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Part 2 of 2: Constitutive Behavior Modeling of Steels Under Hot-Rolling Conditions

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Part 2 of 2: Constitutive Behavior Modeling of Steels Under Hot Rolling Conditions

Authors:

Y. W. Cheng and P. T. Purtscher

Performing Organization Names and Address

National Institute of Standards and Technology (NIST)
325 Broadway
Boulder, Colorado 80303-3328

Abstract

This report describes the work done at NIST/Boulder for the Microstructural Engineering in Hot-Strip Mills project and is separated into two sections: One section begins with page 1, and the other begins with page 61. The first section deals with the development of constitutive models to predict stress-strain behavior of steels under hot-rolling conditions and the second section deals with the development of models for predicting mechanical properties of hot-rolled steels as functions of chemical composition, microstructural features, and processing variables.

In the first section, we have developed four models that are suitable for predicting the stress-strain behaviors of the following eight steels: A36, DQSK, HSLA-V, HSLA-Nb, HSLA-50/Ti-Nb, HSLA-80/Ti-Nb, and two interstitial-free (IF) grades. The development of these models has been based on experiments conducted at NIST/Boulder. To support the modeling effort, one servohydraulic machine was designed and built to generate data at high strain rates (up to 50 s^{-1}), which was not previously possible at NIST. The model predictions compared favorably with data from CANMET, which have strain rates up to 150 s^{-1} . The model for HSLA-80/Ti-Nb was used in the Sims equation to calculate the rolling forces for rolling an HSLA-80 slab. The calculated results agree well with the mill measurements. All the models are currently used in the Hot-Strip Mill Model (HSMM) developed at the University of British Columbia-Canada.

In the second section, equations have been developed and validated for plain C and HSLA steels produced on the hot-strip mill. The lower yield strength, ultimate strength, and % total elongation are the main outputs from the equations. For thicker, plate-type products, the ductile-to-brittle transition temperature can also be predicted. The HSMM allows for considerable flexibility and detail in the setup for the hot-strip mill and still provides accurate predictions for yield and ultimate strength, about $\pm 35 \text{ MPa}$. The two sets of equations developed for the structure-property relationships are applicable to a wider range in chemistry than just the eight grades in the program. Base strength equations are considered applicable to grades where the microstructure is predominantly polygonal ferrite; prediction of precipitation strengthening is applicable to steels with Nb contents from 0.02 to 0.09%, 0.08 % V, and excess Ti levels up to 0.025%.

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Foreword

This project is a cooperative effort involving the National Institute of Standards and Technology (NIST), the University of British Columbia (UBC-Canada), and U.S. Steel. Only two tasks of the project were performed at NIST/Boulder, and the remaining of the project were done at UBC-Canada. At the start of the project, U.S. Steel provided all the samples for experiments. Later, Dofasco, LTV Steel, Rouge Steel, and Stelco also contributed samples of some steel grades. In addition to these steel companies, the authors also wish to acknowledge the following companies for their support and active participation during the conduct of the project: Bethlehem Steel, Inland Steel, IPSCO Steel, Lukens Steel, National Steel, and The Timken Company.

The project would not have been possible without the support and guidance of the two Directors of the Advanced Process Control Program: Lawrence Kavanagh and Joseph Vehec. This work was performed under the supervision of G. Alers and H. I. McHenry, Materials Reliability Division, National Institute of Standards and Technology, Boulder, Colorado.

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Executive Summary

This report describes the work done at NIST/Boulder for the Microstructural Engineering in Hot-Strip Mills project, and is separated into two sections: The first section begins with page 1, and the second section begins with page 61. The first section deals with the development of constitutive models to predict stress-strain behavior of steels under hot-rolling conditions and the second section deals with the development of models for predicting mechanical properties of hot-rolled steels as functions of chemical composition, microstructural features, and processing variables.

In the first section, we have developed four models that are suitable for predicting the stress-strain behaviors of the following eight steels: A36, DQSK, HSLA-V, HSLA-Nb, HSLA-50/Ti-Nb, HSLA-80/Ti-Nb, and two interstitial-free (IF) grades. The development of these models has been based on experiments conducted at NIST/Boulder. To support the modeling effort, one servohydraulic machine was designed and built to generate data at high strain rates (up to 50 s^{-1}), which was not previously possible at NIST. The model predictions compared favorably with data from CANMET, which have strain rates up to 150 s^{-1} . The model for HSLA-80/Ti-Nb was used in the Sims equation to calculate the rolling forces for rolling an HSLA-80 slab. The calculated results agree well with the mill measurements. All the models are currently used in the Hot-Strip Mill Model (HSMM) developed at the University of British Columbia-Canada.

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CONSTITUTIVE BEHAVIOR MODELING OF STEELS UNDER HOT-ROLLING CONDITIONS

Y. W. Cheng, R. L. Tobler, and B. J. Filla
Materials Reliability Division
National Institute of Standards and Technology
Boulder, CO 80303

K. J. Coakley
Statistical Engineering Division
National Institute of Standards and Technology
Boulder, CO 80303

Abstract

The purpose of this work is to develop constitutive equations that predict the stress-strain curves of steels as functions of processing variables, such as temperature, strain rate, and microstructural features under hot-rolling conditions. Eight steels, A36, DQSK, HSLA-Nb, HSLA-50/Ti-Nb, HSLA-80/Ti-Nb and two interstitial-free (IF) grades, were selectively tested at temperatures in the range from 900 to 1200 °C, strain rates from 1 to 50 s⁻¹, and austenite grain sizes from 0.012 to 0.29 mm. The data were analyzed to support the development of constitutive models.

To cover a wider range of strain rates, a computer-controlled servo-hydraulic laboratory apparatus was designed and built for high strain-rate experiments. The machine has a load capacity of 100 kN and controllable constant strain rate of 50 s⁻¹. It has been used to generate all the data with the strain rate of 50 s⁻¹ in this project.

Models for predicting stress-strain behaviors for the eight steel grades listed above have been developed. Model predictions correlate well with the experimental data, indicating that the models adequately describe the stress-strain behaviors of various steel grades under hot-rolling conditions. Model predictions for A36 and DQSK are further validated with high strain-rate (up to 150 s⁻¹) data obtained at CANMET.

The models have been used in Sims equations to calculate the roll separating forces which agree well with mill measurements. The models have been incorporated in the Hot-Strip Mill Model (HSMM) developed at the University of British Columbia-Canada.

Keywords: Constitutive behavior; dynamic recrystallization; elevated-temperature mechanical properties of steels; hot-strip rolling; IF steels; microalloyed steels; plain-carbon steels; thermomechanical process simulation.

1. Introduction

In 1993, the American Iron and Steel Institute (AISI) sponsored a research project on microstructural engineering in hot-strip mills. The object was to develop a predictive tool that would quantitatively link the properties of hot-rolled products to mill processing parameters. The predictive tool would be a user-friendly computer model that incorporates knowledge of heat flow, microstructural phenomena (recrystallization, grain growth, precipitation, and austenite decomposition), and structure-composition-property relationships. It would compute the thermal and microstructural evolution of steel during hot rolling, as well as the final product properties, as a function of hot-strip mill design and operating practices.

This AISI project is a cooperative effort involving the National Institute of Standards and Technology (NIST), the University of British Columbia (UBC-Canada), and U.S. Steel. Major portions of the research have been performed at NIST and UBC. The University of British Columbia conducted heat-transfer studies, quantitative characterization of kinetics of microstructural evolution, and model verification with plant and pilot-plant measurements. NIST focused on constitutive behavior and structure-composition-property relationships. The goal of the constitutive-behavior studies was to model the flow behavior of steels as functions of chemistry, temperature, strain rate, and other processing variables. This portion of the report presents the results of the constitutive-behavior studies, including experiments to provide stress-strain data and to understand the effects of different parameters, and models to predict stress-strain curves under hot rolling conditions.

2. Background

Characterizing the plastic flow of steels is essential to improve industrial hot-forming operations. During forging or hot rolling, the plastic-flow properties determine the power requirements of industrial equipment, as well as the mechanical properties and dimensional accuracy of finished products.

Deformation at high temperatures and high strain rates involves a sizable number of variables. Material response is affected by changes in temperature and strain rate, as well as steel chemistry and microstructural factors including grain size, grain growth, recrystallization, and precipitation during processing. The demand for accurate flow-property predictions under hot-forming conditions has increased in recent years due to research pursuits in the field of deformation simulation and to commercial needs to manufacture products with ever increasing efficiency.

Figure 1 presents the block diagrams (rectangular blocks) of a typical hot-strip rolling operation and the various computer models for different stages (oval-shape blocks) of the operation. Steel experiences the following conditions during its conversion from slab to hot strip:

- Temperature may reach 1250°C at the first stand of the roughing mill, and 900°C or lower at the last stand of the finishing mill.

- Strain rate may reach 1 s^{-1} at the first stand of the roughing mill, and $100\text{-}200 \text{ s}^{-1}$ or higher at the last stand of the finishing mill.
- Grain size may exceed 0.250 mm at the first stand of the roughing mill, but it may be as small as 0.015 mm at the last stand of the finishing mill.

Obviously, the material's flow stress varies considerably during different stages of processing. The research objective was to develop and validate constitutive equations for use in strip-rolling simulations using numerical techniques, such as the finite element method. Constitutive equations describe the nonlinear relationships that exist among the process variables and the internal structures of materials, and they permit us to predict or estimate the steel flow properties of interest for specified strip-rolling conditions.

Figure 2 shows some experimentally determined flow-stress curves for DQSK steel. The true-stress-versus-true-strain relationship is called the flow-stress curve because it plots the stress required for plastic flow at a given strain. The shape of the flow-stress curve relates to different structural changes going on in the steel as it undergoes hot deformation in the austenite phase. The three shapes of these curves indicate three possible responses to hot deformation at high temperatures and high strain rates. The Type I flow-stress curve (at $900 \text{ }^{\circ}\text{C}$) rises continuously, owing to strain hardening in the austenite. The Type II curve (at $1000 \text{ }^{\circ}\text{C}$) rises at first, and then reaches a plateau. The plateau is caused by the occurrence of dynamic recovery in amounts sufficient to counterbalance the effect of strain hardening in austenite. Finally, the Type III curve (at $1100 \text{ }^{\circ}\text{C}$) rises to a peak and then subsequently decreases. The decline of the flow stress in a Type III curve is caused by dynamic recrystallization, which begins when the critical strain, ϵ_c , is reached.

An important factor that determines which type of the flow-stress curve will manifest is the Zener-Hollomon parameter [1]:

$$Z = \dot{\epsilon} \exp \frac{Q}{RT} \quad (1)$$

Here, $\dot{\epsilon}$ is strain rate, T is temperature, R is the universal gas constant, and Q is the apparent activation energy of the rate-controlling dynamic softening process. High Z promotes strain hardening, whereas low Z favors dynamic recrystallization, which fosters the decline of flow stresses following a peak stress. Deformation at intermediate Z corresponds to a balance between strain hardening and dynamic recovery, which is characterized by a constant flow stress after initial hardening. If strain rate is held constant in eq (1), then Z increases with decreasing temperature and the flow stress at low temperature increases. Conversely, the flow stress will decrease at low strain rates.

The flow-stress curve is fundamental to materials engineering. From it, the important properties of a steel are interpolated, and the mechanisms of deformation can be determined. Many attempts have been made to fit mathematical expressions to empirical flow curves at selected temperatures and strain rates. Exponential equations or power laws and their derivatives

are commonly used. The parametric constants that relate true stress, σ , and true strain, ϵ , in these equations must be determined experimentally.

Table 1. Stress-strain relationships cited in the literature.

Equations		References
$\sigma = K \cdot \epsilon^n$	(2)	2
$\sigma = \sigma_0 + K \cdot \epsilon^n$	(3)	3
$\sigma = b + c \cdot \epsilon^{0.4} + d \cdot \epsilon^{0.3} + e \cdot \epsilon^{1.2}$	(4)	4
$\sigma = K(\epsilon_0 + \epsilon)^n$	(5)	5
$\sigma = K \cdot \epsilon^{n_1} + \exp(K' + n_2 \cdot \epsilon)$	(6)	6
$\sigma = \sigma_0 + (B + \epsilon)^n$	(7)	7
$\sigma = \sigma_0 + B \cdot \log(\epsilon)$	(8)	7
$\sigma = \sigma_s [1 - \exp(-B \cdot \epsilon)]^n$	(9)	8
$\sigma = \sigma_0 + B [1 - \exp(-C \cdot \epsilon)]^n$	(10)	7,9
$\sigma = \sigma_0 - (A - B) \cdot \exp(-C \cdot \epsilon)$	(11)	10
$\sigma = B [1 - \exp(-C \cdot \epsilon^{1/n})]^{(1/n)}$	(12)	7
$\sigma = 1.64 \cdot \sigma_p \cdot \epsilon^n$	(13)	11
$\sigma_p = \text{peak stress}; n = 0.22 + 0.001 \cdot \sigma_p$		

Table 1 lists the various equations that are cited in the literature [2-11] and have been proposed for use in modeling. These equations were applied to describe Type I or Type II behaviors. Most equations in Table 1 were developed to fit stress-strain relationships at relatively low temperatures; they may fit the experimental data well, but only for certain materials, and for limited experimental ranges. For steels at elevated temperatures, eq (4), which is a modified form of Ludwik's power-law equation of eq (3), has been shown to fit some experimental data rather well [4]. At NIST, we successfully used eq (9) to describe the stress-strain behavior of bar steels [8]. The parameters B and n are nearly constant for medium-carbon microalloyed steels when σ_s is taken as the flow stress in the steady state. Regarding Type III flow curves, no single governing equation has been found. Different approaches [9,12-14] have been used to account for the amount and kinetics of softening due to dynamic recrystallization.

3. Constitutive Equations and Factors Affecting Flow Stress

Constitutive equations describe nonlinear relationships that exist among the process variables such as effective stress, effective strain rate, and temperature at different levels of strain. Such equations are unique for each material and processing condition. Usually, for practical applications they are developed using data obtained under simplified experimental conditions.

The constitutive equations are usually described using a rate expression:

$$\dot{\epsilon} = f(\sigma, T, S), \quad (14)$$

where $\dot{\epsilon}$ is strain rate, σ is stress, T is temperature, and S is the structural state of the material. In general, S is affected by deformation (strain and strain rate), and temperature.

One of the goals in constitutive-behavior modeling is to predict the flow stress of a given steel under various processing conditions. For this purpose, rate expressions, such as eq (14), are rearranged to express the flow stress explicitly as a function of strain, temperature, and strain rate.

3.1 Temperature and Strain Rate

The rate at which strain is applied can have an important influence on the flow stress. In general, increasing strain rate increases flow stress. Also, the shape of the stress-strain curve is strongly dependent on temperature. Decreasing temperature, in general, increases flow stress. Typically, the temperature and strain-rate dependence of the rate expression is described by the following hyperbolic sine equation [15],

$$\dot{\epsilon} = A[\sinh(\alpha\sigma)]^m \exp\left[\frac{-Q}{RT}\right], \quad (15)$$

where A , α , and m are experimentally determined constants. At low stresses ($\alpha\sigma < 0.8$), the equation is reduced to a power relation such as the equation used to describe creep behavior,

$$\dot{\epsilon} = A_1\sigma^m \exp\left[\frac{-Q}{RT}\right], \quad (16)$$

and at high stresses ($\alpha\sigma > 1.2$), it reduces to an exponential relation,

$$\dot{\epsilon} = A_2 \exp(\beta m_2 \sigma) \exp\left[\frac{-Q}{RT}\right]. \quad (17)$$

Equation (15) was derived based on results of steady-state creep experiments. It has been observed that the constants in eq (15) decrease with strain, especially at small strains, where

steady state has not been reached. After the steady-state stress is reached, these constants approach a constant with little variation. It has been shown that eq (15) describes the hot-deformation behavior of steels better than eqs (16) and (17)[16]. Rearranging and expressing the steady-state stress as a function of Z , eq (15) becomes

$$\sigma = \frac{\sinh^{-1}\left(\frac{Z}{A}\right)^{\frac{1}{m}}}{\alpha} \quad (18)$$

Low Z promotes dynamic recrystallization (Type III behavior in fig. 2). As Z increases, the peak in the curve broadens and the strain that corresponds to the peak stress also increases. With further increase of Z , Type III behavior approaches Type II behavior and the peak disappears.

Many hot-deformation studies are devoted to determining the constants A , α , m , and Q in eq (15). Once these constants are determined for a given steel, eq (15) can be used to interpolate or extrapolate to conditions for which data are unavailable.

3.1.1 Apparent Activation Energy, Q

For C-Mn steels the values of Q fall in the range 262-366 kJ/mol [1,4, 17-20]. The values of Q tend to increase with additions of certain alloying elements including Nb. Values reported for some Nb microalloyed steels fall in the range 325-461 kJ/mol [12,20,21]. The apparent activation energy is frequently, related to the activation energy for self-diffusion of iron in austenite, and the reported values are similar. However, some materials show distinct differences between activation energies determined under hot-working conditions and those measured during creep [22]. Such materials include steels [22]. In such cases, one should beware not to extrapolate to conditions beyond the range of actual experimental observations since Q might vary with strain rate. Abnormal behavior of Q has been reported for Nb steels [4].

3.1.2 Constants A and α

The unit for the constant A is s^{-1} , and values are usually expressed in $\ln A$. Some reported values for $\ln A$ are in the range 25-32 [4,12]. The values of α are usually fixed at 0.012 [12] or 0.013 [4], and the unit for this constant is MPa^{-1} .

3.1.3 Value for m

The values of m reported in the literature range from 3.2 to 5.0 [4,12,20,22]. This dimensionless parameter relates to the strain-rate sensitivity of the material which is a measure of the dependence of flow stress on strain rate. The higher the strain rate, the higher the flow stress. It is widely reported that the strain-rate sensitivity increases with increasing temperature [23,24]. However, for hot-worked steels, the value of m remains constant after the steady-state condition is reached [16]. Note that m , as expressed in eq (15), decreases with an increase of the strain-rate sensitivity.

