with reasonable accuracy the effects of strain hardening, dynamic recovery, and recrystallization, the effect of the austenite grain size on flow stress was not included. Grain size is known to have a significant effect on the flow behavior of metals. A well known example is the Hall-Petch relationship. This chapter describes the development of a microstructure dependent flow stress model using Laasraoui and Jonas’s approach which takes into account the effect of austenite grain size and also describes a computationally efficient procedure to integrate this model into FEM.

6.2 Experimental Procedure

Experimental flow stress data were obtained by conducting uniaxial hot compression tests on TMS-80R microalloyed steel using a Gleeble thermomechanical simulator. Cylindrical specimens (10 mm dia. X 15 mm long) were used in the hot compression tests. A full factorial experiment was designed with strain rate, temperature
and austenite grain size as the control variables. Table 6.1 shows the test matrix used in experiments.

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Table 6.1 Test matrix for flow stress modeling

The experimental procedure for modeling flow stress (figure 6.1) consisted of heating the specimen to temperature $T_0$ and holding there for a controlled amount of time $t_0$ to obtain the required austenite grain size before being cooled to the test temperature.
The specimen was then held at the test temperature for 60 seconds followed by uniaxial compression to a strain of 0.75 at a predetermined strain rate.

![Thermomechanical cycle for flow stress modeling](image)

**Figure 6.1** Thermomechanical cycle for flow stress modeling

### 6.3 Development of the model

A typical stress-strain curve is shown in figure 6.2. In the initial part of the curve, the increase in dislocation density and pile up of dislocations at the grain boundaries causes the stress to rise with strain. This is referred to as strain hardening. The strain hardening rate $d\sigma/d\varepsilon$ decreases with increasing strain due to dynamic recovery which results in annihilation of dislocations in the deformed grains. Under some conditions (high temperatures and low strain rates) the stress-strain curve displays a clear peak at the peak strain ($\varepsilon_p$) after which, the flow stress decreases due to dynamic recrystallization.
Dynamic recrystallization occurs only after a critical strain is exceeded. The critical strain for the onset of dynamic recrystallization \( (\varepsilon_c) \) is reported to be somewhat lower than the peak strain and depends on temperature, strain rate and the grain size. Many researchers estimate the critical strain using the following relationship:

\[
\varepsilon_c = 0.83 \cdot \varepsilon_p
\]  

The equation for the peak strain (which is a function of the grain size, strain rate and temperature) obtained from regression analysis is:

\[
\varepsilon_p = 1.20 \times 10^{-3} d_0^{0.27} \left( \dot{\varepsilon} \exp \left( \frac{280000}{RT} \right) \right)^{0.18}
\]

Following Laasraoui and Jonas's approach, the stress-strain curve was divided into two regions (figure 6.2).

- The region before the critical strain where dynamic recovery is the predominant softening mechanism \( (\varepsilon < \varepsilon_c) \).

- The region after the critical strain where dynamic recovery and dynamic recrystallization act together \( (\varepsilon > \varepsilon_c) \).

In tests that displayed a clear peak due to dynamic recrystallization, the stress-strain curve before \( \varepsilon_c \) was extrapolated to a hypothetical saturation stress \( (\sigma_{ss}^*) \). This represents the saturation stress if recovery were the only softening mechanism during deformation. The difference between the saturation stress and the steady state stress \( (\sigma_{ss}^* - \sigma_{ss}^{\text{drel}}) \) is a measure of the additional softening taking place due to dynamic recrystallization.
Figure 6.2 A typical stress-strain curve
6.3.1. Effect of Austenite Grain Size

The yield stress \( \sigma_{y} \), the saturation stress \( \sigma^*_{\alpha} \) and the steady state stress \( \sigma_{\alpha}^{\text{sat}} \) are important attributes in the stress-strain curve. The model for flow stress proposed in this study makes use of models for \( \sigma_{y} \), \( \sigma^*_{\alpha} \), and \( \sigma_{\alpha}^{\text{sat}} \) in computing the flow stress. The primary emphasis of this effort is on modeling the effect of the initial austenite grain size (in addition to thermomechanical variables) on the flow stress. Consequently, the effect of austenite grain size on these important attributes was analyzed.

1. Effect on yield stress

The yield stress was determined graphically from the recorded stress-strain data as the stress corresponding to 0.2% permanent strain. According to McQueen and Jonas (1975), evidence exists that for grain sizes above 20 microns, the high temperature yield stress increases with decreasing grain size. Researchers have shown that the strengthening component due to decreasing grain sizes follow the Hall-Petch relation. Figure 6.3 shows the variation of yield stress with grain size \( d_{\alpha} \) for some of the strain rates and temperatures used in this study. It was observed that grain size did have a significant effect on the yield stress.

2. Effect on saturation stress

Figure 6.4 shows the variation of saturation stress \( \sigma^*_{\alpha} \) with initial grain size. As in the case of yield stress, \( \sigma^*_{\alpha} \) showed a strong dependence on the grain size. The
saturation stress was seen to be larger for smaller austenite grains and vice-versa. On account of the significant effect grain size had on the saturation stress, it was deemed important to include grain size as a variable in the model for saturation stress.

Figure 6.3 Effect of grain size on yield stress

Figure 6.4 Effect of grain size on saturation stress
3. Effect on steady state stress

Figure 6.5 shows the variation of steady state stress ($\sigma_{\text{ss}}$) with the initial austenite grain size under different conditions of strain rates and temperatures. Unlike yield stress and saturation stress, the steady state stress was not affected significantly by a change in the initial grain size. This is a result of the formation of a steady state substructure accompanying dynamic recrystallization which annuls the effect of initial grain size in the steady state condition.
6.3.2 Modeling of yield stress

In addition to initial grain size, the yield stress showed a strong dependence upon temperature and strain rate. Figure 6.6 shows how flow stress varies with temperature at a strain rate of 1.0/sec. In accordance with the relationship displayed in figures 6.3 and 6.6, an expression of the following form was used for modeling the yield stress.

\[
\sigma_{yp} = A_1 \cdot \dot{\epsilon}^m d_0^n \exp \left( \frac{Q_{yp}}{RT} \right)
\]  

(6.3)

In this model, \(Q_{yp}\) the apparent activation energy, and \(A_1, m_1 \& n_1\) the material constants, were determined by means of a multiple regression analysis.

\[
\sigma_{yp} = 3.25 \cdot d_0^{-0.09} \exp \left( \frac{315000}{RT} \right)^{0.073}
\]  

(6.4)

Figure 6.6 Effect of temperature on yield stress
6.3.3 Modeling of flow stress in the strain hardening regime (region 1: $\varepsilon < \varepsilon_c$)

