Modeling the Microstructural Evolution during Hot Working of C-Mn and of Nb microalloyed Steels using a Physically Based Model

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Abstract

Recrystallization kinetics, during and after hot deformation, has been investigated for decades. From these investigations several equations have been derived for describing it. The equations are often empirical or semi-empirical, i.e. they are derived for certain steel grades and are consequently only applicable to steel grades similar to these. To be able to describe the recrystallization kinetics for a variety of steel grades, more physically based models are necessary.

During rolling in hot strip mills, recrystallization enables the material to be deformed more easily and knowledge of the recrystallization kinetics is important in order to predict the required roll forces. SSAB Tunnplåt in Borlänge is a producer of low-carbon steel strips. In SSAB’s hot strip mill, rolling is conducted in a reversing roughing mill followed by a continuous finishing mill. In the reversing roughing mill the temperature is high and the inter-pass times are long. This allows for full recrystallization to occur during the inter-pass times. Due to the high temperature, the rather low strain rates and the large strains there is also a possibility for dynamic recrystallization to occur during deformation, which in turn leads to metadynamic recrystallization after deformation. In the finishing mill the temperature is lower and the inter-pass times are shorter. The lower temperature means slower recrystallization kinetics and the shorter inter-pass times could mean that there is not enough time for full recrystallization to occur. Hence, partial or no recrystallization occurs in the finishing mill, but the accumulated strain from pass to pass could lead to dynamic recrystallization and subsequently to metadynamic recrystallization.

In this work a newly developed physically based model has been used to describe the microstructural evolution of austenite. The model is based on dislocation theory where the generated dislocations during deformation provide the driving force for recrystallization. The model is built up by several submodels where the recrystallization model is one of them. The recrystallization model is based on the unified theory of continuous and discontinuous recovery, recrystallization and grain growth by Humphreys.

To verify and validate the model, rolling in the hot strip mill was modeled using process data from SSAB’s hot strip mill. In addition axisymmetric compression tests combined with relaxation was modeled using experimental results from tests conducted on a Gleeble 1500 thermomechanical simulator at Oulu University, Finland. The results show good agreement with measured data.

Keywords: austenite, modeling, hot deformation, microstructure evolution, static recrystallization, dynamic recrystallization, metadynamic recrystallization.
Preface

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Linda Lissel
Borlänge, September 2006
Thesis

This thesis consists of an introduction and the following three appended papers:

**Paper 1**
Prediction of microstructural behavior during hot rolling  
L. Lissel and G. Engberg

Conference proceeding of the 8th International Conference on Technology of Plasticity (ICTP), Verona, Italy 2005

**Paper 2**
A physically based microstructure model for predicting the microstructural evolution of a C-Mn steel during and after hot deformation  
G. Engberg and L. Lissel

Manuscript have been submitted to Steel Research International

**Paper 3**
Modeling the microstructural evolution of a Nb microalloyed steel during and after hot deformation  
L. Lissel

Manuscript to be submitted to an international journal
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**APPENDED PAPERS**
1 Introduction

During hot strip rolling of steels, the microstructural evolution that occurs in the material is dependent on the size of the reductions, the strain rate, the temperature and the length of the holding times between reductions. The amount of reduction, the rolling velocity, the temperature and the length of the inter-pass times are set for each pass in a so called rolling schedule dependent on the desired final mechanical and geometrical properties of the strip.

During hot deformation, work hardening and dynamic recovery occurs simultaneously in the work piece. The stored energy due to the accumulated dislocations is then the driving force for subsequent recrystallization, when new dislocation-free grains are formed and the energy is thereby lowered. After complete recrystallization the energy in the material can be additionally lowered by grain growth.

For several decades the microstructural evolution of austenite during hot deformation of steels has been investigated [1-7]. In order to predict the roll forces during hot rolling, it is necessary to have a correct description of the stress-strain behavior. As recrystallization causes a reduction of flow stress, knowledge of the recrystallization kinetics is essential. Consequently, several models for predicting the recrystallization kinetics have been developed over the years [1, 2, 5-7]. These models are usually empirical or semi-empirical, meaning that they are based on physical descriptions and modified to apply for certain steel grades.

There is a need for more physically based models in order to describe the microstructural evolution for a variety of steel grades. Such a model would provide the possibility to follow the microstructural evolution during hot rolling and in the end predict the mechanical properties of the finished product. Hence, it would provide the ability to design optimized rolling schedules.

The model used in this work is based on a physical description of dislocation density evolution and is developed by Engberg. The model is implemented in a toolbox for process simulations [8] and in the near future, it will cover the whole hot strip rolling process, from heating the slabs in the furnace to cooling of the strips when coiled. The desire is to be able to describe the microstructural evolution throughout the process and predict the final mechanical properties.

This thesis concentrates on the deformation part in SSAB’s hot strip mill, i.e. rolling in the roughing mill and in the finishing mill.
1.1 Aim of the work

This work is done for SSAB Tunnplåt AB and the calculations are primarily intended for the hot strip mill process at SSAB and for SSAB’s products.