3.2 Dynamic Recrystallization

The occurrence of dynamic recrystallization may profoundly affect the flow stress, as shown in fig. 2. Although dynamic recrystallization and its effects have been widely studied, it has been argued that prospects for the occurrence of dynamic recrystallization during rolling in a single pass are remote since the critical strain, ϵ_c , for dynamic recrystallization is seldom achieved [20,25]. However, the attainment of this critical strain is more probable during multi-pass rolling in the finishing stands. The interpass time in that case is short, and the rolling temperature is low; therefore grains deformed in a previous pass do not have sufficient time to statically recrystallize after deformation, and the strain is retained to the next pass. Thus, strain increments accumulated in the process of several passes can reach ϵ_c and induce dynamic recrystallization. Although the occurrence of dynamic recrystallization in an industrial rolling process seems likely, there has been no report in the literature to substantiate the hypothesis.

A model that considers the occurrence of dynamic recrystallization on stress-strain behavior has been reported [13,14,24]. This model derives from the fundamental theories of hardening and softening (recovery and recrystallization), and it applies to a wide range of materials and test conditions. As a consequence of its broad applicability, this model contains a large number of constants that must be determined experimentally. This necessitates some creep test experiments.

A simpler model that involves four constants has also been proposed for hot-worked steels [12]. Two of these four constants are material properties: the steady state flow stress after dynamic recrystallization and ϵ_c , which is a function of the initial grain size and Z . The steady state flow stress after dynamic recrystallization, σ_{recd} , for a Nb-B steel is expressed as

$$\sigma_{recd} = 8434 \sinh^{-1} \left[\left(\frac{Z}{5.2 \times 10^{11}} \right)^{0.2} \right] \quad (19)$$

3.3 Chemical Composition

In general, the effects of chemical composition on flow properties are embedded in the constants of the stress-strain equations. For example, additions of Nb increase Q in eq (15), as noted previously. Few attempts have been made to express the flow stress in terms of composition [26]. But systematic research quantifying the effects of individual alloying elements is scarce.

3.4 Austenite Grain Size

Grain size is an important parameter governing the low temperature mechanical properties of steels, such as strength and toughness. However, the existing literature for steels contains little information describing the effects of grain size on flow stress under hot-working conditions, aside from its effects on the occurrence of dynamic recrystallization.

It is stated that equilibrium subgrain structures are established early during deformation at high temperatures [22,27]. These subgrain structures determine the flow stress under hot-

It is stated that equilibrium subgrain structures are established early during deformation at high temperatures [22,27]. These subgrain structures determine the flow stress under hot-working conditions, except at the very early stage of deformation before the subgrain structures are established. Thus, the initial austenite grain size plays little role in determining the flow stress of steels. Nevertheless, the austenite grain size does change from one pass to another during rolling. And the initial grain size influences the value of ε_c , which increases with an increase of grain size [1]. Therefore, more research is needed to arrive at conclusive results about the effects of grain size on the flow behavior of steels.

3.5 Precipitation

Precipitation is another factor that has potent effects on low-temperature mechanical properties but is not widely discussed in the field of high-temperature flow properties. During hot rolling, precipitation in C-Mn steels is insignificant, but it does occur in microalloyed steels. Dynamic precipitation is not likely to take place in single-pass or even multipass rolling because insufficient time is available [12,25]. However, the conditions for strain-induced precipitation seem to be sufficient in the finishing stands of a multipass rolling operation, especially in the case of Nb-alloyed steels. Temperature is relatively low during rolling in the finishing stands, and strain is accumulated over several passes. However, even if precipitates do form during rolling, the potency of their effects remains at issue.

3.6 Adiabatic Heating

Adiabatic heating is a factor in tests at high strain rates. Typically, at strain rates exceeding 1 s^{-1} , the temperature of a test specimen will increase continuously while the test is progressing. The heat generated in a specimen during deformation at such rates cannot be dissipated quickly enough to maintain an isothermal condition. If we assume that all the mechanical energy is converted into heating the specimen, then the temperature increase ΔT due to deformation can be estimated [4]:

$$\Delta T = \frac{\sigma_m \delta \varepsilon}{\rho C_p} \quad (20)$$

$(\sigma_m \delta \varepsilon)$ is the area under the stress-strain curve without temperature correction, ρ is the steel's density, and C_p is the heat capacity of the steel. The incremental stress decrease due to the adiabatic heating is

$$\Delta \sigma = \left(\frac{Q}{m \alpha R} \right) \left(\frac{1}{T} - \frac{1}{T + \Delta T} \right) \quad (21)$$

Depending on test conditions, the specimen's temperature rise can be as high as 30°C at a strain of 0.7 [4]. A 10°C increase in temperature will generally cause a 2-5 MPa decrease in flow stress. According to one report, the measured and calculated temperature increases due to adiabatic deformation agree [4].

4. Experimental Approach

The goal of constitutive modeling is to predict material flow properties. Knowledge of the stress-strain curve is needed for efficient processing, along with the yield and flow-strength data. Experiments provide an accurate mechanical property database for each steel composition. Models must then be developed, based on equations that relate the deformation parameters.

In this study, we used uniaxial compression tests and right-circular cylinder specimens to characterize the mechanical behavior of selected steels. There are no standard test procedures for determining flow stress curves at the relatively high temperature and strain rate conditions encountered in industrial processing and metalworking. Compression testing offers advantages compared to other test methodologies, especially in simplicity and economy of experimental apparatus.

Tension and torsion tests are viable alternatives. Tension test specimens are not desirable for hot-deformation studies because they typically neck at strains in the range 10 to 30%. Much larger strains without plastic instability can be achieved using torsion or compression tests. However, the strain and strain rate in torsion tests will vary from a maximum at the surface of the cylindrical specimen to zero at the center. The measured flow stress is then an average stress over a certain amount of material at the outer layer of the specimen. The strain and strain rate in torsion tests are likewise averages. It is assumed that the outer layer of the torsion specimen provides the major contribution to the flow stress because of the stress distribution and work hardening [28]. Torsion test specimens are usually rigidly fixed in the drive system. Therefore, the strain-rate transient depends on the time needed to attain the operating speed. The strain rate in this transient period is usually limited to a rate less than 10 s^{-1} [29].

The right-circular cylinder is often used in tests of metals to simulate thermomechanical processing effects. Cylindrical specimens are small and readily machinable compared to specimens used in other test methodologies. The cylindrical specimen is also recognized in the ASTM Standard Practice for Compression Tests of Metallic Materials at Elevated Temperatures with Conventional or Rapid Heating Rates and Strain Rates (ASTM E 209 - 65).

A disadvantage of the cylindrical specimen is that it barrels during compression testing. Barreling in axisymmetrical compression testing is caused by: (1) the friction that exists at the platen-specimen interface, and (2) the temperature gradient that exists between the platen-specimen interface and the middle of the specimen.

Barreling in compression tests causes nonuniform strain and strain rates. Nevertheless, reports show that barreling does not affect the measured flow stress if the specimen aspect ratio (the specimen diameter divided by its height) is kept between 0.5 and 0.8 [30,31]. Higher loads are needed to deform a specimen when friction is present. However, barreling is usually accompanied by shear instability, which reduces the load. In certain specimen geometries the load increase due to friction is possibly compensated by the load reduction due to shear instability. For this reason, we used no lubricants for the experiments described in this study. Lubricants at the platen-specimen interface may reduce barreling, but they introduce other complications.

during deformation is a major factor influencing the recrystallized austenite grain size after deformation. The nominal strain that is measured during deformation differs significantly from the actual strain that the specimen receives [33]. Therefore, barreling may be ignored if the testing is solely for the measurement of flow stress, but it must be taken into account if microstructural evaluation is involved.

4.1 Materials

Two plain carbon steels, two interstitial free (IF) steels, and four microalloyed steels were tested in this study. Table 2 identifies the eight steels and their chemical compositions. Mechanical properties of the steels at room temperature are reported in the Subtask of Structure-Properties Relations.

Table 2. Chemical compositions (mass %).

Elem	HSLA-V	HSLA-Nb	HSLA-50 Ti-Nb	HSLA80 Ti-Nb	A36	DQSK	IF Nb rich	IF Nb lean
C	0.045	0.082	0.071	0.07	0.17	0.038	0.0028	0.002
Mn	0.45	0.48	0.758	1.35	0.74	0.30	0.17	0.106
S	0.005	0.005	0.013	0.003	0.008	0.008	0.006	0.008
Si				0.13				
V	0.080							
Ti			0.013	0.042			0.035	0.059
Nb		0.036	0.023	0.085			0.035	0.009
Al	0.078	0.024	0.053	0.039	0.040	0.040	0.027	0.033
N	0.0072	0.0054	0.0067	0.0075	0.0047	0.0052	0.0029	0.0041

4.2 Test Specimen

The cylindrical test specimen was 10 mm in diameter and 15 mm high. The cylindrical specimens were machined from crop samples such that the cylindrical axis coincided with the through-thickness direction. The steel types were identified in Table 2.

4.3 Apparatus

Two servohydraulic test machines are available in our laboratory: one has a load capacity of 250 kN and the other 100 kN. In addition to this difference in load capacity, the two machines differ in attainable strain rates, and in heating and cooling techniques. The 250-kN machine is an existing machine and the 100-kN machine was developed and commissioned

during the investigation of this project. Using these machines, we compressed specimens between quartz or ceramic platens located inside an evacuated test chamber at elevated temperatures.

Fused quartz or ceramic rods are used for platens. The ceramic platens are transformation-toughened zirconia (TTZ) in the form of solid white rods. The TTZ rods are more expensive than quartz rods, but they are preferable in our experience since quartz rods are more susceptible to crack after repeated use.

The apparatuses are described elsewhere [34,35], and only a brief description of the apparatuses are given here. The subcomponents of the apparatus include a computer-controlled servohydraulic load frame, a test chamber providing vacuum environment, a furnace equipped with a programmable temperature controller, a microcomputer, and a storage oscilloscope. The axisymmetric compression test methodology offers benefits since the required specimen size is relatively small. This facilitates system design by minimizing the volume and mass of test fixtures.

4.3.1 Computer and peripherals for machine control and data acquisition.

Figure 3 is a flowchart of the plan for machine control and data acquisition. The computer is a 25-MHz 80486 personal computer (PC). Peripherals include a line printer, a 340-MB hard disk, a 1.2-MB floppy disk drive, and a 360-KB floppy disk drive.

The apparatus is equipped with a programmable digital storage oscilloscope for high-speed data acquisition. The oscilloscope includes four 15-bit, 100-kHz digitizers. Each channel stores up to 4000 data points when all four channels are operating simultaneously. The same computer and the storage oscilloscope are used interchangeably with the two test machines.

The computer components and accessories described here are existing pieces of equipment that serve multiple uses in our laboratory. They do not represent up-to-date technology.

4.3.2 Load Frames

Each machine is equipped with a closed-loop servohydraulic load frame. One machine has a load capacity of 250 kN and the other 100 kN. These machines are not conventional in terms of actuator travel speed. The machines have been modified by using high flow-rate servovalves and large accumulators. The modification gives attainable actuator speeds up to 500 mm s⁻¹. With the currently used specimen geometry, this travel speed produces controllable strain rates up to 25 s⁻¹ for the 250-kN machine, and 50 s⁻¹ for the 100-kN machine.

The most desirable deformation mode for rolling simulation is a constant true-strain-rate compression, which greatly eases data analysis and interpretation. Constant true-strain-rate compression is accomplished through displacement control of the load train by a PC using the following equation,

$$h(t) = h_i \exp(-t\dot{\epsilon}). \quad (22)$$

Here, $h(t)$ is the specimen's instantaneous height at time t , h_i is the initial height, and $\dot{\epsilon}$ is the desired constant true strain rate.

In a high strain-rate experiment, the actuator's travel starts from a rest position and then accelerates to the desired speed that results in slow rates at the very beginning of the experiment. To overcome the problem, we retract the actuator at the beginning of a test so that the specimen is compressed using only that portion of the actuator travel that produces constant rate. Then, the value of $\dot{\epsilon}$ is calculated using the actual signals recorded by the data acquisition system [36].

4.3.3 Heating, Cooling, and Vacuum

For the 100-kN machine, a vertically mounted electrical tube furnace (power rating is 6.0 kW at 240 V and 21.6 A) is used for heating. The heating elements are four tubular tungsten-filament quartz lamps, each backed by an elliptical reflecting wall. The vacuum chamber is a quartz tube that encapsulates the ceramic rods and the specimen. The tube is 50.8 mm in diameter and 264 mm long. The elliptical reflectors, which surround the quartz tube, focus radiant energy from the tungsten lamps to the outer surfaces of the specimen in the quartz tube at the center of the heating chamber. The physical sizes of this furnace and its power supply are small, which contributes substantially to the compact design and efficiency of the entire test facility.

Heating of a specimen is achieved through radiation and the heating rate is limited to about $2^\circ\text{C}\cdot\text{s}^{-1}$. There is no forced cooling device for fast cooling of a heated specimen. The cooling is achieved through radiation to the surrounding and through conduction to the loading rods. Thus, cooling of a heated specimen after a test is slow, and it takes 2-3 h to furnace cool the specimen. The vacuum system consists simply of a pressure gage and a mechanical pump by which the chamber is evacuated during a test to minimize oxidation of the specimen at high temperatures.

The 250 kN apparatus uses an induction furnace and a controlled gas-flow cooling process. A 10-kW, 450-kHz induction generator (rf frequency) generates the heat demanded by the temperature controller. Heating coil is a 4.76-mm-diameter Cu tube. A stainless-steel vacuum chamber encloses the specimen, and the heating and cooling coils. Vacuum is obtained with a 100-mm diffusion pump, which can achieve a vacuum of 1.33×10^{-3} Pa (1×10^{-5} Torr) within 20 min. It should be mentioned here that at a pressure of about 1.33×10^{-1} Pa (10^{-3} Torr), gasses are ionized between the specimen and the heating coils. At this pressure, it is impossible to heat the specimen because the rf field arcs and shuts off the rf generator.

Cooling of a specimen for the 250-kN machine is controlled with a PC by regulating gas flow onto the specimen through a 6.35-mm-diameter Cu coil concentric with the heating coil. The side of the cooling coil facing the specimen is perforated to provide a gas flow that cools the specimen uniformly. Either helium or nitrogen gas is used for cooling. Because of its higher specific heat (5230 versus 1030 $\text{J}\cdot\text{kg}^{-1}\cdot\text{K}^{-1}$ at 25°C) and greater mobility, helium is more efficient

than nitrogen in rapid cooling. For a cylindrical steel specimen 10 mm in diameter by 18 mm in height, the cooling rate is controllable to a maximum rate of $25^{\circ}\text{C}\cdot\text{s}^{-1}$.

Both furnaces are operated automatically by means of a programmable temperature controller that has ramp-and-soak and multiple set-point capabilities. Before testing, the desired thermal profile is selected and preprogrammed into the controller. The temperature at any phase of the experiment is then monitored by means of an S-type thermocouple (a Pt and Pt-10%Rh pair) that is spot-welded to the surface of the cylindrical specimen at the midpoint of height.

4.3.4 Machine Performance and Reproducibility

4.3.4.1 Alignment

Both test machines are rigidly constructed. The specimen dimensions were measured before and after compression using a micrometer to confirm the final value of strain, and to confirm that good axial symmetry is retained after testing. The fact that the top and bottom surfaces of specimens remain parallel after compression testing is a qualitative indication of good mechanical alignment.

4.3.4.2 Reproducibility

Figure 4 shows the results of triplicate tests of DQSK specimens using the 250-kN machine. The results demonstrate excellent reproducibility of the machine. Figure 5 shows a pair of stress-strain curves for DQSK specimens that were tested using the 100 and 250-kN machines at nominally identical test conditions. In this example, the steel grain size is 0.114 mm, the test temperature is 1000°C , and the strain rate is 25 s^{-1} (the maximum possible strain rate of the 250-kN apparatus). The near equivalence of the stress-strain curves presented in this figure demonstrates acceptable reproducibility.

4.3.4.3 Strain Rates

At low strain rates, the actuator of the machine follows the command signals quite well. As a result, the nominal constant strain rate can be achieved and maintained. Figure 6 shows the results of tests at nominal constant strain rates of 1 and 10 s^{-1} . Here, the strain rates were calculated using the actuator displacement-versus-time signals.

If the command signals call for instant high speeds, the actuator is not capable of following the demands instantaneously due to inherent characteristics of servohydraulic test machines. The actuator's speed will be slower than the command signals at the beginning and at the end of a command cycle. After receiving the commands at the onset of a test, the actuator needs time to accelerate from a rest position to the desired speed. Similarly, it needs time to decelerate from a high-speed travel to rest as called for by the command signals at the end of the test. This behavior is clearly demonstrated in figure 7.

As mentioned in Section 4.3.2, the problem of slower speed at the beginning of a test has been remedied by actuator retraction. However, the problem at the end of a test has not been

corrected in our laboratory. The problem limits the valid strain range during a high strain-rate test. Thus, the actual strain rate during a high-speed test varies instantaneously, so specified values are necessarily quoted in nominal terms.

4.3.4.4 Specimen Strength

The machine performance is slightly affected by secondary factors such as the specimen's size and flow strength. The data in figure 8 show implicitly that the machine performance is influenced by the properties of the test material. Here, machine performance is gaged in terms of the ratio H_{fc}/H_{fm} . The measured and commanded final heights are H_{fm} and H_{fc} , respectively. A ratio of 1 means the commanded strain value is precisely achieved. Specimens in this series were tested at the same nominal strain rate, using identical machine settings (span of ram travel fixed at a constant value). Larger grain size and higher temperature implies softer steel, so the trend of results in figure 8 reflects the 100-kN machine's ability to meet a commanded value of compressive strain at high strain rate, and how that is affected by the specimen response.