The classical approach to modeling the flow stress in the strain hardening regime consists of analyzing the stress-strain data on a $(\Theta - \sigma)$ plot, where $\Theta$ is the strain hardening rate $d\sigma/d\varepsilon$. Several researchers have found that the data fits the following relationship quite well.

$$\Theta = \frac{A}{\sigma} - B\sigma$$ (6.5)

This equation is a combination of the linear $(\Theta-\sigma)$ relationship proposed by Kocks and the linear $(\Theta-1/\sigma)$ relationship proposed by Roberts. The intercept of this equation with the $\Theta$ axis is the saturation stress ($\sigma^*_s$). Integration of equation 6.5 results in the following expression for flow stress in the strain hardening regime.

$$\sigma_{\text{drec}} = \left[\sigma^*_s - \sigma_0^2 + \sigma_0^2 - \sigma^*_s - 1\right]^{\frac{1}{5}} - \Omega \varepsilon$$ (6.6)

Here, $\sigma_0$ is the initial stress and $\Omega$ represents the ease of dynamic recovery. $\sigma^*_s$ and $\Omega$ are related to $A$ and $B$ in equation 6.5 as follows

$$\sigma^*_s = \sqrt{-\frac{A}{B}}$$ (6.7)

$$\Omega = -2B$$ (6.8)

$\sigma^*_s$ and $\Omega$ were determined for all the test runs using the technique discussed above while $\sigma_0$ values were determined by graphically fitting equation 6.6 to the experimental data. All three parameters showed a strong dependence on strain rate, temperature and grain...
size. In order to use equation 6.6 to compute the flow stress, structure and process dependent models of $\sigma_{ss}$, $\Omega$ and $\sigma_0$ are required.

Model for saturation stress

Laasraoui and Jonas modeled the temperature and strain rate dependence of saturation stress $\sigma_{ss}$ using the hyperbolic sine law

$$\dot{\varepsilon} = A_2 \left[ \sinh (\alpha \sigma_{ss}) \right]^{n_2} \cdot \exp \left( -\frac{Q_{app}}{RT} \right)$$

(6.9)

where, $Q_{app}$ is the apparent activation energy of deformation, $A_2$, $\alpha$ and $n_2$ are material constants. In view of the linear relationship between $\ln(\sigma_{ss})$ and $\ln(d_0)$ observed in this study (figure 6.4), equation 6.9 was modified to equation 6.10 to account for the effect of the initial grain size.

$$\dot{\varepsilon} = A_2 \left[ \sinh \left( \alpha \cdot \sigma_{ss} \cdot \frac{d_0}{75} \right) \right]^{n_2} \exp \left( -\frac{Q_{app}}{RT} \right)$$

(6.10)

The constants $Q_{app}$, $A_2$, $\alpha$, $m_2$ and $n_2$ in this equation were determined for TMS80R using regression analysis. Rearranging terms, the following model was obtained for the saturation stress.

$$\sigma_{ss} = 90.9 \left( \frac{d_0}{75} \right)^{-0.107} \sinh^{-1} \left[ 3.31 \times 10^{-3} \left( \dot{\varepsilon} \cdot \exp \left( \frac{355000}{RT} \right) \right) ^{0.192} \right]$$

(6.11)

In figure 6.7, $\sinh(\alpha \sigma_{ss})$ is plotted against the apparent Zener-Hollomon parameter ($Z_{app} = \dot{\varepsilon} \cdot \exp(Q_{app} / RT) = \dot{\varepsilon} \cdot \exp(355000 / RT)$). The graph indicates a slight but
significant influence of the initial grain size on the saturation stress thereby supporting the form of equation 6.7 over equation 6.6 for modeling saturation stress.

![Plot of saturation stress Vs apparent Zener-Hollomon parameter](image)

Figure 6.7 Plot of saturation stress Vs apparent Zener-Hollomon parameter

**Equation for Ω**

Ω in equation 6.6 is a measure of the ease with which dynamic recovery occurs. Ω is expected to rise with decreasing strain rate and increasing temperature. Following Yoshie et al, a relationship of the form was used in this study to model the ease of dynamic recovery.

\[
Ω = A_Ω d_0^{n_Ω} \dot{\varepsilon}^{m_Ω} \exp\left(\frac{Q_Ω}{RT}\right)
\]  

(6.12)

In this expression, \(d_0\) is the initial austenite grain size, \(Q_Ω\) is the apparent activation energy and \(A_Ω, n_Ω\) and \(m_Ω\) are assumed to be constants and their values determined using a least squares analysis.
Laasraoui and Jonas have reported that \( m_\Omega \) is temperature dependent. In view of this fact, the value of \( m_\Omega \) used in this study would be an average value over the range of temperatures used in the compression tests.

**Equation for \( \sigma_0 \)**

Like \( \Omega \) and \( \sigma^\ast_{\nu} \), the initial stress \( \sigma_0 \) also showed a strong dependence upon strain rate, temperature and grain size. Earlier in this chapter, a model for yield stress (\( \sigma_{yp} \)) was proposed (equation 1). It was found that \( \sigma_0 \) displayed a strong linear correlation with \( \sigma_{yp} \) (figure 6.8). Consequently, \( \sigma_0 \) is determined by the following equation.

\[
\sigma_0 = 0.95 \cdot \sigma_{yp}
\]

(6.14)
6.3.4 Modeling of flow stress during dynamic recrystallization (region 2: $\varepsilon > \varepsilon_c$)

To model softening due to dynamic recrystallization, an Avrami type equation was used. The fractional softening due to dynamic recrystallization $X_d$ is represented by

$$X_d = \frac{\sigma - \sigma^d_{\text{rec}}}{\sigma^* - \sigma^d_{\text{rec}}} = 1 - \exp\left(-K_d\left(\frac{\varepsilon - \varepsilon_c}{\varepsilon_{0.5} - \varepsilon_c}\right)^{n_d}\right)$$  \hspace{1cm} (6.15)

where, $\sigma^d_{\text{rec}}$ represents the flow stress if dynamic recovery were the only softening mechanism (equation 6.6), $\sigma^d_{\text{rec}}$ represents the steady state stress after recrystallization has progressed through the material, $\varepsilon_c$ is the critical strain for the onset of dynamic recrystallization, $\varepsilon_{0.5}$ is the strain corresponding to 50% softening due to dynamic recrystallization and $K_d$ and $n_d$ are constants and take the values 0.693 and 1.46 respectively. Equation 6.15 can be rewritten as follows to obtain a flow stress equation:

$$\sigma = \sigma^d_{\text{rec}} - \left[\sigma^* - \sigma^d_{\text{rec}}\right] \cdot \left[1 - \exp\left(-0.693\left(\frac{\varepsilon - \varepsilon_c}{\varepsilon_{0.5} - \varepsilon_c}\right)^{1.46}\right)\right]$$  \hspace{1cm} (6.16)