The objective of this thesis is the verification and validation of Engberg’s newly developed microstructure model. The model is meant to eventually be valid for the whole range of SSAB’s products. In this investigation, the microstructural evolution of C-Mn steels and Nb steels is modeled.

The aim of this thesis is to be able to describe the microstructural evolution for C-Mn steels and Nb microalloyed steels.

1.2 The hot strip mill at SSAB Tunnplåt

SSAB Tunnplåt AB is a producer of steel strips of a variety of low-carbon steel grades. At SSAB in Borlänge, the first step in the production line is the hot strip mill. In the hot strip mill, slabs are rolled down to thin strips. The strength of the hot rolled strips is ranging from 220 to 700 MPa, i.e. from mild steels to ultra high strength steels. The wide range in strength is mainly achieved by adding different contents of alloying elements. High strength is mainly obtained by microalloying with Ti and/or Nb.

In the hot strip mill at SSAB Tunnplåt, slabs with height of ~220 mm, length of 5500 - 11000 mm and width of 650 - 1650 mm are rolled down to thin strips with thickness 1.6 - 16 mm. The hot strip mill consists of, from start to finish, reheating furnaces, descaler, reversing roughing mill, coilbox, crop shear, descaler, finishing mill with 6 stands, run out table with accelerated cooling and finally down coilers as seen in Figure 1.

Figure 1. The hot strip mill at SSAB Tunnplåt.

In the reheating furnaces the slabs are heated up to ~1200°C and a fully annealed structure is achieved. The heating temperature and the time in the furnaces are dependent on the steel grade and for microalloyed steel they are chosen in order to dissolve the right amount of particles. While heating, small to average size particles that were formed during the continuous casting of the slab are dissolved, but large particles remain undissolved. Moreover,
particles that are stable at higher temperatures than the chosen reheating temperature are of course not dissolved.

When the slab is withdrawn from the furnace, the surface is covered with oxide scale which is removed in the descaler after the furnace. Descaling is also conducted before some of the passes in the roughing mill. In the reversing roughing mill, five to seven passes are conducted and the total reduction is ~82-88%. The inter-pass times in the roughing mill are long, which depends on the fact that the bar is rolled back and forth. Usually, after rolling in the roughing mill, the transfer bar is transported down to the coilbox where it is coiled, but sometimes it is directly transferred down to the finishing mill. When the transfer bar is coiled in the coilbox the temperature along it is evened out.

During rolling in the roughing mill and during transportation, the temperature of the transfer bar decreases and before entering the finishing mill it is ~1000°C. Before entering the finishing mill, the new layer of oxide scale is removed and the irregular head and tail end are generally cut off, leaving an even edge to enter the first pass. In the finishing mill, the bar is rolled down to the desired thickness. Due to the fact that the finishing mill is a continuous process, the inter-pass times are shorter than in the roughing mill.

After rolling in the finishing mill, the strip is cooled down to desired temperature at the runout table and then it is coiled.

The temperature in the hot strip mill is only measured at a few places in the process: after the roughing mill, after the finishing mill and at the down coiler. It is only possible to measure the temperature at the surface of the strip and the temperature gradients therefore have to be calculated. Also the temperature in between measuring points has to be calculated. An example of the calculated temperature during rolling using STEELTEMP® [9] is presented in Figure 2.

![Figure 2. Calculated temperature during rolling in the roughing mill and in the finishing mill.](image-url)
The temperatures at the surface and at the centre of the work piece are shown in the figure and also the measured surface temperatures after the roughing mill and after the finishing mill. The figure show rolling of a 220 mm slab to a 4 mm thick strip. In the first passes there are large temperature gradients but these are reduced during rolling and at the end of the finishing mill, there are practically no remaining temperature gradients.

As mentioned above, after heating in the furnaces, the material is fully annealed. The dislocations are arranged in configurations of lower energy and the dislocation density is quite low. The grains are quite coarse due to the fact that the energy is lowered by the reduction in grain boundary area which is accomplished by grain growth.

During deformation, the stored energy of the material is raised by the presence of dislocations and boundaries. Simultaneously, because of the high temperature, dynamic recovery occurs. During recovery, rearrangement and annihilation of dislocations occurs which lowers the stored energy. The remaining stored energy is the driving force for subsequent recrystallization. Dynamic recovery and recrystallization are important processes because they lower the flow stress of the material enabling it to be deformed more easily. During the inter-pass times, the materials energy is lowered by static recovery and possibly recrystallization. After complete recrystallization the grains may coarsen by normal grain growth.

In the roughing mill, the temperature is high and the inter-pass times are long, allowing for full recrystallization to occur between passes. Because of the high temperature, large strain and rather low strain rate, dynamic recrystallization is likely to occur which in turn will trigger metadynamic recrystallization [10]. In the finishing mill the temperature is lower and the inter-pass times are shorter which could mean that no or partial recrystallization occurs between passes. If enough strain is accumulated between passes, dynamic recrystallization can occur. These softening mechanisms: recovery, recrystallization and grain growth are described more thoroughly below.

