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MODELLING OF SOFTENING IN NIOBIUM-HSLA STEELS DURING HOT ROLLING

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Abstract

Niobium HSLA steels are more difficult to hot roll with consistent mechanical properties than plain carbon steels. In this paper the causes of this different behavior are explained and illustrated with a recently developed physical model. Advantages, restrictions, applications and future developments of this model are discussed.

The Hot Rolling of HLSA Steels

Microalloyed or HSLA (High Strength Low Alloy) steels find many specialized applications, such as in the automotive and construction industry. They are important products for the steel industry, because these steels provide a higher added value to the customer.

However, while hot rolling of plain carbon steels is a relatively straightforward task, hot rolling of HSLA steels is much more a challenge. For HSLA steels it is more difficult to get products with consistent properties. Consistent properties are important, though, especially with the introduction of new fabrication techniques such as laser welding and hydro forming [1].

For a better control of the hot rolling production process of HSLA steels a model describing the appearing processes quantitatively is needed. To improve the modeling of hot rolling of HSLA steels three European steel companies and one research institute (Thyssen Krupp Stahl, Arcelor, CEIT and Corus) worked together in the ECSC sponsored project CAMSIP (Characterization And Modeling of Strain Induced Precipitation) during the period 2001-2005. Aims of this project were collecting data on model alloys, comparing different measuring methods for the characterization of softening and precipitation, and testing models published in literature. A synopsis of the experimental work in this project is presented in [2]. This paper will focus on the model of Zurob and co-workers, describing the complex interaction of recovery, recrystallization and precipitation. Testing and implementing this model was another activity of the CAMSIP project. In this paper a concise discussion of this model is given, and the advantages and shortcomings are discussed.

The difficulties in the process control for HSLA steels are caused to a great extent by the complex metallurgical processes at high temperatures in these steels. During hot rolling, precipitation, recovery and recrystallization can interact. Therefore, the softening in HSLA steels after deformation evolves differently from for instance CMn steels, where only recovery and recrystallization take place.

Titanium, niobium and to a lesser extent vanadium form precipitates with nitrogen and carbon (and in some cases also with sulphur and boron) at hot rolling temperatures. Precipitation influences the hot rolling process in two ways: first it can increase the rolling loads due to

precipitation hardening. However, the precipitation has an even larger influence on the recovery and recrystallization of the material. Due to recrystallization the microstructure of a deformed steel is changed drastically by the growth of new grains. The precipitates, however, pin the dislocations (which were formed in the rolling process) and pin the grain boundaries of the new, recrystallized grains. This pinning force can be so large that below a certain temperature, the growth of recrystallized grains stops and so does the recrystallization.

Partial recrystallization will also have consequences for the austenite-ferrite transformation. If the microstructure is partly recrystallized at the entrance of the run out table the stored energy in the deformed grains will accelerate the nucleation of the ferrite grains resulting in a fine ferrite microstructure. Furthermore, the elongated form of the unrecrystallized austenite grains will influence the microstructure of the ferrite after transformation.

The metallurgical processes and their interactions strongly depend on the chemical composition and process parameters as temperature, strain and strain rate. This makes the changes in the microstructure complex and the process control of the hot rolling of HSLA steels difficult. To get a better grip on the rolling process, it is necessary to develop a better understanding of the microstructural processes and to describe them quantitatively in a model. An important class of such models is empirical relations [3,4]. These relations have the advantages of easy use and accuracy for the materials and process conditions for which they were tested. But at the same time this is a restriction of empirical relations, because extrapolation outside the material range tested or range of process parameters can give very inaccurate results. Furthermore, empirical relations do not aid to the understanding of the mechanisms.

A model based on a physical description of the relevant phenomena, will largely circumvent these drawbacks. Such a model is the subject of this paper.

A model for the interaction between precipitation, recovery and recrystallization

In the period 2001-2003 Zurob, Hutchinson, Brechet and Purdy published a number of papers in which they developed a physical model for strain-induced precipitation [5-7]. In this model recovery, recrystallization and precipitation are described with a specified submodel for each phenomenon. Each submodel is based on a physical description of the relevant phenomenon. Most of these submodels have already been published. For recovery, for instance, they used the model constructed by Verdier et al. for recovery in aluminum [8] and for precipitation with nucleation on dislocations, a model for aluminum developed by Deschampes and Brechet [9], was applied. Recrystallization has been modeled using a Johnson-Mehl-Avrami-Kolmogorov approach. In the model for the recrystallization the solute drag phenomenon, the drag exerted on moving grain boundaries by solutes, is also taken into account.