It may be noted that the fractional softening in equation 6.15 does not represent the extent of dynamic recrystallization. Rather, it is only a measure of softening due to dynamic recrystallization. Nevertheless, it assumes a sigmoidal form for the softening kinetics. This form of the equation has been used successfully by other researchers. The strain for 50% softening was modeled using an equation similar to that used by Sellars et al (1985).

$$\varepsilon_{0.5} = 1.43 \times 10^{-3} \cdot d_0^{0.239} \cdot \dot{\varepsilon} \cdot \exp\left(\frac{256305}{RT}\right)^{0.21}$$  \hspace{1cm} (6.17)
Model for steady state stress

Following the approach used by Laasraoui and Jonas, a hyperbolic sine equation (equation 6.18) was used to model the steady state stress ($\sigma_{ss}^{\text{dref}}$).

$$\dot{\varepsilon} = A_3 \cdot [\sinh(\alpha_3 \dot{\sigma}_{ss}^{\text{dref}})]^{n_3} \cdot \exp\left(-\frac{Q}{RT}\right)$$

(6.18)

The constants in the equation were obtained using regression analysis. Rearranging terms in equation 6.18, the following equation for the steady state stress was obtained.

$$\sigma_{ss}^{\text{dref}} = 90.91 \cdot \sinh^{-1}\left[1.97 \times 10^{-1} \cdot \left[\dot{\varepsilon} \cdot \exp\left(\frac{300000}{RT}\right)\right]^{0.247}\right]$$

(6.19)

Figure 6.9 shows the dependence of $\sigma_{ss}^{\text{dref}}$ on the Zener-Hollomon parameter Z. The hyperbolic sine equation (6.18) was found to give a good fit to the data.

Here, $Q$ is the activation energy of deformation, $A_3$, $n_3$, and $\alpha_3$ are assumed to be constants. Unlike saturation stress, the data fails to show any significant effect of initial grain size on the steady state stress. This justifies the exclusion of initial grain size from the equation for steady state stress.
Figure 6.9 Plot of steady state stress vs. Zener-Hollomon parameter
6.3.5 Flow stress model summary

The flow stress model for TMS80R steel is as follows:

1. Critical strain: \( \varepsilon_c = 0.83 \varepsilon_p \).
   \[ \text{where} \]
   \[ \sigma^{\text{drec}} = \left[ \sigma_{\alpha}^2 + (\sigma_0^2 - \sigma_{\alpha}^2) \right]^{0.5} \]

2. For \( \varepsilon < \varepsilon_c \)
   \[ \sigma = \sigma^{\text{drec}} - \left[ \sigma_{\alpha}^* - \sigma_{\alpha}^{\text{drec}} \right] \left[ 1 - \exp \left( -0.693 \left( \frac{\varepsilon - \varepsilon_c}{\varepsilon_{0.5} - \varepsilon_c} \right)^n \right) \right] \]

where: \( \varepsilon_p = 1.20 \cdot 10^{-3} \cdot d_0^{0.263} \cdot \dot{\varepsilon}^{2.16} \cdot \exp \left( \frac{230474 \times 0.216}{RT} \right) \)

3. For \( \varepsilon > \varepsilon_c \)
   \[ \sigma = \sigma^{\text{drec}} - \left[ \sigma_{\alpha}^* - \sigma_{\alpha}^{\text{drec}} \right] \left[ 1 - \exp \left( -0.693 \left( \frac{\varepsilon - \varepsilon_c}{\varepsilon_{0.5} - \varepsilon_c} \right)^n \right) \right] \]

where:
\( \sigma_0 = 3.25 \cdot \dot{\varepsilon}^{0.073} \cdot d_0^{0.09} \cdot \exp \left( \frac{314223 \times 0.073}{RT} \right) \)

\( \Omega = 39.25 \cdot \dot{\varepsilon}^{-0.076} \cdot d_0^{-0.08} \cdot \exp \left( \frac{-228215 \times 0.076}{RT} \right) \)

\( \sigma_{\alpha}^* = 90.9 \left( \frac{d_0}{75} \right)^{-0.107} \sinh^{-1} \left[ 3.31 \times 10^{-3} \left( \dot{\varepsilon} \cdot \exp \left( \frac{354819}{RT} \right) \right)^{0.192} \right] \)

\( \sigma_{ss}^{\text{drec}} = 90.91 \cdot \sinh^{-1} \left[ 1.97 \times 10^{-3} \left( \dot{\varepsilon} \cdot \exp \left( \frac{290000}{RT} \right) \right)^{0.247} \right] \)

\( \varepsilon_{0.5} = 1.43 \times 10^{-3} \cdot d_0^{0.239} \left( \dot{\varepsilon} \cdot \exp \left( \frac{256305}{RT} \right) \right)^{0.21} \)

n in equation 3 is 1.46
Effect of temperature

![Graph showing the effect of temperature on flow stress.](image)

a. ($\dot{\varepsilon} = 1.0/\text{sec}, d_0 = 250\mu$)  
b. ($\dot{\varepsilon} = 0.1/\text{sec}, d_0 = 75\mu$)

Figure 6.10 Effect of temperature on flow stress

Effect of strain rate

![Graph showing the effect of strain rate on flow stress.](image)

a. ($T=1000^\circ\text{C}, d_0=75\mu$)  
b. ($T=1100^\circ\text{C}, d_0=250\mu$)

Figure 6.11 Effect of strain rate on the flow stress
6.4 Comparison of model predictions with experimental measurements

Application to single stage deformation

In the first stage of validation, the model was applied to single hit compression tests. Figures 6.10 and 6.11 show a comparison of stress-strain curves predicted by the model and the experimentally measured values. It is seen that the predictions are fairly good over the range of variables considered in the experiment. The model was found to be able to accurately model the effect of strain hardening, dynamic recovery as well as dynamic recrystallization.

Application to multiple stage deformation

Hot rolling is an incremental deformation process in which deformation progresses in multiple stages. After deformation in a roll bite, there is typically a gap of few seconds before the material undergoes deformation in the next roll bite. During the period between the successive roll passes, metallurgical changes such as static recovery, static recrystallization and grain growth change the flow characteristics of the material. The flow stress model, in order to be applicable to a rolling process should be able to predict the stress-strain curve in a multiple hit compression test. To validate the flow stress model, comparisons were carried out between the model predictions and measurements in a two hit compression test. The two hit compression test was earlier used for modeling recrystallization kinetics as described in chapter 5. The test sequence (figure 6.12) consisted of heating the specimen to a temperature $T_0$, holding there for a controlled amount of time $t_0$ to obtain the required austenite grain size before being
cooled to the test temperature. The specimen was then held at the test temperature for 60 seconds followed by uniaxial compression to a chosen strain ($\varepsilon_1$) at a predetermined strain rate. The specimen was then held at the same temperature for a short duration $t$ before being compressed again to a total strain of $2 \cdot \varepsilon_1$.