4.3.4.5 Adiabatic Heating

Figure 9 is an example of specimen self-heating during a typical compression test at a strain rate of 50 s^{-1} . Experiments indicate that the temperature rise, ΔT , at the strain of 0.7 ranges from 5 to 10°C , with modest variation depending on test conditions. Normally, ΔT values less than 10°C may be considered negligible for engineering purposes. On the other hand, the observed values here are certainly smaller than expected. Based on predictions from mechanical-equivalence equations assuming all deformation energy is converted into heat as described in Section 3.6, the calculated ΔT is about 20°C . The disparity may be due to nonuniform heating throughout the specimen. Therefore, additional analysis is needed to clarify the role of adiabatic heating in hot-deformation studies. Because of this uncertainty, the stress-strain results presented in this report are not adjusted for the adiabatic heating correction.

5. Results and Discussion

5.1 Raw Data and Data Analysis

During each test, the load, actuator's displacement, temperature and time were measured, recorded by the storage oscilloscope, and saved onto floppy disks for processing. Outputs of interest are the curves of true stress as a function of true strain, the true strain rate as a function of true strain, and the temperature change during deformation. Information on the software programs for the various data reduction and analysis options is given elsewhere [36].

5.2 Round-Robin Test Program with UBC

A round-robin test program with UBC was initiated to compare the test results using the respective laboratory facilities and analysis techniques. Two steels, the A36 and DQSK grades, from the same slabs after rough rolling were used for the program. The round-robin test program is important because there are no published test standards for the high-temperature and high-

strain rate compression tests, and test procedures vary widely from laboratory to laboratory. Consistent and agreeable stress-strain data generated from different laboratories are essential for any modeling effort. The cylindrical specimens used were 15 mm in height and 10 mm in diameter. The testing parameters included 3 strain rates, 0.1, 1, and 10 s^{-1} , and 4 temperatures, 900, 1000, 1100, and 1200 °C.

Typical test results are presented in figure 10 . The flow stresses at a strain of 0.5 are tabulated in Table 3. In general, the agreement between the results generated in the two laboratories is good. This is within the variability observed for the same steel tested with the same machine.

The flow stresses at a strain of 0.5 were used to fit eq (15). The results are presented in Table 4. Here again the agreement between the results from the two laboratories is reasonably good. The results also agree well with those published in the literature.

However, there is a consistent difference observed in the results from the two laboratories; that is the hardening behavior at strains below about 0.1. This is due to the difference in the way strain has been calculated. At UBC, the uniaxial strain is inferred from the direct measurement of diametrical changes during testing. The measurement is performed directly on the specimen at the mid-section of the specimen. At NIST, the axial strain is calculated from the displacement signals of load train with the assumption of incompressibility of steel during plastic deformation (constant volume assumption). Implicit in this strain calculation is the assumption of uniform deformation.

As mentioned previously, the displacement signals measured at NIST are not taken directly on the specimen. The displacement signals also contain the displacement of the load

Table 3. Summary of test results from UBC and NIST.

Steel	$\dot{\epsilon}$, s ⁻¹	T, °C	σ at $\epsilon = 0.5$, MPa	
			UBC	NIST
DQSK	1	900	135	154
		1000	98	108
		1100	71	66
		1200	43	44
	10	900	171	174
		1000	129	138
		1100	103	102
		1200	75	74
A36	1	900	133	133
		1000	95	89
		1100	66	57
		1200	46	44
	10	900	171	165
		1000	120	131
		1100	104	104
		1200	80	76

Table 4. Constants in hyperbolic sine equation for DQSK and A36 steels.

Steel	Laboratory	α , (MPa) ⁻¹	Q , (KJ/mol)	m	$\ln A$
DQSK	UBC	0.013	293	4.07	25.6
	NIST	0.013	326	4.41	28.4
A36	UBC	0.013	314	4.34	27.6
	NIST	0.013	321	4.39	28.4
C-Mn	Literature	0.013	262-366	3.2-5.0	25-32

train in addition to the displacement due to specimen. With rigid machines such as those at NIST, the displacement due to the load train is negligible. However, even this negligible displacement due to the machine becomes noticeable at small strains. This results in a slight overestimation of strain in the results produced at NIST. The correction for displacement due to machine compliance will be discussed in Section 5.3, "Machine Compliance".

5.3 Machine Compliance

In strain measurements, the accurate way is to measure the displacement, from which strain is calculated, directly on the specimen with strain gages, extensometers, or optical devices. In our experiments, however, the strain has been calculated using displacement signals that are not taken directly on the specimen (due to experimental difficulties); but away from the specimen. The experimental difficulties arise because of space constraint resulting from heating furnace, high temperature, and large strain. The displacement signals away from the specimen will consist of the displacement due to the specimen plus the displacement of the load train.

Our experimental setup has the displacement measured at a location about 1 meter away from the specimen with a linear variable displacement transducer (LVDT). The displacement of the load train typically consists of two parts. First, the loose contact between the anvils and the specimen at the beginning of test. This portion of the displacement will vary from specimen to specimen and the magnitude is relatively large. The other is the elastic deformation of the load train. This portion of the displacement is proportional to the load applied and is reproducible from test to test.

Using aluminum specimens at room temperature (investigation at high temperature is difficult to conduct if not impossible) we determined the displacement of the load train as a function of load experimentally. In this experiment, a clip-on gage was attached to measure the gap between the two anvil surfaces that compress the specimen. This device measured the actual specimen displacement. At the same time the displacement was recorded away from the specimen just like those experiments conducted at high temperatures. The difference between these two quantities is the displacement due to machine. Except for those results in Section 5.2, the stress-strain results presented in this report have been corrected for the machine displacement.

5.4 Mathematical Description of Stress-Strain Curves

The goal of the present study was to provide models that estimated the stress-strain curves of steels under hot-rolling conditions. These models were then used to calculate the roll-separating forces during rolling, and to provide data as input to the numerical tools, such as finite-element method, for detailed calculations of stress, strain, and related properties. The focus had been on modeling the stress-strain behavior as functions of temperature, strain rate, and austenite grain size.

Besides chemical composition, there are three major parameters that have significant influence on the flow behavior of steels: temperature, strain rate, and austenite grain size. For models to be useful, one has to consider these parameters explicitly. In this study, there were not enough systematic variations of chemical compositions to provide enough data for modeling. Thus, there was no modeling effort in this area and each grade of steel would have its own model that considers the effects of temperature, strain rate, and austenite grain size.

The general trends of the effects of temperature, strain rate, and austenite grain size on flow behavior are given in figures 11 through 13. As expected, decreasing temperature increases flow stress (figure 11). Also, increasing strain rate increases flow stress (figure 12).

The influence of austenite grain size is more complicated. As indicated in figure 13, grain-size reduction promotes dynamic recrystallization. Grain size reduction also increases the flow stress and the rate of hardening in the early stage of deformation. As for the effects of grain size on the steady-state flow stresses, either with or without dynamic recrystallization, the results are not very conclusive. However, observations tend to support the theory that grain size has little effect on the steady-state flow stresses [22,27]. Equilibrium subgrain structures are established in the early stage of deformation. The process of subgrain structure development controls early stage of deformation and the equilibrium subgrain structure controls the steady-state stresses. So, the initial grain size has little influence on the steady-state flow stresses.

To describe the stress-strain curve mathematically, our initial choice was eq (9), which fitted the elevated-temperature experimental data well for the stress-strain portion that was prior to dynamic recrystallization. Later, UBC suggested that we use the rigid-plastic model to describe the flow behavior and incorporate yield stress at the start of the stress-strain curve. Thus, eq (9) was modified and the following equation was used,

$$\sigma_1 = \sigma_0 + (\sigma_{ss1} - \sigma_0)[1 - \exp(-\beta \cdot \epsilon)]^n. \quad (23)$$

Here, σ_0 is the 0.2% offset yield stress in MPa, σ_{ss1} is the steady-state stress with dynamic recrystallization in MPa, β is a constant related to the hardening behavior, and n is a constant related to the relationship between applied stress and dislocation density.

The hyperbolic sine equation (eq (15)) is used to describe the combined effects of temperature and strain rate on the steady-state stresses. Grain size is considered to have no influence on the steady-state stresses. The effects of temperature, strain rate, and grain size on σ_0 , and β are treated separately. The constant n is considered to be independent of temperature, strain rate, and grain size.

Different equation forms have been proposed to treat the effects of temperature, strain rate, and grain size on σ_0 , and β . The task has proven to be difficult because the equations must describe the effects that follow some physical principles and still cover a wide range of conditions. The equations are required to meet the following conditions:

- σ_0 increases with decreasing temperature, increasing strain rate, and decreasing grain size.
- Within the range of interest, β increases with increasing temperature, decreasing strain rate, and decreasing grain size.
- Within the experimental range, where the machines are capable of generating reliable data, the equations should be able to reproduce the stress-strain curves.
- Outside the experimental range, where no reliable experimental data exist, the equations should be able to estimate the stress-strain curves reasonably well. The operational

ranges are temperature from 800 to 1300 °C, strain rate from 1 to 400 s⁻¹, and grain size from 0.02 to 0.5 mm.

After some trials and errors, we found that the power law appears to perform well and the following equations are used:

$$\sigma_0 = a_1 \left(\frac{T}{T_0} \right)^{a_2} \left(\frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right)^{a_3} \left(\frac{d}{d_0} \right)^{a_4}, \quad (24)$$

$$\beta = b_1 \left(\frac{T}{T_0} \right)^{b_2} \left(\frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right)^{b_3} \left(\frac{d}{d_0} \right)^{b_4}. \quad (25)$$

a_i and b_i are constants to be determined experimentally and T_0 , $\dot{\varepsilon}_0$, and d_0 are scaling factors. T , $\dot{\varepsilon}$, and d are temperature, strain rate, and grain size in K, s⁻¹, and mm, respectively.

To account for the softening due to dynamic recrystallization, the following equation is introduced.

$$\Delta\sigma = \max \left[0, (\sigma_{ss1} - \sigma_{ss2}) \left(1 - \exp \left(- \left(\frac{\varepsilon}{\varepsilon^*} \right)^{m_4} \right) \right) \right]. \quad (26)$$

$\Delta\sigma$ is the softening amount due to dynamic recrystallization after the peak stress in the stress-strain curve has been reached. ε^* is a scaling factor which is a function of temperature, strain rate, and grain size, and m_4 is a constant to be determined experimentally. The following simple power law is used to describe the dependence of ε^* on temperature, strain rate, and grain size,

$$\varepsilon^* = \eta_1 \left(\frac{T}{T_0} \right)^{\eta_2} \left(\frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right)^{\eta_3} \left(\frac{d}{d_0} \right)^{\eta_4}. \quad (27)$$

σ_{ss2} is the steady-state stress after the occurrence of dynamic recrystallization. The dependence of σ_{ss2} on temperature, strain rate, and grain size is treated with the hyperbolic sine equation (eq (15)). Grain size is considered to have no influence on σ_{ss2} .

Equation (27) is subtracted from eq (23) after the peak stress in the stress-strain curve is reached, and the overall stress component becomes

$$\sigma = \sigma_1 - \Delta\sigma. \quad (28)$$

The determinations of the strain at peak stress and the boundary separating dynamic recrystallization and no dynamic recrystallization are treated in Sections 5.7 and 5.8 later in this report.

Combining eq (18) and eqs (23) through (28) will give a full description of the stress-strain curve. The subsequent task is to determine the coefficients in these equations.

5.5 Statistical Regression for Attainment of Equation Coefficients

At the beginning of the modeling effort in this study, experimental data were fitted to obtain the coefficients in eq (23) for each individual specimen. Then, coefficients in other equations were obtained individually and separately. This procedure has the advantage for detailed analyses of the effects of individual parameter on the overall stress-strain curve, and of the deformation mechanisms. But, it is very tedious and time-consuming. For the purpose of this study, which is to provide models that estimate the stress-strain curve for hot-rolling simulations, some statistical techniques are available that estimate all the coefficients in all the equations with one computer run.

This latter approach, which was adopted in this study, has the advantage of being fast to obtain the coefficients, that will describe the stress-strain behavior adequately. It also tends to smooth out the experimental abnormalities. However, it has drawback, which loses the details of the effects of individual parameter. Also, the original meanings of some parameters are lost. These include the steady-state stresses without or after dynamic recrystallization. And the significance of the value of ε^* , which was originally proposed to be the critical strain for dynamic recrystallization [9], does no longer exist. It is therefore to caution that, as presented in this report, they should not be interpreted as their intended original meanings. They must be treated only as fitting parameters.

5.5.1 Parameter Estimation

The empirical mathematical model just described has 20 adjustable parameters, with σ_{ss1} and σ_{ss2} expressed as follows,

$$\sigma_{ss1} = \frac{\sinh^{-1}(k_1 Z_1^{k_2})}{\alpha}, \quad (29)$$

and

$$\sigma_{ss2} = \frac{\sinh^{-1}(k_3 Z_2^{k_4})}{\alpha}. \quad (30)$$

Here, Z_i ($i = 1$ and 2) is the Zener-Hollomon parameter which is expressed as

$$Z_i = \dot{\varepsilon} \exp\left[\frac{Q_i}{RT}\right]. \quad (31)$$

For each material, there are K experimental stress-strain curves. The empirical model is fitted to the data in two stages. Yield stress, as described in eq (24), is modeled and fitted separately. To estimate the constants, a_1 , a_2 , a_3 , and a_4 , we minimize

$$L = \sum_{k=1}^K \left(\hat{\sigma}_0(k) - \sigma_0(k) \right)^2. \quad (32)$$

In this expression, $\hat{\sigma}_0(k)$ and $\sigma_0(k)$ denote the predicted and observed true stress for the k -th curve (at true strain equal to zero). We minimize L using quasi-Newton method where the Hessian matrix and gradient are approximated numerically [37].

Once the fitted values of α_i , $i = 1, 2, 3, 4$, are obtained, we determine the value of the other 16 parameters by a penalized least squares fit procedure. We minimize

$$L = \sum_{k=1}^K \sum_{i=1}^{N_k} \left[\left(\hat{\sigma}(k, i) - \sigma(k, i) \right)^2 + \lambda_k \left(\hat{\sigma}_{ss2}(k, i) - \sigma_{ss2}^G(k) \right)^2 \right]. \quad (33)$$

Above, $\sigma(k, i)$ and $\hat{\sigma}(k, i)$ denote the measured and predicted true strain for the i -th sample in the k -th experimental curve. For the k -th curve, there are N_k samples in the stress-strain curve. The first term on the right hand side is the sum of the squared prediction errors. The second term is a penalty function which increases L when the predicted asymptotic stress, $\hat{\sigma}_{ss2}$, varies from a reasonable guess, σ_{ss2}^G of what the asymptotic stress should be. The relative contribution of the penalty function is controlled by the value of the (positive) parameter λ_k . Estimates of $\hat{\sigma}_{ss2}$ are not available for all experimental curves. For curves where there is no prior information, we set $\lambda_k = 0$. For cases where we have prior information, we set $\lambda_k = 0.1$. For the cases studied, for $\lambda_k \leq 0.1$, the agreement between the observed and predicted data is about the same. However, when extrapolating the model to higher strains, our judgement is that a nonzero value of λ_k improves the quality of the prediction.

In the fitting procedure, the parameters are constrained to fall between the lower and upper bounds that we select. For the i -th parameter, if the bounds are L_i and U_i , we express the constrained parameter α_i in terms of an unconstrained parameter q_i as follows

$$\alpha_i = L_i + \frac{U_i - L_i}{1 + \exp(q_i)}. \quad (34)$$

We minimize L as a function of the unconstrained parameters using a quasi-Newton method where the Hessian matrix and gradient are approximated numerically [37]. In a complicated optimization problem such as we have, results may depend on the initial guess for the model parameters, and the type of algorithm used to estimate the parameters. As a check, for each material, we refit the model at least 20 additional times. In each additional run, the initial parameter values were randomly perturbed. In all cases, additional runs did not yield a lower value of L . However, sometimes a large value of L was obtained.

The procedure just described above was used to estimate all of the coefficients for grades HSLA-80/Ti-Nb, HSLA-Nb, DQSK, and A36. The stress-strain behaviors of grades HSLA-50/Ti-Nb, and the two IF steels are found to be similar to that of HSLA-Nb. As a result, it is decided that the model for HSLA-Nb is also used for these steels. For the same reason, the DQSK model is used for the HSLA-V grade. The units for some of the parameters in the model are σ in MPa, $\dot{\epsilon}$ in s^{-1} , T in Kelvin (K), d in mm and Q in $\text{J}\cdot\text{mol}^{-1}$. R is the universal gas constant, which has the value of $8.314 \text{ J}\cdot\text{mol}^{-1}\cdot\text{K}^{-1}$. The range of experimental data used for statistical regression is tabulated in Table 5.

Table 5. Range of experimental data used for statistical regression.

	$T, ^\circ\text{C}$	$\dot{\epsilon}, \text{s}^{-1}$	d, mm
HSLA-80/Ti-Nb	900-1200	1-25	0.012-0.2
HSLA-Nb	900-1200	1-50	0.015-0.249
DQSK	900-1200	1-50	0.04-0.29
A36	900-1200	1-50	0.014-0.244

5.6. Experimental Stress-Strain Curves and Model Predictions

This section presents the results of model predictions versus experimental data for each individual grade of steel. The experimental conditions are given on the graphs.

5.6.1. A36

Typical results for A36 are presented in figure 14. In general, the agreement between the experimental data and the model predictions is excellent.

5.6.2 DQSK and HSLA-V

Typical results for DQSK are presented in figure 15. In general, the agreement between the experimental data and the model predictions is excellent.

For HSLA-V grade, the test matrix covers a range of temperature from 900 to 1200°C, strain rate from 1 to 50 s^{-1} , and grain size from 0.03 to 0.3 mm. After completing the experiments, we compared the data of HSLA-V specimens with predictions obtained using the DQSK model. Figure 16 presents some of the typical results showing that the DQSK model predictions agree well with the HSLA-V experimental data.