However, the interesting feature of this model are the couplings between those submodels, in other words, the different ways the three submodels can influence each other. An overview of these couplings is given in figure 1. A few examples of these couplings will be discussed in the following.

The dislocation density, created during deformation in the rolling mill, plays an important role in many of these couplings. It decreases as a result of recovery and recrystallization. However, the dislocation density also determines the total number of available nucleation sites for precipitation, because it is assumed that precipitates nucleate at dislocations. Moreover, the dislocation density is the driving force for recrystallization. On the other hand, precipitates can pin the dislocations in the submodel for recovery: if the precipitate number density is high enough, recovery can stop. Dislocations also play a role in the diffusion, especially of niobium.

Diffusion along a dislocation ('pipe diffusion') is much faster than diffusion in the bulk. Thus, a higher dislocation density leads to a larger, effective diffusion of niobium in the grains.

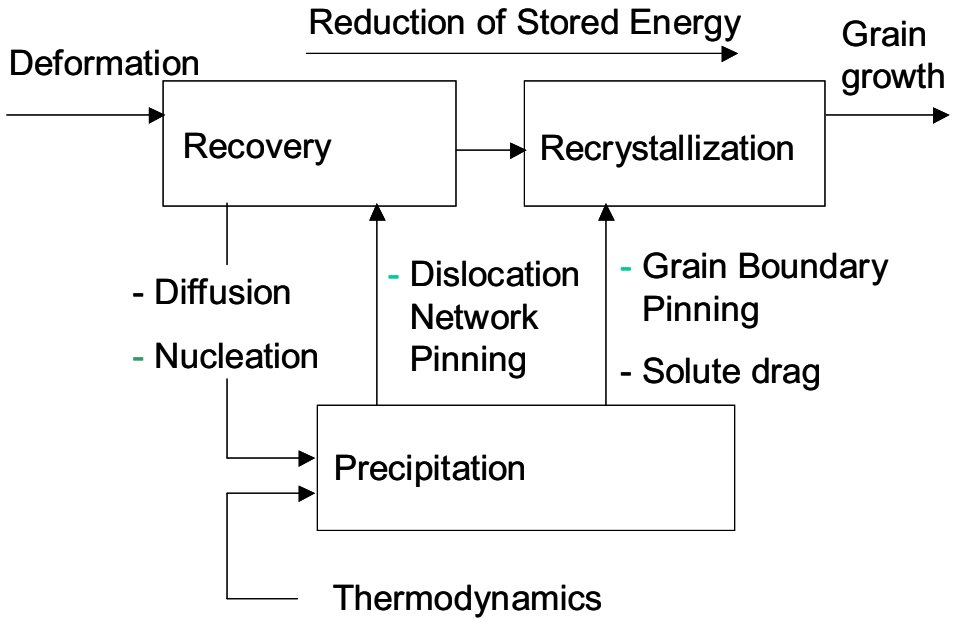

Figure 1. Scheme of the model, showing the processes in the model and the interactions between the processes.

Another example of the mutual influence of the submodels is the role of niobium. Niobium in solution has the strong solute drag effect: it restricts the movement of the grain boundaries of recrystallized grains. However, the concentration of dissolved niobium decreases due to the binding of niobium in precipitates, and, hence, precipitation reduces the solute drag effect. On the other hand, precipitates cannot only pin dislocations (in the recovery submodel) but can also pin the grain boundaries of recrystallized grains. This Zener pinning depends on the volume fraction of the precipitates and the average precipitate size. These are just two examples of the couplings between the submodels. A detailed discussion of all couplings is given in [10].

First simulation results

The model of Zurob has been coded in a computer program, in which the evolution equations, describing the submodels, are solved numerically. A simulation with this program takes a few minutes. Because of these relatively short calculation times the program is also very suitable for optimization and for fitting of experimental data.

With this program, simulations of experiments have been made, which were published in the literature. The Zurob model, however, still contains two parameters, values of which cannot be predicted a priori yet. One of these parameters determines the nucleation rate for precipitation; the other the nucleation density for recrystallization. Values for these parameters were therefore obtained by a least square fitting of the model to experimental data of the softening for each composition and temperature.

Table I. Chemical composition (in wt%) of the steel used for the simulations of Figures 2 and 3 [11].