![Thermomechanical procedure in a two-hit compression test](image)

Figure 6.12 Thermomechanical procedure in a two-hit compression test

One of the important issues that need to be addressed in a multiple deformation process is the handling of partial recrystallization. In a reversing bar mill, the time available between two successive roll passes may be sufficient for complete recrystallization to occur. On the other hand, the interstand time in a continuous rolling mill may not be long enough for complete recrystallization. Researchers have proposed different approaches for handling partial recrystallization. One of the approaches uses the law of averages for the effective grain size and the retained strain. In this method, the average grain size and the retained strain are computed as follows

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\[ d_{\text{avg}} = (X \cdot d_{\text{rex}}^3 + (1 - X) \cdot d_0^3)^{1/3} \]  

(6.20)

\[ \varepsilon_{\text{ret}} = (1 - X) \cdot \varepsilon_1 \]  

(6.21)

where, \( d_{ avg } \) is the average grain size, \( d_{ rex } \) is the recrystallized grain size, \( d_0 \) is the initial austenite grain size and \( X \) is the fraction recrystallized, while \( \varepsilon_{ ret } \) is the retained strain and \( \varepsilon_1 \) is the strain at the end of first hit. Researchers have used the average grain size and retained strain to determine recrystallization kinetics with reasonable success.

Another approach to handling partial recrystallization was suggested by Karhausen and Kopp. It consists of treating the recrystallized and unrecrystallized fractions separately unlike averaging as above.

In this study, the flow stress predictions using different techniques were tested. Comparisons between predicted and measured flow stress values were carried out for various conditions of strain, strain rate, temperature, initial austenite grain size and inter-hit holding times. One of the cases is presented below. In this test, the two-hit compression test was carried out under conditions listed in table 6.2.

<table>
<thead>
<tr>
<th>Test variable</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>strain per hit</td>
<td>0.25</td>
</tr>
<tr>
<td>strain rate</td>
<td>0.1 / sec</td>
</tr>
<tr>
<td>temperature</td>
<td>1100 °C</td>
</tr>
<tr>
<td>initial grain size</td>
<td>100 microns</td>
</tr>
<tr>
<td>inter-hit hold times</td>
<td>5 sec, 1 sec</td>
</tr>
</tbody>
</table>

Table 6.2 Conditions in a two-hit compression test
In order to model the recrystallization during the inter-hit hold time, a recrystallization (kinetics and grain size) model developed earlier for the same grade of steel was utilized. The model predicted 94.4% recrystallization under the test conditions for a hold time of 5 seconds and 43.7% recrystallization for the 1 second hold time. The recrystallized grain size was estimated to be about 59 microns.

Figure 6.13 Flow stress prediction using the average grain size and retained strain showed poor agreement between measurements and predictions for the second deformation cycle (1 second inter-hit time)

6.4.1 Using average grain size and retained strain

In this method, the average grain size and the average retained strain computed after modeling the recrystallization kinetics during the inter-hit period are used in computing the flow stress for subsequent deformation. Figure 6.13 which corresponds to
1 second hold time and represents a case of partial recrystallization, shows a poor agreement between prediction (using the averaging technique) and experimental measurements in the initial part of the second hit although there is a fairly good agreement in the latter part of the second hit. This result implies that the use of average grain size and average retained strain does not work well for modeling flow stress.

6.4.2 Using modified form of Karhaussen and Kopp’s technique

A modification of Karhausen and Kopp’s method was used. While Karhausen and Kopp compute the volume fractions at every small time increment during deformation, the nature of the flow stress model used in this study obviates these cumbersome computations. The procedure adopted in this study is illustrated in figure 14. The new sub-structures are assumed to evolve during the period between two successive deformations. The recrystallization model is applied to each existing substructure to determine the grain sizes and the volume fractions of the evolving substructures. Next, the flow stress model is applied to each of the substructures to determine their individual flow stresses. The weighted average of these individual flow stress is the flow stress of the partially recrystallized material.

\[ \sigma = \sum f_i \cdot \sigma_i \]  

where, \( f_i \) is the fraction of substructure \( i \) with flow stress \( \sigma_i \).
Figure 6.14 Evolution of substructures

Figure 15(a,b) shows the comparison between measured and predicted flow stress curves using this approach. The agreement between the two is seen to be very good. The model predicts accurately the drop in flow stress due to dynamic recrystallization in the previously unrecrystallized fraction as well as the increased strain hardening rate in the fully recrystallized case due to grain refinement. The advantage of this model over the Karhaussen and Kopp model is that the number of substructures to be handled during deformation does not increase, thereby resulting in faster computation and requiring significantly less computer memory. The problem however with this approach is that the number of fractions to be handled rises exponentially in a multi-stage deformation process due to the multiplication effect of the evolving substructures (figure 14). The number of substructures existing after the n-th stage of deformation is $2^n$. Storage of thermomechanical history of these substructures requires an enormous amount of
computer memory thereby making this technique unsuitable for application to multi-pass hot rolling.

Figure 6.15 Comparison between measurements and predictions in two-hit compression.

(a) Five seconds inter-hit time

(b) One second inter-hit time
6.4.3 Approximation using the rule of mixtures

While the modified Karhausen and Kopp's technique was shown to be accurate and convenient for modeling a few stages of deformation, a modification is proposed to counter the difficulty in application to multi-stage deformation. This approximate method is illustrated in figure 16. Consider the second hit during which two substructures (1 and 2) exist in the material. After deformation, the grain sizes and the accumulated strains in the two substructures are averaged to obtain the effective grain size and the average accumulated strain which are then associated with substructure 3. Static recrystallization model is then applied to substructure 3 to determine the fraction and size of the evolving substructure (substructure 4). Consequently substructures 3 and 4 are carried forward to the third deformation step. This procedure is repeated over the entire deformation sequence. At any stage of deformation except the first, there are only two substructures that need to be used in computing the flow stress thereby making it computationally very efficient.

![Figure 6.16 Representation of substructures in the approximate method using rule of mixtures](image)

Figure 6.16 Representation of substructures in the approximate method using rule of mixtures
Figure 6.17 compares the predictions of the above procedure with the modified Karhausen and Kopp’s model for a hypothetical five-hit compression test conducted at a strain rate of 0.1/sec at 1100°C with an initial grain size of 100 microns and a 1 second hold time between successive hits. The predictions of the proposed approximate method agreed very well with the predictions using the modified Karhausen and Kopp method. The difference between the two was less than 7% in this case.