To get the desired microstructural evolution during hot rolling and thereby get the desired final mechanical properties, models that describe the microstructural evolution are extremely useful tools when designing rolling schedules.

### 1.3 Controlled rolling

To get the desired final mechanical properties such as a good combination of strength, fracture toughness and weldability, ThermoMechanical Processing (TMP) is utilized [11, 12]. TMP means control of the reheating temperature, the rolling schedule, the cooling rate and the coiling temperature, i.e. control of the entire processing sequence. The purpose with TMP is to obtain optimum ferrite refinement and it is therefore necessary to “maximize the area of austenite grain boundary per unit volume at the onset of phase transformation” [13]. There are different ways of achieving this and the main types of controlled rolling are recrystallization controlled rolling, conventional controlled rolling and dynamic recrystallization controlled rolling [14]. The type of controlled rolling refers to the conditions in the final rolling passes, i.e. in the finishing mill for conventional hot strip rolling. First, the as-reheated microstructure is removed in the roughing mill where complete recrystallization occurs between passes.
Recrystallization controlled rolling means that rolling is conducted above the no-recrystallization temperature, $T_{nr}$, and that full recrystallization is obtained between passes. The no-recrystallization temperature is the temperature above which static recrystallization occurs between passes [15]. Below this temperature recrystallization is retarded due to strain-induced precipitation of second-phase particles. The repeated recrystallization of austenite provides progressively refined recrystallized grains. A schematic time-temperature diagram of recrystallization controlled rolling is shown in Figure 3.

![Figure 3. Schematic time-temperature diagram of recrystallization controlled rolling [14].](image)

Conventional controlled rolling, Figure 4, is conducted below $T_{nr}$ so that pancaking of the austenite is achieved and the severely deformed austenite grains provide numerous nucleation sites.

![Figure 4. Schematic time-temperature diagram of conventional controlled rolling [14].](image)
Recrystallization is retarded by adding small amounts of elements to mild steels (microalloying). High Strength Low Alloy (HSLA) steel is the commonly used name for microalloyed steels. Microalloying generally means the addition of less than 0.1% of elements to mild steel compositions and refers to the addition of Ti, Nb or V [16]. The effect of the concentration of Ti, Nb, V and Al in a 0.07C-1.40Mn-0.25Si (wt%) steel on the recrystallization stop temperature (RST) was determined by Cuddy [17]. The RST is the temperature where complete recrystallization no longer occurs and is thus equivalent with $T_{nr}$.

In Cuddy’s investigation multi-pass deformation was conducted with a total reduction of about 65% and the result of the initial solute content on the RST is shown in Figure 5. As is seen in the figure, the effect of the precipitated Nb(C,N) provide the most effective raise of the RST. Also, recrystallization is impeded at higher temperatures with smaller addition of Nb than with the other alloys. For Ti, V and Al it is the effect of the formation of TiC, VN and AlN that is shown in the figure. Ti nitrides are very stable at high temperature and are instead effective for grain coarsening control [18, 19].

In this investigation, only the addition of Nb is concerned. When reheating the slabs the temperature is usually high enough so that Nb is in solution. During rolling, Nb forms nitrides, carbides and/or carbonitrides which effectively retards recovery and recrystallization. The highly work hardened (pancaked) austenite provides numerous nucleation sites for the ferrite which give very fine ferrite grains [3].

![Figure 5. Effect of the initial solute content on the no-recrystallization temperature, Cuddy [17].](image)

Dynamic recrystallization controlled rolling, Figure 6, means that dynamic recrystallization, DRX, is triggered in one or more passes during rolling and this in turn triggers the fast metadynamic recrystallization, MDRX. DRX is induced either by applying large single deformations or by accumulation of strain from pass to pass. In both methods the critical strain for initiation of DRX is reached.
Figure 6. Schematic time-temperature diagram of dynamic recrystallization-controlled rolling [14].

To be able to practice controlled rolling, knowledge of the microstructural evolution of the austenite during rolling is essential. In order to design optimized rolling schedules that fit the needs and constraints of each case, a microstructural model is invaluable.
2 Microstructural evolution during hot deformation

The stored energy due to the accumulated dislocations during deformation is generally lowered by three processes: recovery, recrystallization and grain growth [20]. Recovery is a process when annihilation and rearrangement of the dislocations occurs, see 2.2. Recrystallization is a process when new dislocation free grains are formed and grow on the expense of the old deformed grains, leaving a new structure with low dislocation density as a result, see 2.3. Grain growth is the process when the grains coarsen and the grain boundary area is lowered. Recovery and recrystallization can take place during and after deformation and to distinguish them they are called dynamic and static, respectively. If the recrystallization after deformation is preceded by dynamic recrystallization it is called metadynamic. Metadynamic recrystallization is the continued growth of the dynamic recrystallization nuclei present at the end of deformation [10].

The characteristic effect of work hardening, dynamic recovery and dynamic recrystallization on the stress-strain curve is illustrated in Figure 7.