C	Si	Mn	P	S	Nb	Al	N
0.076	0.06	1.34	0.0058	0.0026	0.030	-	0.0061

Experimental data for strain induced precipitation are scarce, particularly experiments in which both softening and precipitate diameters have been measured. Simulations were carried out using experimental data obtained by Kang and co-workers [11]. The results of these experiments were also used by Zurob et al [6,7]. Kang and co-workers measured both softening (using the double hit method) and precipitate diameters [11]. The composition of the steel used by Kang, is given in Table I.

The predicted softening and recrystallization curves at 850 °C are shown in figure 2. Softening was measured after deformation with a strain of 0.3 at a strain rate of 3.63 s^{-1}. As shown, the model can represent the experimental softening data quite well. The model prediction for the recrystallized fraction is also given in Figure 2, but for lack of experimental data no verification is possible. Note that the recrystallization only starts after 1000 seconds; for a steel with the same composition, but without niobium a typical recrystallization time would be around one second.

The values of the model parameters obtained by fitting the softening data were also used to predict the precipitate diameters. Figure 3 shows this model prediction together with the experimental data. The relation is good, especially considering that the values for the model parameters were obtained from another type of experiment. In general the model predictions of the softening and precipitate diameters show a high degree of consistency for this temperature.

One advantage of the model is the wealth of information obtained in addition to the prediction of softening. Because the model also predicts the evolution of the dislocation density it allows separating the contributions of recovery, recrystallization and precipitation hardening to the softening. Separating recovery and recrystallization by experimental techniques is difficult and often impossible. In addition to the dislocation density, the model also predicts the volume fraction of precipitates, precipitate number density and the concentrations of niobium, carbon and nitrogen in the matrix. Niobium concentration, precipitate diameters and volume fraction are important for the softening behavior as well as playing a role in the austenite-ferrite phase transformation. Dissolved niobium and fine precipitates will slow down the transformation kinetics by the solute drag effect and the Zener pinning effect.

A similar agreement between experimental data and model predictions is also obtained for 900 °C for the same material and deformation strain and strain rate. However, for temperatures above 900°C the model predictions for the softening and precipitate diameter are much less accurate.

This result is also found for other compositions: below 900 °C the model gives good predictions for the experimental data, but the accuracy is lower for temperatures higher than 900 °C. This discrepancy between the model predictions and measurements is subject of further study. One suggestion is to re-assess the simplifications in this - already quite complex - model. A more accurate description of (at least one of) the phenomena may be able to improve the results above 900 °C. One candidate is the evolution equation for the recovery.

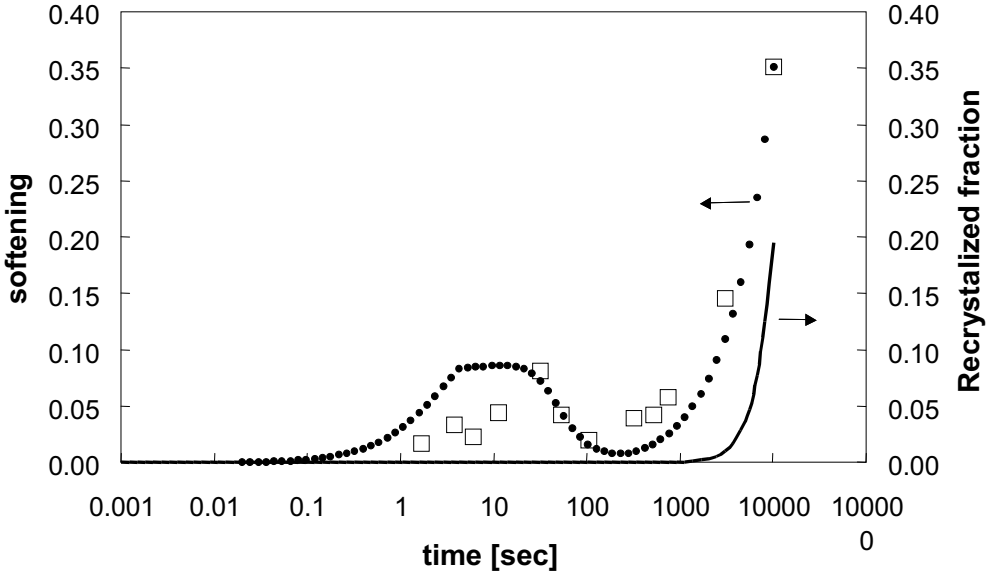

Figure 2. Softening vs. time. Simulation results (dotted line) compared to experimental data (squares) [10] for softening and simulation results for the recrystallized fraction (solid line). Alloy composition is given in Table I, temperature: 850 °C, strain: 0.3, strain rate: 10 s^{-1}.