This technique makes use of the fact that grain size has a lesser effect on the flow stress at higher strains. Consequently this technique has been found to work well for incremental strains larger than 0.1 under the conditions of temperature, strain rates and grain sizes considered in this investigation. This method also makes use of the notion that the average grain size and average strain can provide reasonable predictions for recrystallization kinetics. Support for this notion can be found in the fact that estimates for average grain size (58.3 microns) and the average strain at the end of the last hit (0.42) using this method agreed well with those obtained using the modified Karhausen and Kopp technique (56.8 microns and 0.40 respectively).

To validate this procedure, comparisons were made between predictions and measurements in five hit compression tests conducted on the gleeble machine. The test procedure consisted of heating the test specimen to obtain a desired initial grain size. This was followed by cooling to the test temperature and homogenization before being compressed to a strain of 1.0 in five equal strain increments. A short hold time was provided between two successive hits as in a two hit compression test. Figures 6.18 and
6.19 illustrate two such cases looked at. The predicted flow stress values agreed very well with the measurements thereby validating not only the flow stress modeling procedure but also the austenite evolution models.

![Stress-strain predictions](image)

(a) Stress-strain predictions using the proposed approximate procedure

![Stress-strain predictions](image)

(b) Stress-strain predictions using the modified Karhausen and Kopp's technique

Figure 6.17 Approximate procedure using rule of mixtures vs. modified Karhausen and Kopp's technique
Figure 6.18 Five hit compression test conducted at a temperature of 1100°C and a strain rate of 1.0/sec with an initial grain size of 200 microns. Inter-hit time was 1 second.
Figure 6.19 Five hit compression test conducted at a temperature of 1100°C and a strain rate of 1.0/sec with an initial grain size of 200 microns. Inter-hit time was 0.2 second.
6.4.4 Application of the flow stress model under changing strain rate conditions

To apply these models under changing strain rate conditions, the same approach as was described in section 5.10 was used. Consider for example a case in which deformation up to a certain strain \((\varepsilon_1 + \varepsilon_2)\) is carried out in one deformation step but using two strain rates as shown in figure 6.20. The initial strain \(\varepsilon_1\) is carried out at a strain rate of \(\dot{\varepsilon}_1\) and the final strain \(\varepsilon_2\) is carried out at a strain rate of \(\dot{\varepsilon}_2\). The temperature is assumed to be held constant at some temperature \(T\) during this deformation. Consider two points \(A\) and \(B\) as shown in figure 6.20(a). \(A\) corresponds to the end of the deformation at the strain rate of \(\dot{\varepsilon}_1\) while \(B\) corresponds to the beginning of subsequent deformation at \(\dot{\varepsilon}_2\). Between points \(A\) and \(B\), technically there is no deformation and no additional strain energy gets added to the material. Consequently, the recrystallization kinetics at points \(A\) and \(B\) should be the same, i.e. \(t_{0.5} \big|_A = t_{0.5} \big|_B\). Using this argument, the equivalent initial strain \((\varepsilon_{21})\) corresponding to strain rate \(\dot{\varepsilon}_2\) that satisfies the above condition is given by:

\[
\varepsilon_{21} = \varepsilon_1 \left( \frac{\dot{\varepsilon}_1}{\dot{\varepsilon}_2} \right)^m
\]

at constant temperature

where, \(l\) and \(m\) are constants from equation for time for 50% recrystallization (chapter 5).
(a) The initial deformation is at a strain rate of $\dot{\varepsilon}_1$ and the final deformation is at a strain rate of $\dot{\varepsilon}_2$.

(b) Flow stress curves at different strain rates

Figure 6.20 Procedure for adapting the flow stress model under changing strain rate conditions
To determine the flow stress at the new strain rate, incremental strains up to $\varepsilon_2$ at strain rate $\dot{\varepsilon}_2$ (point C) are added to the initial strain $\varepsilon_1$, to obtain the new equivalent strain $\bar{\varepsilon}_2$ and used to compute the flow stress. This procedure is repeated for any additional strain rate steps in the deformation process. This approach assumes that the grain structure is the same at points A and B.

To validate this procedure, isothermal compression tests were conducted under changing strain rate conditions. The experimental procedure consisted of heating the specimen to obtain an initial grain size of 200 microns. The specimen was deformed to a strain of 0.5 at a temperature of 975°C under different conditions of strain rates. Figure 6.21(a-d) show a comparison between the measured and predicted flow stresses using the procedure described above. A very good agreement between the two was observed.
Figure 6.21 Flow stress predictions vs. Measurements under changing strain rates
Figure 6.21 contd.

(c) case 3

(d) case 4
6.5 Conclusions

This study has shown that grain size has a significant effect on the flow behavior and therefore its inclusion in the flow stress model is important. Accordingly, a microstructure dependent flow stress model was developed and tested. The predictions of the model compared very well with the experimental flow stress curves. Computational procedures for handling partial recrystallization were tested. It was found that a modified form of Karhausen and Kopp’s method gave accurate results. However, its application to multi-stage deformation becomes impractical on account of the enormous computational memory requirement. An approximation using the rule of mixtures was found to give fairly accurate predictions while being computationally efficient.
CHAPTER 7

INTEGRATED SYSTEM: FRAMEWORK, APPLICATIONS AND VALIDATION

7.1 Introduction

The primary objective of this study is the development of an integrated system for analysis of metal flow and austenite evolution during hot rolling. The previous chapters described the finite element model (chapter 4), the development of the microstructural evolution models (chapter 5) and the development of the microstructure dependent flow stress model (chapter 6). In this chapter, we integrate all of them into one system to be used for simulation of metal flow and microstructural evolution in a hot rolling process.

7.2 Framework of the integrated system

The central feature of the integrated system is a 3-D finite element program ROLPAS for simulating multi-pass shape rolling (figure 7.1). The non-isothermal deformation analysis in ROLPAS is based on rigid-viscoplastic assumption of the material behavior as described in chapter 4 and uses eight-node isoparametric hexahedral elements. Deformation within the roll gap is assumed to be kinematically steady. Such an
assumption has been successfully applied earlier to steady state processes such as extrusion and rolling. This FEM model however uses a microstructure dependent flow stress model described in chapter 6.

![Figure 7.1 Framework of the integrated system](image)

Using the integration procedures described in chapters 5 and 6 a microstructure evolution module MICON was developed and integrated into ROLPAS to enable modeling of austenite evolution. MICON uses the thermomechanical history computed by the FEM model in conjunction with microstructure evolution models to model the evolution of austenite during hot rolling. The microstructural changes occurring in bar rolling are primarily due to static recrystallization and grain growth that occur in the interstand region. In cases where accumulated strain is large enough to nucleate dynamic recrystallization, metadynamic recrystallization is modeled in the interstand region following Sellars (1979). The microstructural changes in the interstand are used in computing the retained strain and the grain sizes of the recrystallized and unrecrystallized fractions and used to compute the flow stress during the next pass.
Figure 7.2 Algorithm for modeling microstructural evolution
The last section in the system is a module for modeling phase transformation called AUSTRANS. A parallel study (Phadke, 1997) was conducted to develop the framework of this module. It uses the temperature history after rolling (computed by ROLPAS) and the isothermal transformation data to model the transformation of austenite to ferrite, pearlite, bainite and martensite. This model also uses structure-property relationships to predict the mechanical properties of the rolled product. Details about this module can be found in Phadke (1997).