We believe that vanadium has little effect on the flow response of the steel in the process variable ranges of interest, and that the model developed for the DQSK grade steel is suitable for the HSLA-V steel because the two steels have similar chemical compositions. Figure 16 presents data that support these expectations. As shown in Table 6, detailed analysis of

Table 6. Dynamic recrystallization behavior of DQSK and HSLA-V steels.

		DQSK		HSLA-V	
Temp °C	Strain rate s ⁻¹	Grain size mm	ϵ_p^*	Grain size mm	ϵ_p^*
1200	10	0.29	0.3	0.3	0.33
1200	1	0.29	0.17	0.3	0.185
1100	25	0.114	0.44	0.1	0.42
1100	10	0.29	0.33	0.3	0.35
1100	10	0.114	0.337	0.1	0.335
1100	10	0.114	0.342	0.11	0.335
1100	10	0.114	0.34	0.1	0.335
1100	10	0.114	0.336	0.1	0.335
1100	1	0.29	0.22	0.3	0.235
1100	1	0.114	0.195	0.1	0.225
1000	1	0.29	0.321	0.3	0.33
1000	1	0.114	0.26	0.1	0.3
900	1	0.04	0.331	0.03	0.335

* Strain at which peak stress occurs in the stress-strain curves. This table lists only those specimens that have indication of dynamic recrystallization.

dynamic-recrystallization in the two grades further demonstrates the similarities in flow behavior for these two grades. Therefore, we used exactly the same model for HSLA-V and DQSK.

5.6.3. HSLA-Nb, HSLA-50/Ti-Nb and IF steels

Typical results for HSLA-Nb grade are presented in figure 17. In general, the agreement between the experimental data and model predictions is excellent.

Selected experiments with a scaled down test matrix were completed for HSLA-50/Ti-Nb grade. The experimental conditions span temperatures from 900 to 1150 °C, strain rates from 1

to 25 s^{-1} , and grain sizes from 0.013 to 0.061 mm. We expected the flow behavior of the HSLA-50 grade to be similar to that of HSLA-Nb because the compositions of the two grades are similar. Therefore, the series of experiments performed using this grade was less extensive than those for other grades.

Some of flow stress measurements for HSLA-50 were compared with predictions using the model developed for HSLA-Nb and the results are presented in figure 18. As expected, the flow curves agree well with predictions from the HSLA-Nb model. On this basis, we conclude that the model developed for the HSLA-Nb grade is in fact suitable for HSLA-50 also.

Two IF steel grades, Nb rich and Nb lean, were tested. Identical experimental conditions were used for both grades. Our results lead to the conclusion that the flow behaviors of both grades of IF steel are similar to that of the HSLA-Nb grade within the range of experimental variables studied.

For the Nb-rich IF grade, figure 19 shows some of the results, wherein the experimental stress-strain curves for the IF grade are compared with predictions made using the model for the HSLA-Nb grade. The flow behavior of this IF grade is very similar to that of the HSLA-Nb grade. The same conclusion is obtained from torsion experiments conducted at UBC. The results of the torsion experiments are presented in figure 20. We conclude that the model developed for the HSLA-Nb grade also predicts the flow behavior of the Nb-rich IF steel.

For the Nb-lean grade, figure 21 shows some of the experimental data compared with predictions made using the HSLA-Nb model. We find that the flow behavior of this grade is also similar to that of the HSLA-Nb. However, results obtained in multiple-hit torsion tests at UBC show that the flow stresses of this grade are lower than those of the Nb-rich IF grade when temperatures are below 980°C and total true strains are larger than 1.2.

Scrutiny of the torsion test results has indicated that little softening takes place in Nb-rich grade between hits at the conditions which produce higher flow stresses, whereas substantial softening occurs in the Nb-lean grade. This means that there is little static recrystallization occurring between hits for the Nb-rich grade, whereas the Nb-lean grade almost fully recrystallized. This is probably due to Nb in retarding the static recrystallization for the Nb-rich grade and explains why the Nb-rich grade has higher flow stresses in later stages of deformation in multiple-hit experiments and at low temperatures. We also note that there is evidence showing phase transformation from austenite to ferrite at the test condition of 900°C and 1 s^{-1} in the Nb-lean grade. However, at strain rates of 5 and 25 s^{-1} (900°C), there is no indication of phase transformation.

5.6.4. HSLA-80/Ti-Nb

Typical results for HSLA-Nb grade are presented in figure 22. In general, the agreement between the experimental data and model predictions is excellent.

5.7. Strain at Peak Stress

Dynamic recrystallization is an important phenomenon in hot deformation. The occurrence of dynamic recrystallization not only softens material's resistance to deformation, it also refines the microstructure. It is useful to estimate the conditions under which it will occur.

Based on the experimental data generated in this study, we developed models correlating the strain at the peak stress in the stress-strain curve with temperature, strain rate, and grain size for the A36, DQSK, HSLA-Nb, and HSLA-80 grades. Two different forms of equations were used to make the correlation. One used the power law, and the other used the conventional approach involving the Zener-Hollomon parameter. For the power-law approach, we used the following equation:

$$\varepsilon_p = A_1 \left(\frac{T}{T_0} \right)^{A_2} \left(\frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right)^{A_3} \left(\frac{d}{d_0} \right)^{A_4}, \quad (35)$$

where ε_p is the strain at which the peak stress appears in the stress-strain curve. A_1 , A_2 , A_3 , and A_4 are constants. Other parameters are previously defined in Section 5.4. The results are plotted in figure 23.

For the conventional approach incorporating the Zener-Hollomon parameter, we used the following equation:

$$\varepsilon_p = B_1 Z^{B_2} d^{B_3}, \quad (36)$$

where B_1 , B_2 , and B_3 are constants. Other parameters are previously defined. The results are plotted in figure 24. It should be noted that the reported value of B_2 in the literature is 0.15, while the values for B_3 range from 0.18 to 0.5 [29,38]. As shown in figures 23 and 24, the two approaches have similar statistical correlations.

5.8. Boundary Separating Dynamic Recrystallization and No-Dynamic Recrystallization

As described in Section 5.7, if strains exceed ε_p as described by eq (35) or (36) the steel will undergo dynamic recrystallization. However, there is experimental evidence suggesting that under certain conditions, dynamic recrystallization will not occur with a single deformation even though strains exceed those described by eq (35) or (36). To determine whether dynamic recrystallization will occur during a single deformation regardless of the conditions set forth by eq (35) or (36), we use the Zener-Hollomon parameter as a criterion. If Z is greater than a critical value, Z_c , then dynamic recrystallization will not occur in a single deformation. The model, as shown below, also takes grain size into consideration

$$Z_c = A \cdot \exp(B \cdot d) + C, \quad (37)$$

where A , B , and C are experimentally determined constants. Z and d are defined previously. The values of the constants have been determined here.

5.9 Effects of Prior Heat Treatments: HSLA-Nb

We also conducted experiments to evaluate the effects of precipitates on the flow behavior of HSLA-Nb steel. In earlier experiments we obtained small austenite grain sizes using low reheating temperatures. Under these conditions, the Nb precipitates were not dissolved in the matrix when the experiments were conducted. This represents a departure from mill conditions where the reheating temperature is high and the Nb precipitates are dissolved.

Specimens used in this series of experiments were reheated to 1200°C, held for 30 minutes, and water quenched to the ambient temperature. The majority of the precipitates would be dissolved after this treatment. The specimens were then reheated to 1000°C with a heating rate of 5°Cs⁻¹, held at that temperature for 150 seconds, cooled to 900° at a rate of 1°Cs⁻¹, and held for 30 seconds before testing. In tests at 1000°C, the specimens were tested immediately after the 150 seconds hold at that temperature.

Figure 25 presents a few typical experimental results. For some test conditions there is no discernible difference in the flow behavior with or without prior heat treatment. However, at the test condition of 900°C and 1 s⁻¹ (figure 25), there is a significant difference in flow behavior after a strain of 0.4. In this case, the specimen without prior heat treatment shows evidence of dynamic recrystallization, whereas the specimen with prior heat treatment shows continuous hardening with no indication of dynamic recrystallization. It is likely that the dissolved Nb retards the kinetics of dynamic recrystallization in the specimen with heat treatment.

Although mill conditions will not produce a strain rate of 1 s⁻¹ at 900°C, the observed experimental effect deserves closer attention. The effect vanishes at higher strain rates or higher temperatures.

5.10 Effects of Reheating Temperature and Model Adjustments: HSLA-80/Ti-Nb

In comparison with those samples that were reheated to 1200°C, UBC has observed that in torsion experiments, the flow stresses are lower when samples were reheated to 1250°C and held for 30 minutes. To confirm this observation, we conducted experiments with the following two different reheating conditions:

1. Reheated at 1250°C and held 10 minutes to obtain a grain size of 0.3 mm
2. Reheated at 1200°C and held 10 seconds to obtain a grain size of 0.1 mm

Typical results of the experiments are presented in figure 26. It is found that under condition 2 there is no significant difference between the experimental data and the model predictions, which have been based on the experiments using lower reheating temperatures. However, a higher reheating temperature and a longer holding time (condition 1), appears to lower the flow stress.

The effects of reheating temperature on the flow behavior of HSLA-80 were examined further in experiments using different test temperatures (unreported). The flow stresses for specimens with the reheating temperature of 1250°C and holding time of 10 minutes appear to be

approximately 9% lower than those for specimens reheated at a lower reheat temperature. This 9% drop is consistent throughout the whole stress-strain curve and has been observed for specimens tested at various temperatures. Thus, the model described in Section 5.5 should be adjusted for this effect if the reheating temperature and holding time are high and long.

5.11 Double-Hit Experiments: HSLA-Nb

Using a double-hit experimental technique, we evaluated the effects of retained strain on the flow behavior of HSLA-Nb steel. The strain retained from one pass to the next significantly increases the rolling forces but it also affects microstructural development. The retained strain is

Table 7. Test matrix for double-hit experiments: HSLA-Nb.

Case #	Soaking temp., °C	Soaking time, s	Grain size, mm	Testing temp., °C	Strain for the first hit	Strain rate, s ⁻¹	Inter-hit time, s
1	1200	600	0.249	900	0.2	10	20
2	1200	600	0.249	900	0.6	10	10
3	1200	600	0.249	900	0.6	25	5
4	1200	600	0.249	1000	0.2	10	5
5	1000	150	0.015	900	0.2	10	5

usually small, and the effects are negligible in plain-carbon (A36 or DQSK grades) and V-treated steels because static recrystallization is rapid in these steels. Retained strain is important for the HSLA-Nb grade in the last few finishing passes because the interpass time is short and the process of static recrystallization is comparatively sluggish.

The test matrix of the double-hit experiments is given in Table 7. Figure 27 presents the results which suggest that, under the experimental conditions described in Table 7, little softening takes place between the two hits. As indicated in these figures, the established model predicts the flow behavior well if the retained strain is taken into consideration.

5.12 Comparison With Data from CANMET

From CANMET we received sets of high-strain rate stress-strain data for three steels: A36, DQSK, and HSLA-Nb. These experimental data covered temperatures from 900 to 1100°C and strain rates from 10 to 150 s⁻¹. Austenite grain sizes ranged from 0.014 to 0.250 mm for A36, from 0.038 to 0.155 mm for DQSK, and from 0.015 to 0.249 mm for HSLA-Nb.

We compared CANMET's data with our own model predictions using the experimental data for A36 and DQSK. Figure 28 shows examples of comparison between CANMET's data and our model predictions for the DQSK grade. The agreement is generally good for stresses in high strain regions (strains larger than 0.6). However, poor agreement is observed in the small strain regions (strains less than 0.4).

In general, the agreement between CANMET's data and our model predictions for A36 is not as good as that for DQSK, and in a few cases the agreement is poor for the entire stress-strain curve, as shown for example in figure 29.

Two reasons are advanced to explain the disagreement observed at small strains. First, CANMET's strain values include displacements due to machine compliance, which are accounted for in NIST's analyses. Including displacements due to machine compliance in strain calculations promotes large errors in the stress-strain curve at small strains. Secondly, we have assumed rigid-plastic behavior in our models, so the flow stress begins at the yield stress when strain is zero.

In view of these differences in data treatment, we performed a second comparison. We compared our experimental data (without machine compliance correction) with CANMET's data at similar test conditions. The results in figures 30 through 32 show fair agreement. CANMET's data tend to exhibit lower strain hardening at the onset of deformation and the hardening continues to large deformations (large strains). In contrast, our own data tend to show a rapid increase in stress due to a high strain hardening rate at the onset of deformation, and thereafter the stress levels off at large deformations. Thus, CANMET's data exhibit lower stresses at small strain regions and higher stresses at large strain regions. The cause of this difference in hardening behaviors between results obtained at CANMET and NIST remains unclear. It should be noted that the hardening behavior observed at NIST is more consistent with some findings reported at UBC.

Efforts have been made to correct for the effect of machine compliance in CANMET's data. When machine compliance is accounted for, the stresses in the early portion of the flow curve increase; thus the agreement between NIST's model predictions (or data) and CANMET's data improves. Figure 32 is a typical example showing good agreement between NIST's model prediction and CANMET's data for A36 after compliance correction.

5.13 Calculation of Roll Separating Forces

One of the important applications of constitutive equations in hot rolling is the roll-force calculation, which requires other parameters, such as roll diameter during rolling, thickness of slab at roll entry and roll exit, and the friction between the workpiece and the rolls. One of the commonly used models for roll-force calculations is the Sims method [39], which made some simplifications in the slab analysis [40] and the assumption that only sticking friction exists in the roll contact zone. The roll separating force per unit width is given by Sims as

$$F = 2kR' \left[\frac{\pi}{2} \sqrt{\frac{h_2}{R'}} \tan^{-1} \sqrt{\frac{\Delta h}{h_2}} - \frac{\pi\phi}{4} - \ln \frac{h(\phi_n)}{h_2} + \frac{1}{2} \ln \left(\frac{h_1}{h_2} \right) \right], \quad (38)$$

where k is defined as the yield strength of the workpiece in pure shear, R' is the radius of deformed roll, h_1 and h_2 are the entry and the exit slab thicknesses, Δh is the draft which equals to $(h_1 - h_2)$, ϕ is the contact angle which is approximated as

$$\phi = \sqrt{R' \Delta h}, \quad (39)$$

and $h(\phi_n)$ is the slab thickness at the neutral point, which is the point along the surface of contact between the roll and the workpiece that has the surface velocity of the roll equal to the velocity of the workpiece.

According to von Mises yielding criterion, k is equal to 57.7% of the yield strength of the workpiece in uniaxial tension or compression; it is 50% if Tresca criterion is applied. In this investigation, von Mises criterion is adopted and the uniaxial compression yield strength is taken to be the stress calculated with models presented in Section 5.5. The strain used to calculate the

Table 8. Comparison of calculated roll forces with actual mill measurements.

Pass #	Roll forces (ton)		
	Measured	Calculated	% difference
1	3324	3608	8.54%
2	3456	3326	-3.76
3	3690	3646	-1.19
4	3113	3226	3.63
5	2678	2957	10.42
6	2137	2370	10.90
7	1827	2035	11.38
8	2990	3046	1.87
9	3080	3182	3.31
10	2720	3153	15.92
11	2610	2587	-0.88
12	2428	2754	13.43
13	2250	2393	6.36

stress is the true strain computed from h_1 and h_2 or the cumulative strain if there are retained strains from previous passes.

The value of R' is dependent on the roll material, the original roll radius, R , the roll separating force, and the draft. The relationship among these parameters is given as [26]

$$\frac{R'}{R} = 1 + \frac{CF}{\Delta h}. \quad (40)$$

For steel rolls, the value of C is between 0.00021 and 0.00025 mm²/N, and is taken to be 0.000221 in the present investigation [41].

The angle at neutral point is calculated from

$$\frac{\pi}{4} \ln\left(\frac{h_1}{h_2}\right) = 2\sqrt{\frac{R'}{h_2}} \tan^{-1}\left(\sqrt{\frac{R'}{h_2}} \phi_n\right) - \sqrt{\frac{R'}{h_2}} \tan^{-1}\left(\sqrt{\frac{\Delta h}{h_2}}\right). \quad (41)$$

The thickness at neutral point is

$$h(\phi_n) = 2R'(1 - \cos(\phi_n)) + h_2. \quad (42)$$

The results of roll-force calculations along with actual mill measurements for a 13-pass rolling on a HSLA-80 steel are presented in Table 8 and figure 33. Based on precipitation calculation using the model developed by Dutta and Sellars [42], precipitation of Nb(CN) particles occurred after the first pass at 965 °C. No static recrystallization between passes in all the 13 passes was assumed and strain was accumulated from one pass to the subsequent ones. The predicted results agree reasonably well with the actual measurements. Due to the assumption of sticking friction, the Sims equations, however, are expected to slightly overestimate the rolling loads.

6. Summary and Conclusions

The purpose of this work was to develop constitutive equations that predict the stress-strain curves of steels as functions of processing variables, such as temperature, strain rate, and microstructural features under hot-rolling conditions. Eight steels, DQSK, A36, HSLA-Nb, HSLA-50/Ti-Nb, HSLA-80/Ti-Nb and two interstitial-free (IF) grades, were selectively tested at temperatures in the range from 900 to 1200 °C, strain rates from 1 to 50 s⁻¹, and austenite grain sizes from 0.012 to 0.29 mm. The data were analyzed to support the development of constitutive models.

To cover a wider range of strain rates, a computer-controlled servo-hydraulic laboratory apparatus was designed and built for high strain-rate experiments. A constant strain rate of 65 s⁻¹ has been obtained with the machine. The usable strain range at this rate is about 0.4. The strain rate drops off substantially at strains larger than 0.4. At a constant strain rate of 50 s⁻¹, the usable range is about 0.6. The strain rate of 50 s⁻¹ has been used for the majority of tests in this project. The machine has a load capacity of 100 kN and its radiant furnace can heat a specimen (a cylinder with 10 mm diameter and 15 mm long) to 1200 °C. The maximum heating rate is about 2 °C/s. The machine has multiple-strike capabilities. The machine is equipped with a rough-vacuum pump to prevent oxidation during test at high temperatures. The machine and data acquisition are computer controlled.