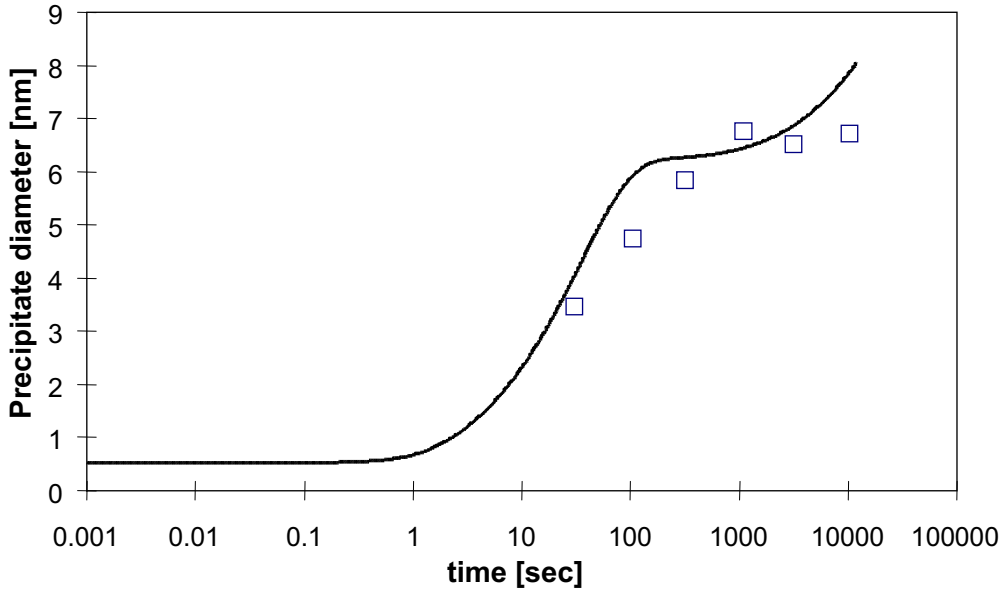

Figure 3. Average precipitate diameter vs. time. Simulation results (solid line) compared to experimental data (squares) [11]. Alloy composition is given in Table I, temperature: 850 °C, strain: 0.3, strain rate: 10 s^{-1}.

Behaviour of HSLA Steels in the Hot Rolling Mill

The control of the production of Nb-HSLA steels is much more difficult than for CMn-alloys. With the results of Zurob's model this can be illustrated. The cause of the difficulty lies in the different phenomena that make up the softening process. These are recovery and recrystallization on one hand, reducing the flow stress and promoting the softening, and precipitation on the other hand, increasing the flow stress and retarding the softening. The softening kinetics depends on the chemical composition, deformation parameters and temperature. The softening curves may contain no plateau (figure 4a), one plateau (figure 4b), a 'hump' (figure 2 shows this type of

curve) and even two plateaus (figure 4c). The type of curve found in an experiment at constant temperature depends on the chemical composition, strain, strain rate and temperature.

At high temperature only a sigmoidal curve (figure 4a) is observed. At these temperatures nucleation and growth of recrystallized grains occur fast and precipitate nucleation is slow (due to a low driving force). Thus, recrystallization is not hindered by precipitation, which leads to a 'normal' recrystallization curve. The kinetics of the recrystallization will only be lowered by the solute drag effect of niobium.

However, when the temperature is decreased the kinetics of recrystallization and precipitation become comparable and the interaction between both phenomena gains influence. When recrystallization is strongly retarded (due to the solute drag and Zener pinning effects) recovery is the only remaining process for softening. However, the recovery process as such can be hampered by the pinning of the dislocations because of precipitates, nucleating on these dislocations. This hindrance of the recovery causes the plateau in the softening curve. In course of time the precipitates will grow and coarsen. Coarsening leads to a decrease in the precipitate density and thus to depinning of the dislocations. When the depinning begins the recovery process and the softening starts again.

Softening curves can also contain a minimum (a 'hump'), as shown in figure 2. This minimum is caused by precipitation hardening. This hardening mechanism is determined quantitatively by the precipitate volume fraction and the precipitate size distribution. Both quantities change with time, such that the precipitation hardening first increases and subsequently decreases. If the increase of the hardening is larger than the increase of the softening due to recovery, a minimum in the softening curve can occur.