7.3 Procedure for modeling microstructural evolution

The evolving austenite is known to significantly affect the flow stress of the material while on the other hand, the material flow affects recrystallization kinetics. This situation calls for an iterative approach in modeling metal flow and austenite evolution in ROLPAS which models the deformation and the evolution of microstructure separately.

Figure 7.2 shows the algorithm for modeling austenite evolution. For the first pass, an initial preheated grain size is input to the program. The FEM model provides the thermomechanical history (strain, strain rate and temperature at different instants in time) at every node in the workpiece for the first pass and the first interstand. After deformation and heat transfer computations for each pass, the microstructure evolution module in conjunction with the heat transfer analysis module computes recrystallized fraction and the austenite grain size at each node in the interstand region. If the fraction recrystallized
is found to be greater than 95%, complete recrystallization is assumed. Otherwise, partial recrystallization condition persists. In the event of complete recrystallization, grain growth after recrystallization becomes important in determining the recrystallization kinetics of the next pass. Partial recrystallization is handled using the rule of mixtures as described in chapter 5.

7.4 Example applications of the integrated system

7.4.1 Simulation of a plate rolling process

To verify the accuracy of the system predictions, a simplified schedule for rolling a C-Mn steel plate of 20 mm thickness from a 250 mm slab using equal reductions of 15% in each pass with a 19 second inter-pass time was simulated. This rolling schedule is identical to that studied by Sellars (1979).

The change in the austenite grain size predicted during this process starting with an initial grain size of 300 microns and an initial temperature of 1180 °C is illustrated in figure 7.3. The temperature shown in this graph is the average temperature across the section of the billet. It was observed that after the initial passes when temperature is high, complete recrystallization occurs followed by grain growth. However, in the latter passes when the temperature is lower, partial recrystallization occurs. In simulating this process, the austenite evolution models for C-Mn steels proposed by Sellars, Roberts and Yada were used. The final austenite grain sizes proposed by the three models are in the range
of 25 μ - 40 μ. This agrees well with Sellars who reported a grain size measurement of about 30 microns.

![Grain size-time curve](image)

**Figure 7.3** Change in austenite grain size during the plate rolling process

### 7.4.2 Sensitivity Analysis

The integrated system enables the user to carry out a quick sensitivity analysis to see how changing certain process parameters affect the product characteristics. For example, the effect of initial temperature on the final austenite grain size is illustrated in figures 7.4 and 7.5 respectively.

To illustrate the effect of initial billet temperature, the same plate rolling sequence was simulated with an initial temperature of 1280°C instead of 1180°C. The result is presented in figure 7.4. The higher temperature in this case resulted in faster
recrystallization and significant grain growth right until the final pass resulting in coarser austenite grains (40-60 microns).

![Grain size-time curve](image)

**Figure 7.4** Grain size vs. time for an initial temperature of 1280 °C

The effect of initial grain size was examined by considering an initial grain size of 1000 microns while keeping all other conditions the same. The result of this simulation is shown in figure 7.5. It is seen that the first few passes resulted in rapid grain refinement due to recrystallization. The latter passes on the other hand were associated with recrystallization and grain growth similar to that seen in the original sequence. The result is that the final austenite grain predictions were about the same as in the original simulation. Sellars reported that the effect of initial grain size on the final austenite grain size was insignificant which agrees with the simulation results.
Figure 7.5 Austenite grain size vs. time for an initial grain size of 1000 microns
7.4.3 Simulation of interstand cooling in a shape rolling process

A six pass industrial rolling schedule from a leading steel company was chosen to study the effect of interstand cooling. The initial billet is 165 mm x 165 mm square and the material is AISI 1117. The six passes are shown in figure 7.6. The initial billet temperature is 1118°C and the distance between subsequent passes is 6.1 m. The initial grain size was assumed to be 300 mm. The simulations were conducted for two different scenarios: namely, with and without interstand cooling.

In the case without interstand cooling, complete recrystallization and grain growth was seen to occur in the interstands due to the high temperatures in the billet. To demonstrate the effect of interstand cooling, a hypothetical case with interstand cooling after passes 3 and 4 was simulated. The finite element model is designed so that the user
can specify location dependent heat transfer coefficients. This feature makes it convenient to simulate interstand cooling conditions.

Figure 7.7 (a) depicts the temperature distribution in interstand 3 using air cooling while figure 7.7 (b) shows the temperature distribution obtained with spray-mist cooling. As expected, interstand cooling resulted in significant reduction in temperatures across the billet. The effect of interstand cooling on the austenite evolution is shown in figure 7.8. From the results it can be seen that very little grain growth was observed in interstands 3 and 4 as compared to the case without interstand cooling. This lack of grain growth after recrystallization resulted in finer final austenite grains. This principle if often made use of Thermo-mechanical Control Processes (TMCP). A microstructure modeling system would enable precise prediction of the effect of process parameters on the microstructure evolution and therefore be of immense utility in the design of processes to attain the desired structure and properties in the rolled products.

The above hypothetical example demonstrates the importance of modeling interstand cooling in analyzing microstructural evolution. The integrated system has the capability to determine not only average values of austenite grain size, retained strain and fraction recrystallization at any section in the billet but also to compute the above values at every node in the interstand. Figure 7.9 shows the austenite grain size predicted by the model at the end of the final pass. This feature is important from the point of view of modeling subsequent phase transformations and prediction of mechanical properties in the rolled bar.
(a) Temperature contour plot after third pass without interstand cooling

Figure 7.7 Temperature contour plot with and without interstand cooling
(b) Temperature contour plot after third pass with interstand cooling
(a) Austenite evolution without interstand cooling

(b) Austenite evolution with interstand cooling

Figure 7.8 Effect of interstand cooling on austenite grain size
7.5 Validation studies

7.5.1 Effect on load predictions

It was mentioned in chapter 1 that the non-integrated approach used in earlier studies resulted in higher predictions of rolling loads using FEM. This was due to the fact that the flow stress model did not take into account the effect of softening due recovery and recrystallization. A seven pass rough rolling sequence from a leading steel company was chosen to study the effect of microstructure modeling on the load predictions. The roll pass sequence shown in figure 7.10 converts a 15” x 15” ingot into a 12” round bar in seven rough rolling passes. Measurements of roll loads were made on the rolling mill. The process was simulated using integrated ROLPAS first with microstructure modeling.
using a microstructure dependent flow stress model and then without microstructure modeling using a flow stress model of the following form:

\[ \sigma_f = A \cdot \sinh^{-1}\left[ B \cdot \dot{\varepsilon} \cdot \exp\left( \frac{Q}{RT} \right) \right]^n \]  

(7.1)

The bar graph shown in figure 7.11 shows a comparison between the measured and predicted rolling loads for the seven passes. The first bar in the graph represents the measured load, the middle bar represents the load predicted without microstructure modeling and the last bar is the load predicted with microstructure modeling. It is seen that the predictions of the rolling loads with microstructure modeling were within 10% of the measurements. On the other hand, the predictions without microstructure modeling were consistently much higher. This example illustrates the advantage of the integrated approach towards obtaining better predictions of rolling loads from the finite element model.