Models for predicting stress-strain behaviors for A36, DQSK, HSLA-Nb, and HSLA-80/Ti-Nb have been developed. The model developed for DQSK is also adequate for HSLA-V, since we found vanadium had little influence on the flow behavior of low-carbon steels, and the chemical compositions of the two grades were similar (except V addition in the HSLA-V grade). The model developed for HSLA-Nb was also found to be suitable for predicting the stress-strain behaviors of HSLA-50/Ti-Nb, and two Nb-treated IF steels, which all contained niobium. Model

predictions correlated well with the experimental data, indicating that the models adequately describe the stress-strain behaviors of various steel grades under hot-rolling conditions.

Models were also developed for correlating the strain at peak stress in the stress-strain curve with temperature, strain rate, and austenite grain size for the following four steels: A36, DQSK, HSLA-Nb and HSLA-80/Ti-Nb. In addition, for the same steels models for predicting the conditions under which there would be no dynamic recrystallization were also proposed. The latter conditions would precede the former conditions.

We also conducted experiments to evaluate the effects of precipitates on the flow behavior of HSLA-Nb grade. It is found that under one condition with low strain rate (1 s^{-1}) and low temperature (900°C), the specimen with undissolved precipitates displays the occurrence of dynamic recrystallization, while the specimen with prior high-temperature soaking heat treatment (precipitates dissolved at the time of experiment) does not show the same behavior. It is likely that the dissolved Nb retards the kinetics of dynamic recrystallization under this condition. At high temperatures and higher strain rates, there is no difference in flow behavior between the specimens (with or without the high-temperature soaking heat treatment). It should be noted that mill conditions rarely encounter a strain rate of 1 s^{-1} at 900°C .

For the HSLA-80/Ti-Nb grade, it is found that specimens with treatment of reheating at 1250°C and holding for 10 minutes (0.3 mm in grain size) reduce the flow stress about 9% when compared with those specimens that were reheated at lower temperatures. This behavior has been observed for various experiments with testing at different temperatures and strain rates. It is, therefore suggested that the 9% reduction in flow stress be corrected if the actual conditions are close to our experimental conditions (reheated at 1250°C and held for 10 minutes).

In HSLA-80/Ti-Nb, we find the static recrystallization is sluggish. Our double-hit experiments show that at testing temperatures of 900 and 1000°C and inter-hit times of 5 to 20 seconds, little softening occurs between hits. If retained strain is properly considered, we show that the model based on single-hit deformation is adequate to describe the stress-strain behaviors under multiple-hit conditions.

Our model predictions for A36 and DQSK are used to compare with Camplastometer's high strain-rate data obtained at CANMET. The CANMET's data covered the strain-rate range from 10 to 150 s^{-1} , which were beyond NIST's experimental capability. The agreement between CANMET's data and NIST's model predictions is good at large strains (> 0.6) but poor in small strain areas (< 0.4). The discrepancy mainly comes from the different data treatment in the two laboratories. NIST did correct the data for displacements due to machine compliance and used rigid-plastic model to describe stress-strain behavior, whereas CANMET did not correct for machine compliance and reported the data as generated (no stress-strain model had been used). After correction for machine compliance in CANMET's data, the agreement did improve and the results indicated that NIST's models could be extrapolated beyond the experimental range that were used to obtain the coefficients in the constitutive equations. However, cautions should be exercised in performing extrapolation because the results might not be reliable.

We used the Sims equation with NIST's constitutive model to calculate the roll separating forces for rolling an HSLA-80 slab. The calculated results were compared with the actual mill measurements. The results are encouraging. Eight of the 13 passes, the calculated loads are within 10% of the measured loads. The worst discrepancy is 15.9% overestimation. The other 4 passes, the discrepancy is slightly over 10%. Ten of the 13 passes, the calculated loads are higher than the measured. This is expected since the Sims equation assumes the sticking friction (maximum friction) between the rolls and the workpiece.

Finally, the actual value for the coefficients for the constitutive equations have not been published in this report, as these are considered proprietary data of AISI.

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Figure1. Block diagram for the hot-strip rolling operation. Rectangular blocks represent the physical stages of the operation, while oval-shape blocks represent models that describe the state of the steel or the microstructural changes at various stages of operation.

Figure 2. Stress-strain curves of DQSK steel. Test temperatures are indicated and strain rate is 10 s^{-1}

Figure 3. Basic components and flowchart of the test apparatus.

Figure 4. Triplicate tests of DQSK steel. The results show good reproducibility of the machine.

Figure 5. Stress-strain curves for two specimens of DQSK using different test machines. The results show the consistency between the two machines

Figure 6. Typical plot of strain rate versus true strain for nominal strain rates of 1 and 10 s^{-1} .

Figure 7. Typical plot of strain rate versus true strain for nominal strain rates of 25 and 50 s^{-1} .

Figure 8. Ratio of measured-to-commanded final heights for A36 steel sample tested at various temperatures and grain sizes.

Figure 9. Temperature versus true strain of an A36 steel specimen showing self-heating during a high strain-rate test.

Figure 10. Test results obtained from UBC and NIST in the round-robin test program.

Figure 11. Dependence of flow stress on temperature.

Figure 12. Dependence of flow stress on strain rate.

Figure 13. Effects of austenite grain size on flow behavior.

Figure 14. Results of model predictions vs. experimental data: A36.

Figure 15. Results of model predictions vs. experimental data: DQSK.

Figure 16. Results of model predictions vs. experimental data: HSLA-V.

Figure 17. Results of model predictions vs. experimental data: HSLA-Nb.

Figure 18. Results of model (HSLA-Nb model) predictions vs. experimental data: HSLA-50/Ti-Nb.

Figure 19. Results of model (HSLA-Nb model) predictions vs. experimental data: Nb-rich IF steel.

Soaking: 5 °C/s to 1200 °C hold 15 min; strain rate=1/s; strain=1, 0.36, 0.51, 0.41, 0.33, 0.30, 0.25, 0.10;
 interpass time=10, 8.0, 4.8, 3.2, 2.3, 1.7, 1.3 s; target temperature=1100, 1041, 979, 955, 934, 919, 908, 898 °C.

Figure 20. Torsion test results which show similar flow behavior between the Nb-rich IF steel and the HSLA-Nb grades.

Figure 21. Results of model (HSLA-Nb model) predictions vs. experimental data: Nb-rich IF steel.

Figure 22. Results of model predictions vs. experimental data: HSLA-80/Ti-Nb.

Figure 23. Results of dynamic recrystallization model predictions vs. experimental data: power-law model.

Figure 24. Results of dynamic recrystallization model predictions vs. experimental data: Zener-Hollomon parameter approach.

Figure 25. Effect of prior heat treatment on flow behavior: HSLA-Nb. There is no discernible differences in the flow behavior below strain of 0.4 between the two cases. Above 0.4, the behavior of Case 2 shows dynamic recrystallization while Case 1 shows continuous hardening. Case 1: with prior heat treatment; Case 2: without prior heat treatment.

Figure 26. Effect of reheating temperature on flow behavior: HSLA-80. The differences between the experimental data and model predictions are insignificant.

Figure 27. Results of double-hit experiments: HSLA-Nb. Little softening took place between hits. If retained strain is taken into consideration, model based on single-hit experiments is adequate for predictions of the flow behavior under multiple-hit deformations.

(a)

(b)

Figure 28. Examples of comparison between CANMET's high strain-rate results and NIST's model predictions: DQSK.

Figure 29. Example of comparison between CANMET's high strain-rate results and NIST's model predictions: A36. The agreement is poor.

Figure 30. Example of comparison between CANMET's high strain-rate data and NIST's data: A36. No machine compliance correction for both sets of data. The agreement is fair.

(a)

(b)

Figure 31. Examples of comparison between CANMET's high strain-rate data and NIST's data: DQSK. No machine compliance correction for both sets of data. The agreement is fair.

Figure 32. Comparison of CANMET's data (A36) with NIST's model prediction. It is shown that the agreement is improved after machine compliance correction.

Figure 33. Results of roll-separating-force calculations. The agreement between the measurements and predictions is excellent.

STRUCTURE –PROPERTY RELATIONSHIPS IN STEELS AT AMBIENT TEMPERATURES

P.T. Purtscher, Yi-Wen Cheng, and C.N. McCowan
Materials Reliability Division
National Institute of Standards and Technology

Abstract

The purpose of this work was to develop equations to predict the properties of steels as functions of chemistry and microstructure variables, such as ferrite grain size and fraction at ambient temperatures after hot-rolling. Eight steels, A36, DQSK, HSLA-Nb, HSLA-50/Ti-Nb, HSLA-80/Ti-Nb and two interstitial-free (IF) grades, produced by three different companies were used for evaluation of the equations.

To develop and test the equations, testing of mill-produced products was supplemented by laboratory-produced samples. For the plain-C grades, the laboratory-produced samples provided a wider range of microstructural variables than those supplied by the mills. For the HSLA grades, the laboratory treatments were successful in simulating the precipitation coarsening kinetics only.

Equations to predict the yield strength, ultimate strength, and total elongation for the eight steel grades listed above have been developed. For A36, an equation to predict the ductile-to-brittle transition temperature in the Charpy impact test has also been developed. Model predictions correlate well with the experimental data available, indicating that the equations adequately describe the properties of the eight steel grades that were considered. The equations' predictions for strength in plain-C steels are further validated with data obtained from one manufacturer. The predicted strength properties of HSLA steels were compared to those obtained from similar Nb-containing grades from two steel producers to demonstrate the application of the equations to other HSLA grades.

The equations have been used to calculate the properties in the Hot-Strip Mill Model (HSMM) developed at the University of British Columbia-Canada/

Keywords: grain size; hot-strip rolling; HSLA steels; mechanical properties of steels; plain-C steels; precipitation strengthening.

1. STRUCTURE-PROPERTY GOALS

This report summarizes the work related to development of structure-property equations at ambient temperatures. There are three main technical hurdles associated with this work.

1. characterization of fine, sometimes complex microstructural features,
2. development of non-process specific constants in the equations, and
3. the relationship between the other work being done at the UBC and the proposed structure-property equations including verification of the predictions from the resulting computer program with mill data.

The responsibility for the first two hurdles is self-contained within the subtask and is reported on here. Responsibility for the third hurdle is ultimately shared along the partners in the program, NIST, UBC, and the AISI members and can only be partially evaluated with the information available at the time.

This report is organized as follows: first, there is a short summary of the accomplishments for each year; second, there is a discussion of the approach we have taken for the development of the structure-property equations; third is a presentation of significant results for each steel grade in the program; next is a discussion of the relevant literature and how those equations are applied to the hot-strip mill including fine tuning and validation; the discussion ends with a section on the limitations and possible further applications for the equations.

2.0 ANNUAL SUMMARY OF ACTIVITIES

FY93 Began the subtask at NIST with guidance from the University of British Columbia (UBC), and U. S. Steel (USS). We performed a literature review related to this work. We have considered that the work would have three components: first, laboratory testing of crop samples simulated the range of processing conditions that are used in typical operations; second, evaluation of mill samples; and finally, development of equations to describe precipitation strengthening. We evaluated samples from the initial set of crop and coil material (DQSK, A36, V-HSLA, and Nb-HSLA grades) submitted by USS.

FY94 We continued the investigations with additional coils of A36 from USS. The steel grades to be included in the program were modified; A572 is dropped and two Nb-Ti HSLA sheet steels (350 and 550 MPa grades) as well as two IF sheet steels (Ti-rich and Nb-Ti) are added. These changes shift the emphasis to low-C sheet products, de-emphasizing the hot mill plate product with a base chemistry of 0.15-0.20% C and 0.7-0.9 Mn where the prediction of DBTT is important. An additional task for UBC has been added to extend the model's capability to include a Steckel mill configuration. No additional work related to the Steckel mill was

additional task for UBC had been added to extend the model's capability to include a Steckel mill configuration. No additional work related to the Steckel mill was expected for the prediction of structure-properties. In this fiscal year, models for the plain-C steels were selected. The properties that the model must predict included yield strength (YS), ultimate tensile strength (UTS), ductile-to-brittle transition temperature in impact testing (A36 only), and elongation (uniform and total) in a tensile test.

- FY95 We continued the investigations with additional coils of DQSK, V, and Nb steel. In this fiscal year, we selected different equations to describe the strength of plain-C steels (referred to as the base strength) to reflect a larger data base and to extend the range of applicability for the equations. The measurement of grain size in the plain-C steels was studied with a round-robin between NIST, UBC, and the steel companies. Companies submitted information on the HSLA steels that they make. This information was used to focus the structure-property subtask on the microalloyed steels of concern to participating companies.
- FY96 We continued the investigations with additional mill coils from different suppliers. We verified that the predictions from the plain-C steels would work for the base strength of HSLA steels. The laboratory testing of crop samples was dropped as the results did not correlate well with the mill coil samples. Isothermal aging of mill coils from the V HSLA grade provided all the input needed to apply the Shercliff-Ashby process model to describe the change in strength due to precipitate coarsening that occurs in the coiling model for the HSLA V grade.
- FY97 We finished the investigations with additional samples of the extreme tail end of Nb and Nb-Ti 80 grades. Results verified the torsion test results from UBC and literature data. Base strength equations from IRSID for plain-C steels were applied to IF grades without any modification; for the base UTS prediction of Nb-Ti grades, an increase in the grain-size dependence was applied for finest grain sizes and the base YS was decreased by 25 MPa. Mill data for Nb steels and Nb-Ti were used to validate the prediction of YS, UTS, and total elongation. The equations developed for the structure-property relationships are applicable to a wider range in chemistry than just the eight grades in the program. Base strength equations are considered applicable to grades where the microstructure is predominantly polygonal ferrite; prediction of precipitation strengthening is applicable to steels with Nb contents from 0.02 to 0.09, 0.08 V, and excess Ti levels up to 0.025.

3.0 STRUCTURE-PROPERTY APPROACH

The room-temperature mechanical properties, such as yield strength (YS) and the ductile-to-brittle transition temperature (DBTT), are requirements for many hot-rolled steel products. These properties are usually related to product composition and ferrite grain size. Empirical equations, based on physical metallurgy principles that relate properties to composition and grain size, are widely available for the air-cooled C-Mn steel. However, correlation of structure-properties with accelerated cooled or microalloyed precipitation hardened steels have not been well characterized in the literature.

Another problem related to the development of structure-property equations based on physical metallurgy principles is the lack of good physical models to represent a given mechanical property. The simplest mechanical property of interest would be YS, but even here, there are complications. The yield point elongation (YPElong) can vary significantly depending upon the ferrite grain size and the volume fraction of second phase, and will change the measured lower YS (LYS); yet no approach to separate the microstructural dependence of YPElong from the true structural dependence of YS exists.

The basis of most equations for structure-property relations is the assumption that the different mechanisms responsible for the properties can be described individually by the composition and microstructure and that the effects of each are additive. For example, the YS has been described as a sum of grain refinement ($k_y \times d^{-0.5}$), dislocation σ_{disl} , transformation σ_{tr} , texture σ_{tex} , precipitation σ_{ppt} , and solid solution σ_{ss} as shown below:

$$YS = \sigma_o + k_y d^{-0.5} + \sigma_{disl} + \sigma_{tr} + \sigma_{tex} + \sigma_{ppt} + \sigma_{ss} , \quad (1)$$

where σ_o is assumed to represent the lattice friction for pure Fe. This approach makes the implicit assumption that the contributions from the different mechanisms are additive. This assumption is useful from a practical point of view, and is the approach taken by most researchers. Experiments where strengthening mechanisms are not additive have been reported [1].

Examples of some of the equations for UTS found in the literature are shown in Table 1 [2-7]. Several variations on the basic approach are evident. Several groups have tried to predict hardness of the individual microstructural components from the transformation temperatures and then infer the UTS from hardness. Processing parameters like the cooling rate are sometimes used as input. Several researchers have specifically incorporated precipitation strengthening into their predictions, but without the use of the related microstructural features like the particle size and spacing between particles. The carbon equivalence (CEV) is used in one case to describe the solid solution strengthening where welded applications are of interest. An alternative and more rigorous approach to predict UTS, not shown in Table 1, has been proposed by Tomota et al. [8] where the whole stress-strain curve is predicted and the UTS is given by the point where the true

flow stress is equal to the strain hardening rate.

For this project, we have incorporated some of the details of the different approaches that seem to work best for our intended goal. In the next sections, the individual components of the equations have been discussed briefly with respect to this project.

3.1 Grain refinement

Grain size is a critical factor influencing YS; grain size is also the most important parameter that can change with different processing routes in the hot strip model to allow the predicted property to change. Yet no physical model for the interaction between grain size and YS has been proposed and validated for the type of ferritic sheet steels that we are studying. Mintz [9] has discussed the importance of the grain-size dependence (Hall-Petch slope k_y) in determining YS of steels and summarized the literature on this subject. Variations in k_y for commercial compositions range from a low of 14 to a high of 24 MPa • mm^{0.5}. When all of the C and N are removed from solid solution, values as low as 5 to 6 MPa • mm^{0.5} have been reported. Physically, the variation is almost certainly related to segregation of interstitial atoms to grain boundaries, which increases the stress needed to nucleate new dislocations from the boundary regions. Part of the variation is also related to how grain size is measured. This factor has been discussed in more detail by Gladman and Pickering [10].

The grain size dependence of the UTS is certainly less than that of YS. In the equations for YS and UTS proposed by Hodgson and Gibbs [7], the factors are 19.7 and 11 MPa • mm^{0.5}; Pickering [11,2] uses 15.1 and 7.7 MPa • mm^{0.5} for k_y in his equations for YS and UTS. In both cases, the dependence of YS on grain size is about twice that for UTS. Campbell [12] has proposed a different trend where the k_y for YS is 11.8 and 15.9 for UTS. The fact that equations from different sources will predict different trends in a given property reflects more upon the data base used to generate the equations than any fundamental characteristic of the steel's behavior.