The softening curve can even have *two* plateaus. Zurob et al. explain one plateau from the hindrance of recovery due to the pinning of dislocations and the other plateau by a balance of recovery (softening) and precipitation hardening. The changes in the contributions of these two mechanisms can be (almost) equal for a short period of time, resulting in the plateau. However, as recovery proceeds and precipitation hardening decreases, recovery becomes the dominant mechanism and softening increases again.

In addition to recovery, recrystallization can contribute to the softening. However, the dynamics of recrystallization in HSLA steel can be different from steel grades without niobium. With sufficient deformation recrystallization always occurs in plain carbon steels. In HSLA steels, however, the recrystallization is retarded and, depending on the chemical composition and process parameters recrystallization may only be partial. In the remaining, unrecrystallized part of the material softening only occurs by recovery.

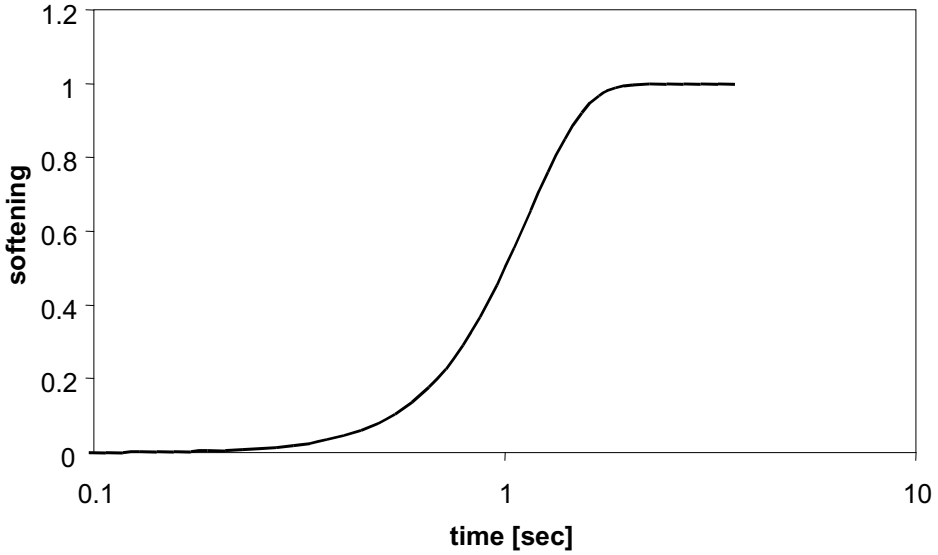

Figure 4a. Softening curve of sigmoidal form.

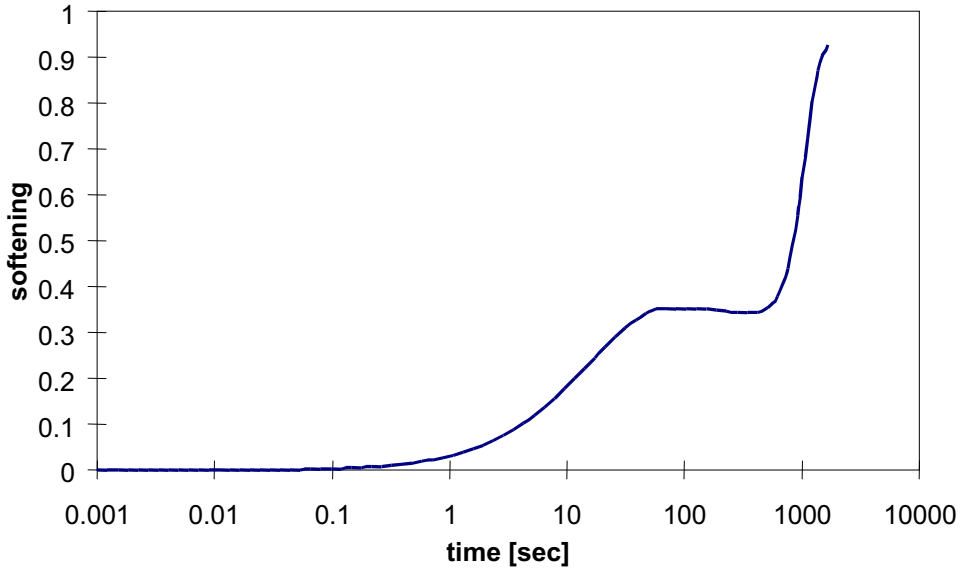

Figure 4b. Softening curve with one plateau.