Figure 7.10 15”x15” to 12” round roll pass sequence
7.5.2 Effect on material spread

Accurate prediction of material spread is important from the point of view of designing a roll pass sequence for rolling of shapes. Experience with process modeling using FEM has shown that predictions of material spread are strongly dependent upon the flow stress model. The integrated approach allows us to model microstructural evolution and metal flow simultaneously.

Figure 7.12 shows a three pass rough rolling schedule being used in a steel company to convert a 6-5/8" x 6-5/8" square billet to a 5" diameter round billet. This
process was chosen to illustrate the effect of microstructure modeling (using an integrated approach) on the material flow. Figure 7.13 shows the deformed mesh at the exit of the rolls in the second pass as predicted by the finite element model with and without microstructure modeling. A sketch of the actual shape seen at the end of the second pass is also shown. The simulation without microstructure modeling used the steady state flow stress model developed in chapter 6 (equation 6.19) which is a function of strain rate and temperature only. It can be easily seen that the finite element model without microstructure modeling grossly under predicts the material spread. It also fails to predict the bulge profile of the workpiece. On the other hand, predictions of material spread with microstructure modeling are more accurate and the shape predicted is closer to what is seen in practice. This is due to the fact that the material at the center which is at a higher temperature compared to the material at the surface in contact with the rolls recrystallized more readily in this case thereby reducing its flow stress making it easier to flow.

Figure 7.12 Three pass breakdown sequence (6-5/8" x 6-5/8" square to 5" round)
Figure 7.13 Comparison of spread and shape predictions in the intermediate pass
7.5.3 Validation of prediction of austenite grain size

Experiments were conducted to validate the predictions of austenite grain size. The predictions from simulations were compared with actual measurements in a production environment. The measurements agreed very closely with the predictions of the integrated system. The grain size predictions were found to be within 10% of the measurements. The details of the validation experiments and the results are however not available for publication.

7.6 Modeling of Phase Transformation: Preliminary Developments

As the hot rolled product is cooled, phase transformation takes place to form transformation products such as ferrite, pearlite, bainite and martensite. Modeling of phase transformation is important from the point of view of prediction of mechanical properties in the rolled product. Preliminary work on modeling of phase transformation and linking it with ROLPAS was conducted in a parallel study (Phadke, 1997). The following write up summarizes the key features of the phase transformation module.

In the case of hypoeutectoid steels, the formation of proeutectoid ferrite, pearlite and bainite occurs by means of diffusional transformation. Diffusion is a time dependent phenomenon. The diffusional transformation depends very strongly on temperature and proceeds by a nucleation and growth mechanism. The progress of diffusional transformation is usually described by the Johnson-Mehl-Avrami (JMA) equation. The
JMA equation gives the amount of transformed product $X(t)$ at time $t$ at a fixed temperature as

$$X(t) = 1 - \exp\left[-\beta_x \left(\frac{t}{t_x}\right)^k\right]$$

(7.2)

$$\beta_x = -\ln(1-x)$$

(7.3)

where $t_x$ is the time required for transformation to a known fraction $x$, $\beta_x$ is the diffusion coefficient and $k$ is the exponent of transformation. $\beta_x$ and $k$ are functions of temperature, and represent the nucleation and growth rates, conditions of nucleation and the geometry of the growing phase. For ferrite, the exponent $k$ is assumed to be 1, while for pearlite and bainite, the exponent is computed from the IT data. Umemoto et al. have proposed a modification of the JMA equation to account for the effect of prior austenite grain size on the transformation kinetics.

The kinetics of diffusional transformation are completely described by the isothermal transformation (IT) diagram. Some form of digital representation of the various curves in the IT diagram is required to perform transformation modeling and for prediction of properties. Researchers have used various techniques of digitizing the IT diagram. Some of the commonly used techniques are: spline interpolation, function interpolation and actual measurement (Phadke, 1997). In the spline interpolation method, the curves in the IT diagram are represented by splines. On the other hand, the function interpolation method, uses a time-temperature function to represent the IT curves. The actual measurement method is the easiest to digitize and probably the most accurate. It consists of tabulating the transformation start and finish times for the various phases at
different temperatures and using interpolation for intermediate values. In the current work, this technique was used to digitize the IT diagram which includes the ferrite start curve, pearlite start and end curves and bainite start and end curves.

Unlike diffusional transformations described above, the progress of martensitic transformation is independent of time, and depends only on the temperature. The transformation of martensite has been described by a semi-empirical relation developed by Koistinen and Marburger (1959). They used precise X-ray techniques to estimate the amount of martensite formed during transformation. The amount of martensite formed $X(T)$ at a given temperature $T$ is given as:

$$X(T) = 1 - \exp[-1.10 \times 10^{-7}(M_s - T)]$$

(7.4)

where, $M_s$ denotes the martensite start temperature. Similar relations have been developed by other researchers.

The Johnson-Mehl-Avrami equation used for describing the transformation kinetics for diffusional transformation (equation 7.2) like recrystallization models is valid only under isothermal conditions. However, the transformation during cooling of the hot rolled steel product is a non-isothermal process. To adapt the isothermal transformation JMA models to a rolling process, a time discretization procedure like the one used for recrystallization (section 5.9) was used.
Figure 7.14 Integration of AUSTRANS with ROLPAS
Figure 7.14 illustrates the integration of the austenite transformation module AUSTRANS with ROLPAS. AUSTRANS uses the temperature history and microstructural details like austenite grain size, fraction recrystallized, retained strain, etc. in the last interstand to model transformation of austenite. AUSTRANS also needs as input the information about the isothermal transformation curves. Figure 7.15 shows the algorithm used for transformation modeling. At any instant $t$ during transformation, there can be up to five different phases present in the material, namely; austenite, ferrite, pearlite, bainite and martensite. Initially all the nodes are assumed to be in the austenitic state. Also, at any instant $t$, the sum of all volume fractions should be unity:
\[
\sum_{i=1}^{5} V_{i,t} = 1 \quad \text{for } t > 0
\]  
(7.5)

where $V_{i,t}$ is the volume fraction of a phase $i$ at time $t$. The interstand region is discretized into several small time steps. For each time step, the temperature is assumed to be constant at temperature
\[
\bar{T} = \frac{T_t + T_{t+\Delta t}}{2}
\]  
(7.6)

where $T_t$ is the temperature at time $t$ and $T_{t+\Delta t}$ is the temperature at time $t+\Delta t$. Using the JMA equation, the volume fractions of the ferrite, pearlite and bainite at each time step are determined. In the case of martensitic transformation, the Koistinen-Marburger equation is used. This procedure is repeated for every node until it is completely transformed or till it reaches the room temperature. After transformation modeling, the module computes the average ferrite grain size, pearlite spacing, etc. of the transformed
products and uses structure-property relationships to estimate the mechanical properties of the rolled product.