![Figure 7. Schematic illustration of work hardening, dynamic recovery and dynamic recrystallization during hot deformation.](image-url)
The flow curve is dependent on the conditions of the deformation, such as temperature, \( T \), and strain rate, \( \varepsilon \), i.e. by the Zener-Hollomon parameter [12]

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

where \( Q_{\text{def}} \) is the activation energy for deformation and \( R = 8.314 \text{ J mole}^{-1}\text{K}^{-1} \).

The model used in this work describes the microstructural evolution by several submodels, which are outlined in Figure 8. The precipitation submodel is not used in this investigation and the box is therefore filled and dotted.

![Figure 8. Outline of the submodels.](image)

### 2.1 The grain structure

The kinetics of recrystallization and grain growth depends on the migration of grain boundaries. Grain boundaries are regions of considerable atomic misfit and act as strong barriers to dislocation motion [21]. A grain boundary is a boundary that separates regions of different crystallographic orientations and the misorientation between two crystals (grains) is an angle (\( \theta \)) which is the smallest rotation required to make the two crystals coincide [22], see Figure 9.
Grain boundaries are usually separated into the categories of low and high angle grain boundaries which are dependent on the size of the misorientation [22]. Low angle grain boundaries (LAGB) or subgrain boundaries are boundaries misoriented by an angle less than 10-15°. High angle grain boundaries (HAGB) are boundaries misoriented by an angle greater than 10-15°.

2.2 Dislocation generation and recovery

Most of the applied work during rolling turns into heat, only a small part remains as stored energy (~1%) [22]. The increase in stored energy is due to the accumulation of dislocations which is caused by both tangling of existing dislocations and the generation of new. Also, the energy is raised by the increase of grain boundary area. The stored energy in the material is the driving force for subsequent recovery and recrystallization.

Recovery of the material is a process that occurs prior to recrystallization and is primarily due to changes in the dislocation structure. During recovery, the dislocations rearrange in configurations of lower energy. Recovery is actually a series of events: formation of cells, annihilation of dislocations within cells, formation of low-angle subgrains and subgrain growth [22], see Figure 10. During subsequent recrystallization these subgrains function as the nucleus of recrystallization, see 2.3.

During hot deformation, dislocation accumulation due to deformation and annihilation and rearrangement of dislocations due to dynamic recovery occurs simultaneously. The remaining stored energy that is built up by the dislocation accumulation is the driving force for the possible subsequent recrystallization.

If dynamic recovery is the only form of restoration that occurs in the material, the flow-stress in a stress-strain curve reaches a plateau and then holds a steady-state flow-stress, see Figure 7. This depends on the fact that the rate of recovery and work hardening reaches a dynamic equilibrium.
The flow stress during deformation is dependent on the dislocation density $\rho$ and is usually described by

$$\sigma = \sigma_0 + m \cdot \alpha \cdot G \cdot b \cdot \sqrt[3]{\rho}$$

(2)

See for example the work by Bergström [23]. In the equation, $\sigma_0$ is a constant, mainly due to the strengthening contribution due to precipitation, $m$ is the Taylor factor and is dependent on the deformation due to the development of a deformation texture [13], see Figure 11, and $\alpha$ is a proportionality constant (~0.5). $G$ is the shear modulus and $b$ is the Burgers vector.

![Figure 10. The stages of recovery [22].](image)

(a) Dislocation tangles  
(b) Cell formation  
(c) Annihilation of dislocations within cells

(d) Subgrain formation  
(e) Subgrain growth

![Figure 11. Taylor factor as a function of applied strain for uniaxial compression of a fcc metal [13].](image)
The evolution of dislocations during hot deformation can be separated in two parts – dislocation generation and dislocation recovery

\[
\frac{d\rho}{dt} = \frac{d\rho_{\text{gen}}}{dt} - \frac{d\rho_{\text{recovery}}}{dt}
\]

(3)

A detailed description of this is given in Paper 2.

### 2.3 Recrystallization and grain growth

The definition of recrystallization according to Doherty et al. [20] is “the formation of a new grain structure in a deformed material by the formation and migration of high angle grain boundaries driven by the stored energy of deformation”. Consequently, recrystallization is a process when new dislocation-free grains grow on the expense of the old deformed grains, leaving a new structure with low dislocation density as a result. The process is divided into nucleation of new grains and growth of the same. It is nowadays acknowledged that “nucleation” of recrystallization cannot be described with classical nucleation theory [22]. Recrystallization is a much more rapid process and the new grains grow from small regions already present in the structure, like subgrains or cells. The subgrains must have an energy advantage, a larger size, to be able to grow, rather than shrink and vanish. Hence, for the recrystallization to take place, a critical subgrain size has to be reached. This critical size is large in comparison with the mean subgrain size, which means that it can only be reached by abnormal growth.

According to Humphreys, [24], no evidence is found that rotation plays an important role in subgrain coarsening and it is a reasonable assumption that both subgrain coarsening and nucleation of recrystallization from pre-existing subgrains are controlled by the migration of low angle grain boundaries. Subgrain coarsening in regions of large orientation gradient is believed to be a method by which recrystallization originates. When an orientation gradient is present in the deformed and recovered material, the subgrains will grow more rapidly than those in an environment where there is no orientation gradient. Faster growing subgrains obtain more misorientation, which results in the creation of a high angle grain boundary and hence a nucleus.