Table 1. Summary of predictions found in the literature for UTS.

Investigator	Ultimate tensile strength, MPa
Pickering [2] 1977, England	$UTS = 294 + 27.7[Mn] + 83.2[Si] + 385X_p + 7.85d^{-1/2}$
Choquet et al. [3] 1990 France	$UTS = UTS_o + UTS_{ss} + 7.24X_\alpha d^{-1/2} + 500X_p$ $UTS_o = 237, UTS_{ss} = 29[Mn] + 79[Si] + 700[P] + 5369[N_{free}]$
Yada [4] 1988 Japan	$UTS = 3.04 H_v$, where $H_v = X_\alpha H_\alpha + X_p H_p + X_b H_b$; $H_\alpha = 361 - 0.347T_m^\alpha + 50[Si]$; $H_p = 175$; $H_b = 508 - 0.588T_m^b$
Mitchell et al. [5] 1993 England	$UTS = 74.1 + 985.1CEV + (31125[N_{free}] - 39)[V]^{1/2} + 181.5t^{-1/2}$ where $CEV = [C] + [Mn]/6 + \{[Cr]+[Mo]\}/5 + \{[Ni]+[Cu]\}/15$
Kwon [6] 1992 Korea	$UTS = 0.095X_\alpha H_\alpha + 0.264X_p H_p + 0.277X_b H_b + 11d^{-1/2} + 75.5X_\alpha d^{-1/2}$ $+ 13.4 X_\alpha (1-X_\alpha) + 3.8$ where $H_\alpha = 27.72\exp(980 / T_m^\alpha) + 39.8[Si] + 9.1$ $H_p = 215.4 - 0.016T_m^p - 34.04[Mn] + 58.26[Si]$ $H_b = 27.72\exp(980 / T_m^b) + 39.8[Si] + 9.1$.
Hodgson and Gibbs [8] 1992 Australia	$UTS = 164.9 + 634.7[C] + 53.6[Mn] + 11d^{-1/2}$

X_α, X_p , and X_b = volume fraction of α , pearlite, and bainite.

d = α grain size in mm.

H_v = Vickers hardness no.; H_α, H_p , and H_b = Vickers hardness for α , pearlite, and bainite.

T_m^α and T_m^b = average transformation temperature in °C for α and bainite.

CEV = carbon equivalent.

$[N_{free}] = [N] - 3.42[Ti]$.

t = thickness of plate in mm.

The grain-size dependence of equations in this report for elongation and DBTT, are substantial, but they are derived mainly from literature reports rather than developed from our own data base. For total elongation, the prediction varies with the UTS.

3.2 Dislocation strengthening

Dislocation strengthening has little influence on properties of most hot-rolled low-C sheet steels because the mill conditions (finishing and coiling temperatures) are such that the structure is usually polygonal ferrite, which is free of excess dislocations. With higher C steels, larger austenite grain size, and high local cooling rates, non-polygonal transformation products can form and excess dislocations are an issue for property prediction. The presence of excess dislocations in the structure (ones that will affect the prediction of properties) can be difficult to detect. Often, light microscopy can only provide hints as to the dislocation content. The best way to characterize dislocations is transmission electron microscopy (TEM), but difficulties in sample preparation and quantitative interpretation of TEM results make its use for model development limited.

For our project, A36 and HSLA 550 MPa are the only grades where non-polygonal microstructures are important, and then only for specific processing routes. Literature data [13] suggest that some modifications to the equations for polygonal ferrite microstructure are necessary to describe the strength of the non-polygonal microstructures. For the limited conditions under which dislocation strengthening are relevant, we have modified the strength equations used for polygonal structures to reflect the change in strength due to the non-polygonal microstructures. The total elongation is assumed to be influenced only by the relative change in UTS. Predictions for uniform elongation and DBTT are not changed by the non-polygonal structure because we did not have any data base to quantify the effects.

3.3 Transformation strengthening

This is one area of the program where there is difficulty to separate the individual models. The equations that describe transformation and the structure-property equations are intimately related; for example, the equations for strength from Campbell [12] require a pearlite spacing, but the transformation model does not predict that level of detail. The transformation model developed at UBC takes into consideration the steel grade (not exact composition), austenite grain size, cooling rate, and transformation heat. The output of the model gives ferrite fraction, ferrite grain size, and the microalloy content in solution at the transformation temperature. Any precipitates formed in austenite before transformation are assumed to not contribute to strengthening in the product.

3.4 Texture strengthening

Texture plays an important role in the ductility and forming properties of steel, but has only a minor role in determining the strength of the final product. Typically, the contribution of texture to strength is about 30 MPa [11]. The influence of texture on total elongation is assumed to be represented by the change in UTS that is predicted. The prediction of the other properties is unaffected by texture.

3.5 Precipitation strengthening

Precipitation strengthening refers to the increase in strength that is associated with the formation of small particles of a second phase in a matrix, where the elements that make up the particles were initially dissolved in the matrix. Precipitation is a minor factor for plain-C steels (AlN precipitation determines the free N content that contributes to solid solution strengthening), however, it is important in HSLA steels. When a steel is alloyed with Nb, Ti, and/or V, micro-alloy carbonitrides can precipitate in austenite and/or ferrite. Precipitation in austenite retards recrystallization and grain growth, and also increases the deformation resistance of the steel. With regards the mechanical properties at ambient temperatures, precipitation in austenite is generally assumed to limit the amount of alloying elements available to form precipitates in ferrite, which does increase the measured mechanical properties. The precipitation strengthening in ferrite is controlled by the volume fraction of the particles and the coarsening of precipitates; the volume fraction of particles can be estimated from the chemical composition, but information on the kinetics of carbonitride coarsening in ferrite is absent from the literature. A major part of our effort has been to characterize the kinetics for different precipitates that occur during and after transformation. These precipitates are mostly carbonitrides with niobium, niobium/titanium, or vanadium.

3.6 Solid solution strengthening

Solid solution strengthening has been studied extensively and it is widely accepted that the solid solution strengthening adds to the lattice friction in Eq. 1. Theory and experiment have shown that the increase in strength is proportional to the square root of the concentration of the solute atom. For dilute solutions like the low-C sheet steels, a linear relationship is generally assumed. The exact value of the strengthening coefficients varies from one study to the next. The values published for YS and UTS from three different sources are shown in Table 2. The differences are significant and again reflect the different data bases from which the equations were developed. The property predictions from the different sources are similar because the differences tend to cancel out for the compositions typically encountered. Problems will almost certainly arise if the equations are applied outside of the range of compositions in the data base for which each was developed.

Table 2. Element vectors for structure-property relationships (MPa/mass %) in commercial C-Mn and microalloyed steels.

Element	Hodgson & Gibbs [7] LYS	Hodgson & Gibbs [7] UTS	IRSID [3] LYS	IRSID [3] UTS	Pickering [2] LYS	Pickering [11] UTS
Mn	26.1	53.6	23	29	37	27.7
Si	60.2	99.7	53	79	83	83.2
P	759	651.9	700	700	680	---
Cu	212.9	---	---	---	38	---
Ni	---	472.6	---	---	33	---
N	3286.0	3339.4	5000	5369	5000	---

4.0 RESULTS

4.1 A36 (0.16C, 0.70Mn)

A summary of all the tensile tests is shown in Table 3. Both as-received coil and heat treated samples are included. The microstructure of the thinnest A36 coil is shown in Fig. 1. Grain size and ferrite fraction change with the processing conditions. Additional tests were performed to evaluate the influence of microstructure on the DBTT in this grade. Laboratory processing of A36 was a very valuable procedure to expand the range of microstructures to be evaluated beyond those supplied by a manufacturer.

Table 3. Summary of tensile tests on A36 coils.

coil #	gauge, mm	location	coil temp., °C	EQAD, μm	f_{α}	YS, MPa	UTS, MPa	elong., %
934848	9.4	MC	665	5.4	0.81	285	445	30
162331	6.1	MC	669	5.0	0.85	312	474	34.5
163634	9.1	MC	685	5.5 (6.5)	0.85	277	426	39.5
163715	4.8	MC	684	6.5 (5.5)	0.87	319	457	37
668290	4.3	HC	688	5.9		315	442	36
668290	4.3	TC	688	6.2		302	455	34

4.1.1 Through-thickness gradient in microstructure

The A36 grade presented some different problems because it has the highest C content of any of the grades in the program and because the samples received from the mill covered a wider range in thickness (approximately 4 to 10 mm). For the thickest coil, there was a gradient in the microstructure; the top surface exhibited nearly all non-polygonal ferrite grains and the center and bottom surface exhibited polygonal grains. Hardness profiles through the thickness of the two thickest coils are shown in Fig. 2, and quantitative metallography from the thickest coil is shown in Fig. 3. These variations are produced in the hot strip mill because the top surface of a thick plate does cool faster (a sheet of water tends to cover the top surface and promotes better heat transfer).

For the prediction of properties in the case where there is a gradient in micro-structure, the following guidelines have been adopted. The cross-section is divided into 100 evenly spacing points where there is a thermal-mechanical history to reference. Given the different histories, the microstructure is predicted from the transformation model. The cross section is then divided into quarters and the YS for the cross section is taken to be the lowest average YS for any quarter while the UTS for the cross section is simply the average of all 100 points.

4.1.2 Non-polygonal microstructures

A complication resulting from the microstructure gradient is that the coiling temperatures can be high enough to temper the non-polygonal structure, changing its strength. A modification to the equations for A36 has been developed to decrease the predicted mechanical properties of the non-polygonal microstructures found near the top surface of coils with a thickness greater

500 and 700°C. The ratio of tempered hardness to as-quenched hardness changes at nearly a constant rate for the temperature range of interest, that is, the temperature at which the strip is coiled.

Without a correction to the properties for coiling temperature, the model for A36 predicts strength properties that are higher than expected for those regions of the coil where non-polygonal microstructures are found. Given this limited set of data for 1016 that has nearly the same composition as A36, we suggest that the predicted properties be modified to reflect the influence of coiling temperature in a similar fashion. Therefore, whenever the local transformation temperature is lower than the coiling temperature, the model will need to multiple the current strength predictions by a factor Φ that varies from about 0.9 to 0.72, depending on the coiling temperature within the range of 500 to 700°C. Given the limited amount of data for which this modification is based on, a linear dependence of Φ on coiling temperatures between 500 and 700°C seems appropriate.

For example, the A36 coil C934848 had a measured coiling temperature of about 660°C and the model predicts non-polygonal microstructures for the top 3 mm of the thickness. If we consider this area of non-polygonal microstructure, the model without correcting for coiling temperature predicts the YS would be about 445 MPa and the UTS would be 635 MPa. Given the coiling temperature, our Φ would be 0.75. The predicted strengths corrected for coiling temperature would now be 333 and 476 MPa respectively. Now the predicted values agree quite well with the estimated values.

4.1.3 Toughness

For toughness prediction, the Pickering equation is taken from the literature [11], and we have tested two different conditions for A36 to verify the equation's prediction. Thermomechanical processing was performed on A36 crop samples, 30 mm thick, to provide samples for toughness evaluation. The blanks for processing were 80 mm long with a 13 x 30 mm cross section. The blanks were soaked at 1100°C for 10 minutes, furnace cooled to the deformation temperature, deformed, cooled rapidly (about 50°C / s at the surface) to about 700°C, and then air cooled. The deformation conditions and the resulting microstructural characterization are summarized in Table 4. Results from tensile testing are summarized in Table 5. Figure 5 shows the results of the Charpy V-notch tests. The relationship between the measured transition (27 J used as the measured transition) and predicted from Pickering's equation is shown in Fig. 6. The difference between measured and predicted values is about 5°C.

Table 4. Summary of processing and resulting microstructure for toughness samples (A36 crop).

condition	deformation temperature, °C	strain	ferrite grain size, mm	pearlite content, %
A	1000	0.90	0.0084	30
B	1050	0.26	0.0110	30

Table 5. Summary of tensile properties for the toughness samples (A36 crop).

condition	YS, MPa	UTS, MPa	uniform strain, %	reduction of area, %
A	317	467	22	67
B	270	450	17	68

The effect of the non-polygonal transformation product on toughness is not clear. Acicular ferrite and tempered bainitic microstructures can have relatively good toughness; however, the individual effects of grain refinement and alloy additions for non-polygonal microstructures have not been determined.

4.2 DQSK

A summary of all the tensile tests is shown in Table 6. Both as-received coil and heat treated samples are included. Figure 7 shows the microstructure that is typical of DQSK. The grains are nearly equiaxed and polygonal in nature. Grain size changes with processing conditions, but the ferrite fraction is relatively constant. Laboratory processing of DQSK was a very valuable procedure to expand the range of microstructures to be evaluated beyond those supplied by the manufacturer.

Table 6. Summary of tensile tests on DQSK.

Coil #	coil temp., °C	Measured YS, MPa	measured UTS, MPa	grain size, μm
659361	665	320	390	7.9
301978	550	310	375	9.4
301979	550	244	335	10.4
301980	550	245	331	10.3
659361	915*	259	325	13.1
659361	950*	254	318	15

* reheat temperature

The biggest unknown factor in the evaluation of the DQSK samples is the N content dissolved in the ferrite. N is an interstitial alloying element that has a large solid-solution strengthening effect (about 50 MPa / 100 ppm) when it is dissolved in the matrix. Under equilibrium conditions, there is more than enough Al to combine with all of the N; however, we know that with a low coiling temperature like 550°C, most of the N remains in ferrite and does not precipitate out as AlN.

To evaluate the free N content, we performed a series of strain aging tests where the samples were pulled in tension to a strain of 5% and then aged in boiling water for 4 hrs before resuming the tension test. Table 7 shows these results. The aging index was about 50 MPa for all of the as-received coil samples and was zero for the laboratory annealed samples.

Table 7. Summary of strain aging experiments on DQSK coils.

Coil/Location	Strength, MPa			Elongation, %			AI ^H , MPa
	YP	YS	UTS	YP	uniform	total	
659361/head, middle	368	330	N/A	2.8	N/A	N/A	46
659361 ^I /head, middle	346	321	N/A	4.9	N/A	N/A	0
301978/tail, left edge	282	269	N/A	2.6	N/A	N/A	46
301978/tail, center	322	277-321	375	2.7	18	29	54
301978 ^I /tail, center	280	267	N/A	5.0	N/A	N/A	0
301979/tail, left edge	253	253	339	2.3	23.8	35.3	52
301979/tail, center	266	252	332	3.2	25.8	37	52
301979 ^I /tail, center	256	253	N/A	5.5	N/A	N/A	0

^H aging index; strain 5% and then immerse in boiling water for 4 hrs.

^I annealed at 650°C for 1 hr before strain aging test.

An existing model for AlN formation found in the literature has been adopted in the coiling module to calculate free N [15]. The calculation assumes that Al diffusion in ferrite controls the precipitation. No validation of the model was done because the temperature predictions in the coiling model could not be verified.

4.3 HSLA V steel, 350 MPa grade

A summary of all the tensile tests is shown in Table 8. Both as-received coil and heat treated samples are included. The mechanical properties of the as-received samples show a 50-MPa variation in YS and 45-MPa variation in UTS. Figure 8 shows the microstructure that is typical of this grade. The grain shapes are somewhat irregular compared to those in DQSK, but no special significance could be associated with the irregular shapes. Grain size changes with heat treating conditions, but the ferrite fraction (not shown in the table) is relatively constant at about 0.95. Laboratory processing of the HSLA V crop samples was not a valuable procedure to expand the range of microstructures to be evaluated beyond those supplied by the manufacturer.

Table 8. Summary of all tensile tests on HSLA V grade 350 MPa steel.

Coil #	coil temp., °C	Measured YS, MPa	measured UTS, MPa	grain size, μm
480733-hr*		356	464	
480733-hc*		360	474	
480733-hl*		343	453	
480733-mr*		372	483	
480733-mc*	692	377	493	8.5
480733-tr*		360	474	
480733-tc*		355	468	
480733-tl*		350	465	
480733-tl*	692	327	448	8.5
570128-mc*	693	375	466	10.6
570129-mc*	693	379	475	10.6
608068-hc	650	367	445	7.7
608068-tc	660	360	446	5.1
NIST heat <u>treated</u> 480733-tl	600°C/1180 min.	430	504	
480733-tl	720°C/35 min.	368-401	470	
480733-tl	720°C/134 min.	348-379	440	
480733-tl	720°C/290 min.	326	401	
480733-tl	720°C/1200 min.	231-264	333-357	

* temper rolled

4.3.1 Temper rolling

One feature that is distinctly different for the V grade compared to the DQSK is the YPElong. The low values of YPElong in the V grade indicate that the samples had been temper rolled after coiling. Temper rolling can have some influence on the measured YS, usually decreasing the YS; temper rolling has essentially no effect on UTS so that the net effect is to increase the difference between YS and UTS. From Table 8, the temper rolled coil samples typically have a 100 to 115 MPa difference between YS and UTS while the as-rolled coils show an 80 to 85 MPa difference.

4.3.2 Within coil variations

Samples from one V coil were received in which nine separate areas in the coil could be studied and compared to each other to evaluate within coil variations. The properties varied with position in the coil while the ferrite grain size and ferrite fraction do not. The data in Table 9 show that there can be a 50-MPa variation in YS and UTS depending upon the location within the coil.

Table 9. Tensile test results from as-received V-HSLA coil #480733.

	Strength, MPa			Elongation, %		
	YP	LYS ^b	UTS	YP	uniform	total
head section						
right edge	357	357	468	0.5	15	27
	356	356	460	0.7	16	31
center	*	358	469	*	14	27
	*	367	479	*	16	28
left edge	353	348	458	1.4	17	29
	337	337	448	0.6	17	28
middle section						
right edge	*	374	485	*	14	25
	369	369	482	0.6	16	29
center	*	380	499	*	14	27
	*	375	486	*	13	25
left edge	362	362	476	0.7	15	27
	363	358	472	0.9	16	28
tail section						
right edge	*	362	471	*	16	28
	*	349	465	*	16	28
center	*	351	465	*	16	29
	*	350	465	*	15	29
left edge	*	331	451	*	16	31
	*	324	445	*	18	34

^b lower yield strength or offset yield strength, 0.2%; * continuous yielding, no yield point.