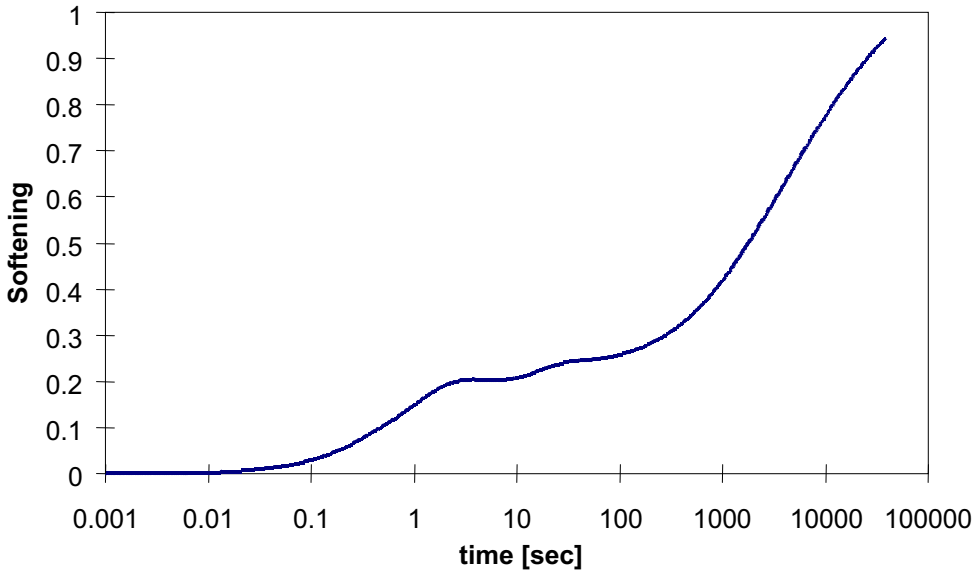

Figure 4c. Softening curve with two plateaus.

This varying behavior can best be explained by considering the total driving force for recrystallization: the total driving force is the difference of the stored energy, introduced by the deformation, and the Zener pinning force, caused by precipitates. While the stored energy decreases monotonously due to recovery, the Zener pinning force initially increases when precipitates nucleate and start growing and subsequently the Zener pinning force slowly decreases as the precipitates coarsen. If the stored deformation energy equals the Zener pinning force the total driving force for recrystallization is zero and recrystallization stops. The instant at which the reduction of the total driving force occurs determines the final recrystallized volume fraction. Depending on the chemical composition and process parameters the driving force may already have fallen to zero before the recrystallization has started.

The partial recrystallization or its complete absence is one explanation for the temperature of no recrystallization T_{NR}. However, there is no common definition of T_{NR} and in practice a pragmatic approach will be used: as shown in Figure 2 the recrystallization can be retarded considerably. The time at which 50 percent of the volume is recrystallized can range from a few seconds to thousands of seconds, depending on the temperature. Interstand times in a hot rolling mill are of the order of seconds or tens of seconds, much smaller than the time scale of recrystallization at those temperatures. Then, T_{NR} is an indication of the temperature at which the time scale of recrystallization becomes comparable to the interstand times in the hot strip mill.

Application of the Model

After calibration of the model using literature data, it can be applied in several way:

- In-process analysis, to understand and solve difficulties in production, focused on short time troubleshooting.
- In the design of HSLA steels, by optimization of the chemical composition.
- In the optimization of the rolling scheme.
- in the prediction of precipitation hardening in ferrite, i.e. the contribution of the precipitates, formed in austenite.

Future Developments

The model of Zurob in its present form should be considered as a basic model, forming a fundament for more complex models and for models for other phases than austenite. A few examples of extensions of the model are:

- Introduction of more complex Ti, V, Nb carbonitride precipitates.
- Introduction of elements, such as manganese and silicon, which do not form precipitates, but influence the solubility products of the precipitates and the solute drag effect.
- A precipitation-recovery-recrystallization model for the ferrite phase. Such a model would enable the simulation of the interaction of recrystallization and the precipitation of aluminum nitride during annealing.
- Simulation of multi deformation experiments or rolling mill, instead of only one deformation in the present model.
- The simulation of T_{NR}.

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