Consequently, for recrystallization to take place, a critical subgrain size has to be reached and the misorientation has to be high enough, i.e. it has to be a high angle grain boundary. Preferable nucleation sites for recrystallization is grain boundaries [22, 25], but when the grains are very large, intragranular nucleation, not deformed at the old deformed austenite grain boundaries, occurs as well [22], see Figure 12.

The recrystallized grain size is reduced by larger deformations which introduces smaller subgrains and thereby increases the number of nucleation sites for recrystallization [5, 26]. This effect becomes less evident with increasing strain and ceases at some critical strain. The recrystallized grain size is also reduced by finer initial grain sizes which increase the number of potential nucleation sites and hence decreases the recrystallized grain size [5, 26]. There is also a strong temperature dependence on the recrystallized grain size. At lower temperature, finer recrystallized grains is obtained due to the lower mobility of recrystallizing grain boundaries and consequently slower growth rate of new grains [26]. The slower growth rate
promotes additional nucleation during recrystallization which consequently gives finer recrystallized gains. As the initial grain size is decreased, this effect becomes less significant.

Figure 12. A schematic picture of the development of recrystallizing grains, where in a-d the initial grain size is large and in e it is small. The dotted lines indicate prior grain boundaries [26].

The nucleation and growth of recrystallizing grains give a sigmoidal shape of the recrystallization curve when fraction recrystallization is plotted as a function of log(time). A typical recrystallization curve is illustrated in Figure 13.

Figure 13. Typical form of a recrystallization curve at isothermal conditions.

Recovery is a thermally activated process and the temperature dependence for static recrystallization is illustrated in Figure 14.
Figure 14. Recrystallization curves for a plain C-Mn steel pre-strained to 0.2 at a strain rate of 1 s⁻¹ deformed and annealed at 850°C-1200°C.

If dynamic recrystallization occurs, the flow-stress raises to a peak value followed by a lower steady-state flow-stress, see Figure 7. Dynamic recrystallization is promoted by low strain rates and high temperature and if the strain rate is low enough the flow stress does not reach a steady-state flow-stress after the peak, but oscillates around a certain value due to successive cycles of recrystallization occurring concurrently with the deformation [10]. At higher strain rates there is a single peak behavior of dynamic recrystallization.

For dynamic recrystallization to take place, a critical strain, ε_c, has to be reached. Some low fraction of recrystallization takes place before reaching the maximum flow stress at the peak strain, ε_p [5]. Hence, the peak strain in the stress-strain curve is always greater than the critical strain required to initiate dynamic recrystallization. According to Sellars [5] the relation between the peak strain and critical strain is ε_c = 0.8·ε_p. For strains below 0.8·ε_p there is a significant strain dependence on the time for 50% recrystallization, t_{0.5}.

Static recrystallization occurs after a deformation below the critical strain and is dependent on the applied strain. Metadynamic recrystallization, which occurs after a deformation above the critical strain, is strain independent. Static recrystallization requires an incubation time, when the nuclei forms, while the nuclei for metadynamic recrystallization is already present at the end of deformation [10].

According to Jonas [27], the softening that takes place after dynamic recrystallization does not provide full softening of the material due to the lightly worked portions in the dynamically recrystallized structure. The metadynamic recrystallization provides ~15% softening and is pursued by static recovery, which in turn supplies additional 10% softening. The recovery provide recrystallization nuclei and static recrystallization (involving incubation time) follow, which gives about 55% softening. In the roughing range of hot strip rolling where the
temperature is high, both metadynamic and static recrystallization are likely to occur very rapidly after the deformation and full softening is expected. Conversely, in the finishing stands when the temperature is lower, metadynamic recrystallization is still likely to occur followed by minor recovery, and the static recrystallization that requires substantial recovery and an incubation time is unlikely to have any influence.

After complete recrystallization, the energy in the material may be further reduced by grain growth. Grain growth is driven by the decrease in energy per unit volume which is accomplished by the reduction of grain boundary area and by the increase of average grain boundary curvature [26, 28].
3 Modeling recrystallization

The recrystallization model used in this work is based on Humphreys’ model and is described in paper 2.

During recrystallization, growth and coarsening of grains happen concurrently. In Engberg’s model, the recrystallizing grains are treated as two classes: surface grains and interior grains, which represent growing and coarsening grains, respectively, see Figure 15. Figure a, shows the old deformed grain boundary with newly developed grains growing along it. Figure b also shows coarsening of one of the new grains and a new grain growing on the “reduced” old grain boundary.

Figure 15. A schematic picture of deformed material (gray) where new grains (white) grow and coarsen. The dotted lines indicate old grain boundaries, the solid lines indicate growing grains and the dashed lines indicate coarsening of a new grain. In a, the new recrystallized grains are shown and also a representation of a reduced deformed grain boundary where new grains can nucleate. In b, coarsening of a new grain is presented and also a growing grain at the reduced boundary.