4.3.3 Influence of test orientation

The influence of test orientation on the hot-rolled coil samples was studied with the first V coil where samples from the head, middle, and tail were tested in longitudinal (0°), transverse (90°), and the diagonal (45°) orientations with respect to the rolling direction. Those results are summarized in Table 10 and they indicate that the 0° orientation has the lowest YS while the 90° samples had about 20 MPa higher strength; the samples from 45° were approximately the average of the other two.

Table 10. Summary of mechanical tests in V steel as a function of orientation.

	Strength, MPa			Elongation, %		
	YP	LYS	UTS	YP	uniform	total
V steel coil # 608068 - head						
long.	386	367	446	2.5	18.5	33
trans.	+25	+18	+5	+1.0	-0.5	-1
45°	+5	+13	+3	+0.4	-0.5	+2
V steel coil # 608068 - middle						
long.	377	370	455	1.5	16.5	31
trans.	+22	+16	-4	+1.0	NA	0
45°	+27	+27	+16	+1.0	+0.5	+3
V steel coil # 608068 - tail						
long.	367	360	446	1.6	19.5	33
trans.	+31	+22	+5	+1.5	+1.5	-2.5
45°	+16	+13	0	+0.4	-0.5	-0.5
V steel coil # 570128 - middle						
long.	377	375	466	1.4	16.5	28.5
trans.	+17	+15	+8	+0.1	-1.5	-1.5
V steel coil 570129 - middle						
long.	381	379	475	1.3	15	27.5
trans.	+14	+14	+1	+0.1	0	0

NA = not available

4.3.4 Aging of coil samples

A special set of tests was initiated to age the lowest strength (most under-aged) section of coil and determine the properties as a function of the aging conditions. Those results are summarized in Table 11 for peak-aged and over-aged conditions. With aging, the YPElong returns to the typical values and the overall strength level peaks after 1160 min at 600°C and then decreases with continued aging at 720°C. The grain size of the samples aged at 720°C for 1200 min was measured and found to be the same as the as-rolled samples; therefore, the increase in tensile properties is attributed to an increased precipitate volume fraction while the subsequent decrease is attributed to precipitate coarsening. At the peak strength, 600°C for 1160 min, the UTS is similar to the sample taken from the middle section, center of width, but the YS of the peak-aged sample is about 50 MPa higher; we attribute the lower YS of the as-received coil to the temper rolling.

Table 11. Tensile Properties of aged coil samples, V HSLA #480733, tail/left edge.

	Strength, MPa			Elongation, %		
	YP	LYS	UTS	YP	uniform	total
600° C / 1160 min -- peak-aged R _B 83						
1	431	431	509	3.6	13.6	27
2	428	428	498	4.4	16.8	29.5
720°C / 35 min -- overaged R _B 78.2						
1	368	368	468	6.6	17.8	34
2	401	401	473	3.8	14.2	26
720°C / 134 min -- over-aged R _B 74.5						
1	379	379	441	3.4	16	28
2	348	348	440	9.8	21	34
720°C / 295 min -- over-aged R _B 69-72						
1	333	333	407	3.1	15.4	27
2	322	319	395	1.1	16.1	29
720°C / 1200 min -- over-aged R _B 63-66						
1	264	264	357	4.4	26	39
2	231	231	333	1.4	24.1	37

4.4 HSLA Nb steel, 350 MPa grade

A summary of all the tensile tests on coil samples is shown in Table 12. Tensile tests from laboratory-treated samples are summarized in Table 13. Both as-received coil and heat treated samples are included. The mechanical properties of the as-received samples are remarkably consistent; there is only a 24-MPa variation in YS and 31-MPa variation in UTS. Laboratory processing of the HSLA Nb crop samples was not a valuable procedure to expand the range of microstructures to be evaluated beyond those supplied by the manufacturer. Figure 9 shows the microstructure that is typical of the HSLA Nb. Laboratory processing of the HSLA Nb crop samples was not a valuable procedure to expand the range of microstructures to be evaluated beyond those supplied by the manufacturer. Grain size changes with heat treating conditions, but the ferrite fraction (not shown in the table) is relatively constant at about 0.95. The grains are polygonal in nature, but not always equiaxed; the grain diameter is typically a factor of two bigger in the rolling direction. No correlation was found between the slight elongation of the grains and the mechanical properties that we are concerned with in this program.

Table 12. Summary for all tensile tests on HSLA Nb 350 MPa grade.

Coil #	coil temp., °C	Measured YS, MPa	Measured UTS, MPa	Nb content	grain size, μm
610918-hc*	662	461	529	0.041	4.2
610918-tc	650	460	528	0.041	4.1
565695-mc	666	472	541	0.035	4.7 (4.5)
544726-hr		480	558	0.043	
544726-hc		482	555	0.043	
544726-hl		481	558	0.043	
544726-mr		473	558	0.043	
544726-mc	577	474	551	0.043	3.6 (4.0)
544726-ml		477	552	0.043	
544726-tr		484	558	0.043	
544726-tc		470	549	0.043	
544726-tr		467	550	0.043	
underaged -tc		466	527	0.036	3.7

* the first letter is the location along the length of a coil, h=head, m=middle, and t=tail; the second letter is the location across the width, r=right, c=center, and l=left.

Table 13. Summary of all laboratory-treated HSLA Nb samples.

ID	coil temp., °C	Measured YS, MPa	measured UTS, MPa	Nb content	grain size, μm
544726	750°C/63 min	422	474	0.043	4.8
544726	750°C/76 min	405	466	0.043	4.9
544726	650°C/4290 min	465	516	0.043	4.9
544726	750°C/234 min	386	434	0.043	5.1
544726	750°C/234 min	343	417	0.043	5.4
544726	700°C/1400 min	439	472	0.043	4.9
544726	600°C/70 min	494	563	0.043	4.0
underaged-tc	600°C/100 min	494	552	0.036	3.7
underaged-tc	550°C/30 min	455	520	0.036	3.7
underaged-tc	600°C/210 min	490	552	0.036	3.7
M-ville rolling-1	as-received	391	471	0.041	5.4
M-ville rolling-2	as-received	433	*	0.041	8.9
M-ville rolling-2	30 min @ 600 C	421	506	0.041	8.9
M-ville rolling-2	100 min @ 600 C	438	523	0.041	8.9
M-ville rolling-2	30 min @ 525 C	473	540	0.041	8.9
M-ville rolling-3	as-received	391	454	0.041	7.3
M-ville rolling-4	as-received	401	468	0.041	7.5
M-ville rolling-5	as-received	442	501	0.041	6.8
M-ville rolling-5	30 min @ 600 C	445	514	0.041	6.8
M-ville rolling-5	100 min @ 600 C	448	519	0.041	6.8
M-ville rolling-5	30 min @ 525 C	455	520	0.041	6.8
M-ville rolling-6	as-received	399	469	0.041	8.1

4.4.1 Within coil variations

Samples from one USS Nb coil were received in which nine separate areas in the coil could be studied and compared to each other to evaluate within coil variations. The properties and microstructure varied little with position in the coil; data in Table 14 show that there is a 24-MPa variation in YS depending upon the location within the coil.

Table 14. Within coil variations in tensile properties, HSLA Nb, coil # 544726.

	Strength, MPa			Elongation, %		
	YP	LYS ^b	UTS	YP	uniform	total
Coil # 544726-- Nb head section						
right edge	488	476	557	NA	NA	23
	493	484	560	NA	NA	25
center	483	480	549	3.6	16	26
	501	484	561	3.5	15	26
left edge	490	481	558	3.2	16	24
	491	482	558	3.1	15	24
coil # 544726 -- Nb middle section						
right edge	494	473	558	3.2	16	26
	488	474	559	2.8	16	24
center-long.	486	474	549	3.6	16	26
	477	475	553	3.6	15	24
	522	517	577	4.3	16	26
center-trans.	516	511	571	3.9	15	24
left edge	489	472	549	3.5	16	24
	488	482	556	3.5	17	28
coil # 544726 -- Nb tail section						
right edge	493	488	557	3.3	16	25
	492	480	559	3.2	17	29
center	487	473	551	3.2	17	27
	482	468	547	3.4	14	22
left edge	475	469	550	2.8	16	24
	478	466	550	2.8	16	26

LYS^b = lower yield strength or offset yield strength, 0.2%

NA = not available

4.4.2 Aging of coil samples

Extra tests were initiated to determine the properties as a function of the aging conditions. Those results are summarized in Table 15 for under-, peak-, and over-aged conditions.

Table 15. Tensile test results from Nb steel after aging.

	grain size, μm	Strength, MPa			Elongation, %		
		YP	LYS	UTS	YP	uniform	total
0.036% Nb HSLA extreme tail coil sample, NIST heat treated							
as-received -- R _B 82, underaged	3.7	466 465 466 467	466 465 466 467	528 527 525 529	3.8 3.8 3.5 3.6	15.7 16.4 15.6 15.6	27.1 30.1 29.0 28.3
AR + 550°C, 30 min -- R _B 81.5, underaged	*	456 456	453 459	518 521	3.6 3.6	17.0 16.0	26.9 27.4
AR + 600°C, 100 min--R _B 87.5, peak aged	*	493 495	493 495	549 554	3.8 3.8	15.7 15.3	25.7 25.2
AR + 600°C, 210 min -- R _B 87.5, peak aged	*	481 491	487 494	549 556	3.8 3.9	15.6 13.5	27.7 24.7
0.043% Nb HSLA coil #544726, NIST heat treated							
as-received --R _B 87, underaged	4.0		465 to 485	550 to 560			22 to 29
600°C / 70 min -- R _B 87.4, peak aged	4.0	511 515	494 494	566 559	3.9 3.0	16.4 12.9	26.5 22.0
650°C / 4290 min -- R _B 84.0, overaged	4.9	498 489	463 468	516 516	5.4 5.5	15.4 16.7	25.5 26.9
700°C / 1400 min -- R _B 82.0, overaged	4.9	464 457	438 439	472 472	6.5 7.7	17.2 16.2	27.9 26.9
750°C / 63 min -- R _B 78.6, overaged	4.8	435 454	421 424	473 474	5.5 5.5	17.5 16.1	28.3 25.2
750°C / 76 min -- R _B 75.8, overaged	4.9	434 411	413 398	467 466	6.3 3.9	16.1 16.2	25.7 24.9
750°C / 234 min -- R _B 71, overaged	5.2	407 358	386 343	434 417	7.1 4.9	20.5 19.7	33.1 30.1

* assume to be the same as for the as-received sample.

4.5 HSLA Nb-Ti steel, 360 MPa grade

A summary of the tensile tests for this grade is shown in Table 16. Only as-received coil samples from LTV are included. The mechanical properties of the as-received samples from head and tail sections are remarkably consistent; there is 15 MPa variation in YS and 33 MPa variation in UTS. The microstructure is similar to the straight Nb grade.

Table 16. Summary of all tensile tests on HSLA Nb-Ti 360 MPa grade.

coil #	location	coil temp., °C	EQAD, μm	YS, MPa	UTS, MPa
1940401	HC	644	4.2	420	493
1940401	TC	644		422	494
1940702	HC	627	5.4	414	507
1940702	TC	627		417	502

The equations used to predict strength and tensile elongation (50 mm long gage section) for the HSLA Nb-Ti 360 MPa grade steel are summarized in Appendix E.

4.5.1 Within coil variations

We received head and tail sections from two different LTV coils (no middle section is included as was the case for the V and Nb HSLA steels); each section included a sample from the left edge, center, and right edge of the coil for a total of six different areas on each of two coils. The properties were extremely uniform, as mentioned earlier; the microstructure was not as uniform as that observed in the USS Nb grade, but the variations are considered to be of little consequence.

4.5.2 Aging of coil samples

Only hardness tests were done on the aged coil samples. The hardness of the starting block for the aging experiment varied from R_B 78 to 81; this variation probably reflects the variations observed in the grain size and is important only on a local scale as the tensile properties, which reflects better the bulk, were uniform. With aging at 650°C, the hardness did not increase above 81 and started to decrease after about 200 min at temperature.

4.6 HSLA Nb-Ti steel, 550 MPa grade

A summary of the tensile tests for this grade is shown in Table 17. As-received and heat treated coil samples are included. The typical microstructure for the HSLA Nb-Ti 550 MPa grade shown in Fig. 10, is finer than that of the straight Nb grade. The grain morphologies are not always polygonal or equiaxed, which makes the characterization of the structure difficult. Laboratory processing of the HSLA Nb-Ti crop samples was not a valuable procedure to expand the range of microstructures to be evaluated beyond those supplied by the manufacturer.

Table 17. Summary of tensile tests on Nb-Ti HSLA, 550 MPa grade

coil #	location	coil temp., °C	EQAD, μm	YS, MPa	UTS, MPa
Coil -- 1, 1995 coil					
841685	HC	648	2.9	626	703
841685	TC	642	2.9	595	675
Coil -- 2, 1996 coil					
845701	HC	644	3.1 (3.3)	650	734
845701	HL	644		633	703
845701	MC	644	3.4 (4.2)	648	720
845701	ML	644		645	722
845701	TC	644	3.4 (3.6)	640	730
845701	TL	644		635	724
Coil -- 3, 1997 coil					
as-rec.	TL	612	2.0 (2.9)	587	644
as-rec.	MC	612	2.6 (3.2)	590	655
aged	TL	525°C/30 min		605	658
aged	TL	650°C/10 min		608	658
aged	TL	650°C/90 min		632	671
Coil -- 2, 1996 coil -- NIST heat treated					
950°C/ 60 min		air cool from 950° C	5.5	372	459
950° C/ 60 min		air cool from 790° C	5.3	349	449
950° C/ 60 min		air cool from 650°C	6.8	328	431
950° C/ 10 min		1°C / min cooling		333	423

4.6.1 Grain size measurements

The second coil received in 1996 that included head, middle and tail sections was used for a interlaboratory comparison between UBC and NIST for grain size measurements. One half of the broken, longitudinal tensile sample from the three positions, center location was distributed to each lab. The undeformed grip area was sectioned, polished and etched according to the standard procedure in each lab, and examined in both the light microscope and scanning electron microscope (SEM). The measurements for the grain sizes in the two labs are included in Table 17; the measurements from UBC are listed first while those from NIST are in parentheses. For the head and tail center locations, there is excellent agreement; in the middle center location, NIST does measure a larger size than that measured at UBC (4.2 compared to 3.4 μm). This 0.8 μm difference in ferrite grain size amounts to a 33 MPa difference in the predicted LYS and a 27 MPa difference in UTS from the preferred equations developed here. The sensitivity of the strength prediction to the measurement of grain size is a concern for the Stelmax 80 grade where the typical grain size is not equiaxed or polygonal.

4.6.2 Within coil variations

We received head and tail sections from one coil in 1995; in 1996, we received head, middle, and tail pieces from a second coil; and in 1997, a middle section and an extreme tail section were sent. The tail section from the 1995 coil did have a lower strength than that of the head (about 30 MPa difference). The 1996 coil was checked in six locations, and the largest variations in strength are seen in the head section where the YS variation was 17 MPa while the UTS variation was 31 MPa. In the 1997 coil, the as-received tensile properties were similar in the extreme tail and in the middle sections.

4.6.3 Aging of coil samples

Both hardness and tensile tests were done on aged coil samples. The tail section from the 1995 coil was aged at 650°C for up to 1000 min and at 550°C for up to 10000 min and the results are shown in Fig. 11. At both temperatures, the peak hardness is the same and the time it takes to reach peak hardness is consistent with literature values for Nb steel [17].

For the 1996 coil, there was less than three Rockwell points (B scale) increase in the hardness with aging and since no significance could be attached to that, no further experiments were performed on this material.

For the 1997 coil, the middle section showed no increase in hardness with aging while the extreme tail section did exhibit some aging response in the hardness test; therefore, additional tensile samples from the extreme tail were machined and aged. The results of these tests are shown in Table 18. The LYS increased about 40 MPa while the UTS was 30 MPa higher after 90 min at 650°C. The relative change in strength is consistent with the increase in hardness observed in the 1995 coil after aging. These changes with aging reflect the relative importance of location within the coil on the strength.

Table 18. Tensile test results from aged samples of the third coil, extreme tail section.

	Strength, MPa			Elongation, %		
	YP	LYS	UTS	YP	uniform	total
as-received						
#1	583	583	650	2.6	10.6	19.4
#2	594	591	637	3.1	12.0	22.8
525°C for 30 min						
#1	608	606	666	2.6	10.4	19.4
#2	608	604	651	3.9	12.0	21.8
650°C for 10 min						
#1	642	610	659	3.4	11.1	20.0
#2	636	606	657	3.6	11.8	20.9
650°C for 90 min						
#1	650	630	669	4.1	10.8	19.0
#2	656	635	673	3.6	10.6	20.0

4.6.4 Effect of heat treatment

Additional tests were performed on the HSLA Nb-Ti 550 MPa grade to confirm the properties of this steel when it has been processed so that the grain size is coarser and the effects of precipitation are gone. The heat treatment took as-received sheets and reheated the material to 950°C for 1 hr and then cooled at different rates. The low, reheat temperature of 950°C insures that little Nb and Ti will be dissolved in the austenite and be available for subsequent precipitation in ferrite. The results are summarized in the overall table of properties, Table 17.

4.7 Nb-Ti and Nb-lean IF grades

The work on this grade for the structure-properties work was limited due to the fact that no IF steels are sold after hot-rolling. A summary of the tensile tests are shown in Table 19. Only as-received coil samples are included. The microstructures for the IF sheets are coarser than that of any of the other grades included in the program and the microstructures revealed some differences in grain size and shapes. The effect of grain shape is not considered to have a significant effect on the properties. A variation in the LYS for the tail sections indicates that there is some effect of processing, but the UTS shows little effect and we conclude that the influence of processing is small and can be neglected.