Figure 7.15 Algorithm for transformation modeling
Figure 7.16 FEM meshes in the roll bites
7.6.1 Application to a Shape Rolling Process

To demonstrate the capability of the system to model phase transformation and to predict mechanical properties, the results of simulation of an eight pass rolling schedule are presented. The simulations were conducted for AISI 1050 steel assuming an initial temperature of 2200°F and an initial reheated grain size of 300 microns. Figure 7.16 shows the 3D FEM mesh in the roll bites.

AUSTRANS first estimates the distribution of volume fractions of various transformation products. Contours showing distribution of volume fractions of austenite, pearlite, bainite and martensite are illustrated in figure 7.17. The center of the rolled bar sees higher temperatures and lower cooling rates resulting in a high percentage of pearlite. Near the surface, where the temperatures and cooling rates are intermediate, significant quantity of bainite is formed while martensite forms at the surface which experiences the highest cooling rates.

One advantage of such a system is that it allows the user to conduct a quick sensitivity analysis to see the effect of varying certain process parameters on the final properties. Figure 7.18 shows the variation in the mechanical properties along the radial direction for three different reheat temperatures. Similar results can be obtained easily by changing other parameters such as the roll pass sequence, interstand cooling, and the post rolling cooling conditions.
Figure 7.17 Contours showing distribution of volumetric fractions of (a) pearlite, (b) bainite and (c) martensite.
Figure 7.18 Variation of mechanical properties along the radial direction for three different initial temperatures
Another use of modeling microstructure evolution is to track progress of phase transformation. The progress of phase transformation for two different steels AISI 1050 and AISI 1566 at a certain node in the bar is shown in figure 7.19. The node for which these results have been plotted is indicated in the legend. AISI 1566 has a little more carbon and manganese as compared to AISI 1050 steel. The transformation kinetics are hence slower and take longer times for completion. The processing conditions simulated were:

- Cooling Rate : \(~30~{^\circ}\text{F/sec.}\)
- Reheat Temperature : \(2200~{^\circ}\text{F}\)
- No interstand cooling

Initially the steel is assumed to be composed entirely of austenite. The austenite volume fraction therefore is initially 1.00 at the start of transformation and then begins to drop as ferrite and pearlite form. When the node temperature falls below the nose of the TTT curve, bainite begins to form. At the instant when the node temperature falls below the martensite start temperature, bainite formation stops, and martensite begins to form.

It should be noted that the cooling curves in both the cases are identical. Slower transformation kinetics in case of AISI 1566 results in formation of lesser amount of pearlite and bainite as compared to AISI 1050. Also as the martensite start temperature for AISI 1566 is lower than that of AISI 1050, martensite starts forming at a later instant.
than AISI 1050. As expected, the final martensite volume fraction is higher in the case AISI 1566.

Formation of bainite and martensite is not desirable in forging grade steels. An analysis like the one presented above can give the process designer a very good estimate about what the run-out table conditions should be for different steel grades to obtain the desired microstructure in the rolled product.

![Figure 7.19 Tracking progress of phase transformation (1050 steel: top and 1566 steel: bottom).](image-url)
7.7 Summary

Microstructure models developed in the previous chapters were successfully integrated with the finite element model. The integrated system was used for simulating various industrial rolling processes to validate the predictions of rolling loads, spread and microstructure evolution. It was shown that the integrated approach to modeling the hot rolling process resulted in improved predictions of material flow and rolling loads.

This chapter also presented several ways this integrated system could be used for designing and analyzing a roll pass sequence. It is expected that a tool like this would be valuable to a roll pass designer. This system offers an inexpensive and powerful alternative to the traditional trial and error approach in roll pass design.

Some preliminary results related to modeling of post rolling phase transformation were presented. However, lot more work still needs to be done to understand better the mechanics and kinetics of phase transformation and developing models that take into account the effects of residual strain and precipitation of second phase particles on phase transformation.
CHAPTER 8

CONCLUSIONS AND FUTURE WORK

8.1 Summary and conclusions

Hot rolling is used for the manufacture of bar sections of various steel grades. These bars are used as raw material by the forging industry and other industries. For many of the applications, the structure and properties of the rolled product are important. The capability to model microstructure evolution would be of tremendous use to the rolling process designers. In this study, an FEM based integrated system for simulation of metal flow and microstructure evolution was developed.

In the first phase of this study, a 3D FEM model for the rolling process was developed and validated against various industrial rolling schedules. The second phase constituted the development of mathematical models for modeling microstructure evolution. In this part of the work, controlled laboratory tests were performed using the Gleeble thermomechanical testing machine. Broadly, the following models were developed.
• Austenite grain growth model
• Recrystallization kinetics models
• Models for recrystallized grain size
• Microstructure dependent flow stress model

Procedures were also developed for integrating these material models into the finite element model. These models were then integrated into the finite element model. Tests were conducted to validate the predictions of the integrated system. Validation tests resulted in the following conclusions:

• The integrated model gave better prediction of rolling loads
• It also resulted in better predictions of material spread
• Predictions of austenite grain sizes were found to be within 10% of measurements on rolled bars

Some work related to modeling of post rolling phase transformation was carried out. However, the work is still in a preliminary stage and lot more effort will be needed to model accurately the effects of residual strain, precipitation of second phase particles, etc. on the phase transformation kinetics. At present, there aren’t many quantitative models available in published literature that address these issues satisfactorily.
8.2 Future Work

With respect to the development of the integrated system, the work is still incomplete from the point of view of modeling phase transformation and correlation of mechanical properties. The system in its current form can still provide very useful results to a process designer. However, the end product, i.e. the final microstructure after phase transformation along with estimates of mechanical properties is what is most useful to the designer and the customer. With this in mind, future work has to be directed towards understand better the mechanism and kinetics of phase transformation and developing quantitative models for correlating mechanical properties to the microstructure.


Campbell, P.C., Hawbolt, E. B. and Brimacombe, J.K.; “Microstructural Engineering Applied to the Controlled Cooling of Steel Wire Rod: Part III, Mathematical Model -