The pinning force by second-phase particles is described by Zener pinning in the model and for the material to recrystallize, the driving force for recrystallization has to be larger than the pinning force by precipitates.
3.1 Empirical models

A characteristic curve of the recrystallization kinetics is presented in Figure 13 and this shows the typical sigmoidal form. This sigmoidal form has been proven to be well described by a modified Johnson-Mehl-Avrami-Kolmogorov (JMAK) equation

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

where \( X \) is fraction recrystallized in time \( t \) and \( n \) is a constant.

The static recrystallization kinetics expressed as the time to reach a certain level of recrystallization depends on chemical composition, applied strain (\( \varepsilon \)), strain rate (\( \dot{\varepsilon} \)), initial grain size (\( d_0 \)) and temperature (\( T \)) [1, 5]. Sellars [5] derived an equation for describing the time for 50% static recrystallization including strain, strain rate, composition, grain size and temperature

\[ t_{0.5} = B \varepsilon d_0^p Z^q \exp \left( \frac{Q_{\text{rex}}}{RT} \right) \]  

where \( B, p, q \) and \( s \) are constants dependent on the chemical composition of the steel and \( Q_{\text{rex}} \) is the activation energy of recrystallization.

Metadynamic recrystallization kinetics is independent on the strain and initial grain size and according to Sellars [5], the time to reach 50% recrystallization is given by

\[ t_{0.5} = B' Z^r \exp \left( \frac{Q_{\text{rex}}}{RT} \right) \]  

where \( B' \) and \( r \) are constants dependent on the chemical composition of the steel.

The condition for metadynamic recrystallization to occur is that the critical strain for dynamic recrystallization was reached during deformation [5]. The critical strain is proportional to the peak strain (~0.8 \( \varepsilon_p \)), which depends on the initial grain size and the strain rate and follows the relationship

\[ \varepsilon_p = A d_0^{1/2} Z^n \]  

where \( A \) and \( m \) are constants dependent on the chemical composition.

Hodgson [1] derived an equation for the time to 50% recrystallization where the effects of solute drag of Nb is taken into account

\[ t_{0.5} = \left( -5.24 + 550[Nb] \right) \cdot 10^{-18} \varepsilon^{-4.0+77[Nb]} d_0^2 \exp \left( \frac{330,000}{RT} \right) \]  

where \([Nb]\) is the Nb in wt%. The equation is valid for Nb contents of 0.013-0.030 wt.%.

When the effect of precipitates is modeled it is often assumed that once the time for 5% precipitation is reached there will be no static recrystallization [1, 14]. In Paper 3 where the effect of the precipitates on recrystallization is calculated, the fraction and size of the particles are set to constant values when precipitation is predicted to occur.
4 Experimental procedure

The recrystallization kinetics can be determined by metallographic examination of quenched specimens, by using double compression tests or by using stress relaxation tests. The stress relaxation technique is used in this research since it is the most efficient way to measure reliable recrystallization kinetics. Metallographic determination of the recrystallization kinetics is tedious and quite difficult because of the phase transformation. Determination of the recrystallization kinetics by double compression tests means one tests per dot on the curve and numerous tests has to be done.

To determine the material behavior during hot compression, several compression and stress relaxation tests were carried out at Oulu University, Finland. The tests were performed on a Gleeble 1500 thermomechanical simulator.

The stress relaxation test is described by Karjalainen et al [31, 32], where both the analysis technique and the characteristics of the test is described. The technique used to analyse the relaxation curve is also described in paper 3. Karjalainen showed that the stress relaxation technique is applicable to several steel grades. The technique can measure both static and metadynamic recrystallization.

The test samples in this investigation have diameter 10 mm and height 12 mm and were machined from the head end of transfer bars after rolling in the roughing mill in SSAB’s hot strip mill.

The preparation of a sample is shown in Figure 16. The anvils are made of tungsten carbide and three foils are used in the tests: a graphite foil for friction, a Ni-foil for preventing carbon diffusion into the specimen at the high temperatures and a Ta foil to prevent sticking to the anvils. The sample is heated by resistance. In the picture, the thermocouples attached to the sample for measuring and controlling the temperature of the sample is shown. Also, the spray devices for quenching the samples are shown.

Figure 16. Preparation of a sample in the Gleeble machine: at room temperature and at 1200°C.
Due to the fact that there is a phase transformation when quenching the samples, the old austenite grains are not easy to detect in the microstructure. Preferably, martensite should be formed when quenching the samples, in order to freeze the austenite structure. In these tests, the quenching method was insufficient and ferrite transformation has started on the former austenite grain boundaries. Hence, there were some problems in determining the old austenite boundaries, but a good estimation of the prior austenite grain size could be done.
5 Summary of appended papers

5.1 Paper 1

In this paper Engberg’s microstructure model was used to describe the microstructural evolution of C-Mn steels during hot rolling in SSAB’s hot strip mill.

The submodel describing the flow stress had been verified in an earlier study [13] and here, only the recrystallization model was considered. The recrystallization model was first validated using experimental data from Medina et al. [33], where good agreement was obtained, see Figure 17.

![Figure 17](image)

**Figure 17. Calculated fraction recrystallized with Engberg’s model (lines) compared to experimental results (dots) [33].**

The fraction recrystallized during hot rolling calculated with Engberg’s model was compared to the fractions calculated with an empirical model which is proven to give good results according to Sicilano et al. [34]. The results showed some minor differences. Both models predict full recrystallization in the roughing mill. In the finishing mill the empirical model predicts full recrystallization after the first two passes and thereafter partial. Engberg’s model predicts partial recrystallization in the entire finishing mill, although it is nearly complete after the first two passes.

The calculated flow stress during hot rolling was verified using process data collected from SSAB’s hot strip mill. The verification was done by comparing the mean flow stress from the
microstructure model with the mean flow stress calculated with a friction-hill roll-force model. The development of the mean flow stress during the rolling process is shown in Figure 18.

![Figure 18. Mean Flow stresses (MFS) for each pass in the hot strip mill calculated with the friction-hill model and the microstructure model.](image)

The agreement with the friction-hill model in the roughing mill is good, but in the finishing mill it is less accurate. One explanation could be the uncertainty in the calculated temperatures, where the calculations can only be adapted to the measured strip surface temperatures.

In paper 1, the coefficient of friction is never discussed, but surely it affects the results. When calculating the mean flow stresses with the friction-hill model, the coefficient of friction was set to a constant value of 0.3. In the roughing mill the roll torque can be determined and with both the roll torques and the roll forces known, the coefficient of friction could be defined. The calculated coefficient of friction in the roughing mill was ~0.3 and hence this value was used. In the finishing mill, the roll torque could not be accurately determined and the coefficient of friction had to be estimated. It is possible that the coefficient of friction in the finishing mill varies from stand to stand, but because this could not be verified, the same value as in the roughing mill was used. There is a possibility that the mean flow stresses calculated with the friction-hill model in the finishing mill are too low and that the agreement with measured data is in fact better than what is illustrated in Figure 18.

It was made clear from this investigation that the hot rolling process is too complex to use as “an experimental tool” to verify and validate the model. Controlled experiments, where all the data is confirmed are needed.
5.2 Paper 2

In this paper, the model was used to describe experimental data conducted on a Gleeble 1500 thermomechanical simulator at Oulu University, Finland. With the intention of verifying that the model can describe dynamic, static and metadynamic recrystallization, compression tests combined with stress relaxation were performed at various temperatures, strains and strain rates, with the following experimental parameters:

- Reheating rate: 4°C/s
- Reheating temperature: 1200°C
- Soaking time: 180 s
- Cooling rate to deformation temperature: 10°C/s
- Holding time before deformation: 15 s
- Deformation temperature: 850°C - 1200°C
- Applied strain in each hit: 0.2-0.6
- Strain rate: 0.1 – 10 s⁻¹

The same set-up of parameters was used to describe the microstructural evolution for all the tests. For the compression tests, the agreement between the experimental and calculated values is good, but the model does not accurately describe the shape of the curves, see Figure 19 to Figure 21.

![Figure 19](image1.png)

**Figure 19.** Calculated and experimental stress-strain curves at various temperatures at a strain rate of 0.1 s⁻¹.

![Figure 20](image2.png)

**Figure 20.** Calculated and experimental stress-strain curves at various temperatures at a strain rate of 1 s⁻¹.
When dynamic recrystallization occurs in the calculations the material is divided into one deforming and one recrystallizing part. This results in oscillations of the curves. Also, in the initial stage of deformation the agreement with experimental data is rather poor, especially at higher strain rates.

Calculations of the stress relaxation tests show good agreement with experimental data, see for example Figure 22.

For almost every stress relaxation test at strain rate $1 \text{ s}^{-1}$, temperatures from 850°C - 1200°C and strains of 0.2, 0.4 and 0.6, the model accurately describes the recrystallization kinetics. Also at strain rate $0.1 \text{ s}^{-1}$ at temperatures 1000°C and below, the model give good agreement with experimental values. There are some problems describing the relaxation when substantial dynamic recrystallization has occurred during the deformation. Hence, at strain rate $0.1 \text{ s}^{-1}$ at temperature 1100°C and above, the model can not reproduce the experimental flow stress during relaxation.

A validation of the model was done by calculating a multi-step test where good agreement with flow stress values and grain sizes was obtained, see Figure 23. The outline of the multi-step test was arranged to simulate the hot rolling process. The first reduction was large in order to quickly reduce the grain size. It was then followed by a long holding time in order to
get a fully recrystallized structure before the next reduction. The succeeding reductions were followed by shorter holding times.

Figure 23. Experimental and calculated average grain size (a) and flow stress (b) at four subsequent deformations with different holding times in between.
On the whole, the calculations show good result which is very promising for the intention of predicting the microstructural evolution during hot rolling. The next step was to verify that the model can describe the microstructural evolution for microalloyed steels.

5.3 Paper 3

In this paper, the model was used to describe the microstructural behavior of a Nb microalloyed steel and compare the results with the plain C-Mn steel used in paper 1. Hence, further compression tests combined with stress relaxation was conducted on the Gleeble 1500 thermomechanical simulator at Oulu University, Finland. The experimental parameters are the same as those described in paper 2 above.

Because of Nb in solid solution, the parameters controlling the mobility of dislocations for recovery and the growth rate of both subgrains and of recrystallizing grains had to be reduced. By only changing these three parameters, the microstructural evolution for the Nb steel could be described with reasonable results.

The calculated flow stress for the compression tests is shown in Figure 24 to Figure 26. As is seen in the figures, the experimental data is quite well reproduced by the model with the exception of the oscillations in the curves, as in paper 2. Also, the agreement between the measured and the calculated flow stress in the initial stage of the deformation is still rather poor but it has been improved compared to in paper 2. The hardening at higher strain rates is better described than in paper 2, but instead, at lower strain rates the hardening is too strong initially.

![Figure 24. Calculated and experimental flow-stress curves at strain rate 0.1 s⁻¹.](image)
At 1000°C and above, when the Nb is in solution, the calculated results of the relaxation tests compared to experimental data show good results. At the final stage of the relaxation test, the experimentally higher final stress for the Nb steel is partly described by using a lower mobility than for the C-Mn steel as seen in Figure 27.

The effect of the Nb carbonitrides that are precipitated during relaxation at 900°C is accounted for by Zener pinning and the start of precipitation of niobium carbonitrides was calculated with the equation of Dutta and Sellars [35]. To model the retardation due to second-phase particles, a volume fraction \( f \) and a radius \( r \) of the precipitated particles was set in the input data at the calculated time for the start of precipitation. After recrystallization has been retarded the stress is still lowered by recovery but the model predicts too strong recovery and lowers the stress too much. This is due to the fact that the pinning force on recovery is best described for the initial part after unloading in the model. It might also be explained by the fact that the ratio \( f/r \) was set to a constant value once precipitation was started.
Figure 27. Experimental and calculated relaxation curves at 900°C and 1000°C for both the C-Mn steel and the Nb steel. Figure a, shows the results after a strain of 0.2 at a strain rate of 0.1 s\(^{-1}\) and figure b, shows the results after a strain of 0.6 at a strain rate of 1 s\(^{-1}\).

In Figure 28 the experimental and calculated flow stress and grain size for a double compression test are presented. The calculated results show good correlation with the experiment.

Figure 28. Results for the double compression test. Figure a shows experimental and calculated flow stress and figure b shows measured and calculated grain size.

The model presents a good description of the microstructural evolution for the conditions when recrystallization is only retarded by Nb in solid solution. When Nb(C,N) are precipitated the description of the microstructural evolution is not so good. As a next step, a precipitation model is to be used and this will probably give better fit also at lower temperatures.

On the whole, Engberg’s model shows good agreement with experimental data. The model has now been proven to be able to describe the microstructural evolution for both plain C-Mn steels and Nb microalloyed steels.
6 Conclusions and future work

In the first paper it was concluded that the recrystallization model developed by Engberg provides good agreement with experimental data for recrystallization. The calculations of the mean flow stress for process data in the roughing mill show good agreement with the calculated mean flow stress using a friction-hill model. The recrystallization during rolling gives similar values as an acknowledged empirical model. Still some modifications have to be made to get better agreement for the finishing mill. It was also concluded that the results were promising for the purpose of modeling the microstructural evolution in hot strip mills, but also that the hot rolling process is too complex to use to get a good set-up of parameters.

In the second paper it was concluded that the model provides reasonable agreement with experimental data for all the temperatures and strain rates. The model does not accurately describe the shape of the curves, especially for small strains and low strain rates but the agreement between experimental and calculated values is good. Also, the calculated relaxation curves provided good fit with experimental data. At some cases when substantial dynamic recrystallization has occurred during the deformation, there were some problems describing the relaxation. Also, the calculations for the multi-step test show good agreement with both flow-stress values and grain sizes and on the whole, the results give very promising results for the intention of predicting the microstructural evolution during hot rolling.

In the third paper it was concluded that only by changing the parameters describing subgrain growth, recrystallizing grain growth and the mobility for recovery of dislocations, because of the effect of solute drag, the model can quite well describe the flow stress during compression and relaxation for Nb microalloyed steels. For the compression test the calculated flow stress was in good agreement with experimental data at all strains, strain rates and temperatures. The recrystallization kinetics was well described at temperature 1000°C and above, but at 900°C when recrystallization is retarded due to precipitated second-phase particles the recovery is too strong. For the conditions when the Nb is still in solution, relaxation is well described by the model. Modeling of the double compression test showed good agreement with both experimental flow stress and measured grain size.

Now, as a first assumption, the ratio of the volume fraction precipitates and of the particle radius was determined and set to constant values at the time for start of precipitation. In the near future a model for the precipitation kinetics will be used and the future work with the model will be to calculate the microstructural evolution for more highly microalloyed steels. The model will also be implemented in a commercial FEM-program.
7 References

18. Siwecki T., Grain Coarsening Behaviour, Microstructure Development During Hot Rolling and As-Rolled Strength in Ti--V and Ti--V--Nb Microalloyed Steels,