The equations used to predict strength and tensile elongation (50 mm long gage section) for the IF grade steels are the same as those for DQSK.

Table 19 Tensile test results from the IF steel coils.

	Strength, MPa			Elongation, %		
	YP	LYS ^b	UTS	YP	uniform	total
#146846 tail center	272	260	332	3.6	26.0	45
	276	259	330	3.3	24.6	43
#150158 head center	261	251	329	3.5	25.4	42
	260	249	327	3.3	25.2	43
	253	244	329	3.3	23.8	47
	255	246	330	2.2	23.0	45
#150158 tail center	296	296	334	3.3	21.9	42
	292	281	333	3.3	21.4	40
	283	265	333	3.6	22.4	41
	263	249	327	3.3	25.5	45
#150159 head center	258	255	330	3.0	21.7	38
	259	248	327	2.7	22.4	38
	258	253	332	2.4	21.4	40
	262	254	333	2.4	22.4	41
Coil #18810-01 longitudinal						
Head	237	231	310	3.3	23.6	41
	232	231	312	3.6	23.9	38
	230	226	312	3.3	22.5	44
	225	225	308	2.7	23.6	40
Tail	273	258	316	4.1	22.8	41
	291	280	308	3.5	20.8	41
	261	250	307	3.8	27.9	44
	262	255	317	4.6	25.0	45
Coil #18810-02 longitudinal						
Head	230	220	311	2.8	23.9	45
	227	217	301	2.4	24.3	45
Tail	264	256	316	3.3	23.0	45
	267	258	323	3.6	22.8	45
Coil #18810-02 transverse						
Head	276	256	303	5.4	24.3	45
	284	266	319	3.6	22.7	43
Tail	282	263	320	5.4	22.8	46
	295	271	321	5.7	23.0	45

LYS^b = lower yield strength or offset yield strength, 0.2%

4.8 Other Nb HSLA grades

Samples from these other grades have been collected from different companies to supplement the data base available for the development of property predictions. Three Companies have participated to some extent. The B-coils had been temper rolled while the others were as-rolled. The chemical compositions are all similar, but each has a different mill setup. The tensile test results are summarized in Table 20. The microstructure was checked and the grains are always polygonal and nearly equiaxed.

Table 20. Summary of tensile tests for HSLA Nb coils from other mills.

coil #	location	coil temp., °C	EQAD, μm	YS, MPa	UTS, MPa	Remarks
845758	TC	678	4.9 (5.3)	436	509	A
43654*	HC	675	4.4	422	533	B
43654*	TC	675	3.9	418	527	B
43654	TC	600 C/100 min	3.9	469	530	B
43622*	HC	675	4.2	408	505	B
43622*	TC	675	5.0	405	495	B
99459*	HC	550	4.8	427	525	B
99459*	TC	550	4.4	423	530	B
99460*	HC	550	4.7	426	527	B
99460*	TC	550	4.6	428	535	B
99460	TC	600 C/100 min	4.6	481	545	B
96466*	HC	600	5.9	405	512	B
96466*	TC	600	5.6	406	515	B
96466	TC	600 C/100 min	5.6	431	488	B
96467*	HC	600	6.0	405	510	B
96467*	TC	600	5.0	393	515	B
824877	MC	655	5.2	386	468	C

* temper rolled

4.8.1 Within coil variations

We received samples from nine different locations for the C sheet; head, middle, and tail sections were included that spanned the full width of the coil. The results of the tensile tests are shown in Table 21. The largest variation in the LYS was 21 MPa while the UTS variation was 19 MPa. The samples from the other steel companies did not allow for a significant comparison of within coil variations.

Table 21. Tensile test results from the C- Nb steel coil.

	Strength, MPa			Elongation, %		
	YP	LYS	UTS	YP	uniform	total
Head-left	399	393	474	2.0	19.5	31.2
-center	383	380	462	NA	NA	31.0
-right	380	375	460	1.9	18.9	30.7
Middle-left	404	401	474	3.3	19.2	30.4
-center	392	386	468	2.7	18.6	31.0
-right	398	395	467	3.1	20.3	31.0
Tail-left	400	387	455	4.1	19.7	30.4
-center	403	397	455	4.7	21.4	34.0
-right	398	392	456	3.3	19.2	31.0

NA = not available.

4.8.2 Aging of coil samples

Both hardness and tensile tests were done on aged coil samples from B. The as-received coils were aged at 600°C for 100 min. Each showed some increase in hardness with aging, between 2 and 3 points Rockwell B. To document the corresponding change in tensile properties, the aging was repeated on samples large enough for tensile testing from three of the coils with different thickness and coiling temperatures. The results are shown below in Table 22. In each case, there was an increase in the LYS with aging, from 26 to 54 MPa; the biggest increase was for the sheet coiled at 550°C. For the UTS, only the sheet coiled at 550°C exhibited an increase; the UTS actually went down by 26 MPa for the sheet coiled at 600°C. The fact that the as-received coils were temper rolled could account for at least a 20 to 30-MPa difference in the LYS and no change in the UTS. In conclusion, it appears that the sheets coiled at temperatures above 550°C were at their maximum strength possible given the ferrite grain size produced on the run-out table. For the sheet coiled at 550°C, the low coiling temperature resulted in a slightly lower strength than possible, about 15 to 25 MPa at most.

Table 22. Tensile test results from B- Nb coils.

	Strength, MPa			Elongation, %		
	YP	LYS	UTS	YP	uniform	total
coil #A43654 -- 1.9 mm thick, 675°C coiling						
as-received	420	420	530	1.0	16	24.5
600°C / 100 min	475	469	530	5.0	19.5	28.5
coil #T99460 -- 4.0 mm thick, 550°C coiling						
as-received	430	427	530	0.8	17.0	28
600°C / 100 min	498	481	545	3.4	16.0	27.5
coil #T96466 -- 6.2 mm thick, 600°C coiling						
as-received	405	405	514	0.8	17.5	31
600°C / 100 min	438	431	488	3.5	18.4	32.5

5.0 DISCUSSION

This section will focus on the two main contributions to strength, base strength (grain refinement and solid solution strengthening) and precipitation. The first sections will deal with the relevant literature and how those equations are applied to the hot-strip mill. The next two sections involve the fine tuning and validation of the equations for the program. The final sections discuss the limitations and possible further applications for the structure-property equations.

5.1 Equations to predict properties

5.1.1 Plain-C steels - base strength

The approach taken for the two plain-C steel grades considered in the program is that only grain refinement and solid solution strengthening contribute to strength in the as-rolled sheets.

5.1.1.1 Review of literature equations

There are five relevant sets of equations for strength found in the literature that could be applied to the hot-strip mill model. Shown below are equations from Pickering [11,2], Hodgson and Gibbs [7], IRSID [3], Tomota [8], and Campbell et al. [12].

Pickering:

$$YS = 88 + 37[Mn] + 83[Si] + 2918[N_f] + 15.1d^{-1/2} \quad (2)$$

$$UTS = 294 + 28[Mn] + 83[Si] + 3.85[f_p] + 7.7d^{-1/2} \quad (3)$$

Hodgson and Gibbs:

$$YS = 62.2 + 26.1[Mn] + 60.2[Si] + 3286[N_f] + 19.7d^{-1/2} \quad (4)$$

$$UTS = 164.9 + 53.6[Mn] + 99.7[Si] + 634.7[C] + 11d^{-1/2} \quad (5)$$

IRSID:

$$YS = 63 + YS_{ss} + k_y f_\alpha d_\alpha^{-1/2} + (360 + 2600 [C]^2) (1 - f_\alpha), \text{ and} \quad (6)$$

$$UTS = 237 + UTS_{ss} + k_t f_\alpha d^{-1/2} + 500 (1 - f_\alpha), \quad (7)$$

$$\text{where } YS_{ss} = 23 [Mn] + 53 [Si] + 700 [P] + 5000 [N_{free}], \quad (8)$$

$$k_y = 15.4 - 30 [C] + 6.094 / (0.8 + [Mn]), \quad (9)$$

f_α = ferrite fraction,

d_α = ferrite grain size (mm),

$$UTS_{ss} = 29 [Mn] + 79 [Si] + 700 [P] + 5369 [N_{free}], \quad (10)$$

$$k_t = 7.24. \quad (11)$$

Tomota et al.:

$$\sigma = a (b + \epsilon_p)^N, \quad (12)$$

$$\text{where } a = 373 + 62.7[Mn] + 109[Si] + 10.8d^{0.5}, \quad (13)$$

$$b = 0.002, \text{ and} \quad (14)$$

$$N = 5.35 / (10 + d^{0.5}) - 0.024 Mn - 0.037[Si] - 0.12 \quad (15)$$

so that,

$$YS = \sigma \text{ at } \epsilon_p = 0.002 \text{ and}$$

$$UTS = a (N)^N. \quad (16)$$

Campbell et al.:

$$YS = f_\alpha (132 + 11.8d_\alpha^{-1/2}) + (1 - f_\alpha)(408 + 92.2[Mn] + 0.400S_p^{-1/2}) + 79.7[Si] \quad (17)$$

$$UTS = f_\alpha (197 + 15.9d_\alpha^{-1/2}) + (1 - f_\alpha)(592 + 0.791S_p^{-1/2}) + 500[Si], \quad (18)$$

where YS and UTS are in MPa, d is in mm, elemental composition is in mass %, f_α is the % pearlite, and S_p = pearlite spacing.

Tomota's equations are best applied to low C steels with no appreciable free interstitial content; the YS cannot be considered for a typical hot-strip product where there is appreciable YPElong. The equations from Campbell et al. require additional information about the pearlite spacing that is not predicted from the transformation model. For comparison purposes, we assume the spacing to be 0.4 μ m; serious considerations of these equations would require expansion of the transformation model to include interstitial solute content and pearlite spacing. We include them here because they could be used for future applications.

A consideration in choosing the equations to predict strength is the grain size dependence. Plain C steels typically vary from about 6 to 20 μm in ferrite grain size. However, we will use these equations developed from plain-C steels for microalloyed steels also so that we can evaluate the contribution to strength from precipitation strengthening. The ferrite grain sizes in microalloyed steels are typically finer than that observed in plain-C steel; a ferrite grain size as small as 2.5 μm has been observed in a Nb-Ti steel [16]. Therefore, we must verify that the equations will work for the smaller grain sizes as well.

A different, but important issue relates to the way in which the second phase affects the strength. One approach for the prediction of YS, taken by Pickering and Hodgson and Gibbs, assumes that the properties of ferrite control the YS until the volume fraction of second phase reaches 20%, which is sufficient for nearly all of the conditions with which we are concerned. Campbell and IRSID take a different approach, assuming a simple rule-of-mixtures equation to predict the LYS, but over a wider range of f_α than that applicable in the previous approach.

For UTS, each of the four equations deals with variations in f_α in a different fashion. The IRSID equation assumes the simple rule-of-mixtures with a constant strength for the pearlite while the equation from Campbell assumes the rule-of-mixtures, but with a dependence on pearlite spacing. Pickering and Hodgson and Gibbs again limit the application of their equations to less than 20%; Pickering predicts a linear dependence of UTS on the volume fraction and not composition or spacing, while Hodgson and Gibbs assume the contribution of the second phase depends on the C content and not on the spacing or f_α .

5.1.1.2 Application to hot-strip model

For the hot-strip model, the IRSID equations have been adopted to describe the base LYS and UTS for several reasons. First, the equations were developed for a wider range of composition and f_α than we need and therefore should be adequate for all cases; second, there is no need to complicate the transformation model with the prediction of pearlite spacing; finally, the equations predict a reasonable behavior for grain sizes between 20 and 8 μm , see Fig. 12, which is typical for plain-C steels.

An alternate set of equations to use for the prediction of strength in plain-C steels would be the YS equation from Pickering and the UTS equation from Hodgson and Gibbs. A major reason to use Pickering's YS and Hodgson and Gibbs' UTS equations together is that they give a consistent value for the contribution of precipitation in HSLA steels.

5.1.2 HSLA steels - precipitation strengthening

Precipitation strengthening is the result of the interaction between dislocations and the carbonitrides containing V, Nb, and/or Ti that form on cooling HSLA steels. The enhanced strength is a strong function of composition (volume fraction) and the distribution of particle sizes and spacings. There have been numerous approaches taken to predict the strengthening; we will review many of these in the following subsection and then make recommendations on how to implement the prediction.

5.1.2.1 Review of literature equations

We have compiled a list of physically-based equations found in the literature [5,6,18-25] that relate to the strengthening increment due to precipitation and the proposed equations are shown in Table 23. In most cases, the strengthening is calculated from structural parameters measured on carbon extraction replicas like the particle size χ , average particle spacing λ , and/or the particle volume fraction f ; however, there is no consistency between the different equations. Some equations assume that only the average particle spacing is important. Only Melander [19] takes into account the size distribution of particles. In other approaches, the concentration of the metal species (either V or Nb) is taken as an estimate of the volume fraction f . The calculated f , based on the temperature history and chemical composition can be correlated to strength as a linear function as NIST has done for the case of continuous cooling or taken as a function of the square root of the volume fraction estimated by Zajac et al. [26]. All of these approaches have been shown in the literature to work for a given set of data. Two complications for the predictions that just depend on calculated f are first, that they cannot predict over-aging (the case where the volume fraction is maximized and the particle sizes are coarsening), and second, the strengthening from a given f of NbC is different from that of VN particles, see Fig. 13.

A different approach to predicting variations in the precipitation strengthening in aluminum alloys has been proposed by Shercliff and Ashby [27]. Theirs is a process model that takes into account the particle coarsening and changes to the solid solution strengthening. The time constants for each process involved are combined and the effective activation energy for the system is measured so that the kinetics of aging is characterized. This approach requires that a series of experiments be done to evaluate the kinetics, but once the activation energy for a particular system is measured, the aging kinetics for any time-temperature history can be predicted.

Table 23. Equations found in the literature to describe precipitation strengthening.

Source	relationship
Gladman [18]	$(5.9 \bullet f^{1/2} x^{-1}) \ln(x/b) \text{ (nm)}$
Melander [19]	$0.7 \left(\frac{Gb}{\lambda} \right) \ln(\lambda/b) \ln \left[\frac{2d_g \lambda}{(d_g + \lambda)b} \right] / \ln(\lambda/b)^{3/2}$
Martin [20]	$2Gb_{\lambda}^{-1}$
Gawne and Lewis [21]	$(8.9 \bullet f^2 x^{-1}) \ln(1630 x) \text{ (}\mu\text{m)}$
Parilak et al. [22]	$38 + 76.8 \bullet 10^{-8} \lambda^{-2} \text{ (nm)}$
Hodgson and Gibbs [7]	$2500 \bullet Nb_{\text{sol}}$
Mitchell et al. [5]	$(\kappa_1 N_{\text{eff}} - \kappa_2) V^{1/2}$
Österle [23]	$0.519 Gb \lambda^{-1} \ln(x/b)$
Kwon et al. [24]	$\kappa V_{\alpha} f^2 x^{-1}$
NIST * [25]	$373(1 - \exp(-6.33Nb_{\text{sol}}))^{0.426}; \kappa f c$

f = volume fraction of precipitates, subscript c for calculated otherwise measured from replica;

b = burger's vector, 0.25 nm;

λ = interparticle spacing;

χ = average particle diameter;

d_g = geometric mean-particle diameter from extraction replicas;

V_{α} = volume fraction of ferrite;

κ , κ_1 , and κ_2 = fitting parameters;

N_{eff} = effective N content; total N - (%Ti/3.42);

assumes Poisson's ratio = 0.3;

* process dependent.

5.1.2.2 Application to hot-strip model

For the hot-strip model, we use the Shercliff-Ashby model [27]; this approach eliminates the need to characterize in detail the particle size and spacing in HSLA steels. Such measurements are always time-consuming and subject to large experimental errors. The experiments required to implement the Shercliff-Ashby model are simpler to perform and are readily integrated into the hot-strip model. The main problem associated with the Shercliff-Ashby model is obtaining suitably underaged material on which to perform the required experiments. For the HSLA V grade, the head and tail sections from the as-received coils were acceptable. Aging experiments at different temperatures were performed and the hardness from the samples is shown in Fig. 14. The time it takes to reach the peak hardness can be plotted against the aging temperature to calculate the effective activation energy Q .

For the remaining HSLA steels, Nb-containing grades, there is a fundamental problem to obtain material that is underaged. The as-received coil samples were within 3 Rockwell B points of the peak hardness that could be measured after aging experiments. Laboratory testing of crop samples was performed in which the samples were reheated to 1200°C, furnace cooled to 950°C, thermomechanically processed to refine the austenite grain size, and then water cooled. The resulting microstructures contained at least 50% non-polygonal constituents and only a 3-point Rockwell B increase was found after the aging the samples to the peak hardness.

A paper by Vollrath et al. [17] contained aging curves performed on a 0.09 % C, 0.72 % Mn sheet steel with 0.046 % Nb which had been coiled at low temperatures to suppress precipitation during coiling (490 to 540°C). The aging kinetics in Vollrath's work were similar to those found in our aging experiments on as-received coil samples and the thermomechanically processed crop samples of Nb steel. Figure 15 shows a comparison between the aging curves for one of the thermomechanically processed crop samples and that reported by Vollrath [17]. The absolute strengths are quite different, but the kinetics of precipitation appears to be identical. Therefore, the value of Q for the Nb-containing HSLA grades is assumed to be the same as that determined from the literature data [17].

5.2 Fine tuning of equations

Fine tuning of the structure-property equations goes along with validation of the model and is done in this program so that the accuracy of the model predictions can be maximized. This can only be accomplished when there is sufficient mill data along with metallographic samples to verify the structure.

The best example of fine tuning is probably the modifications made to the equations for A36 where there is a non-polygonal microstructure. The results from the mill on the thick samples showed us that the prediction of UTS was too high compared to the values we measured. The literature data from the ASM Handbook of continuous-cooling diagrams contained just enough data to make a correction for coiling possible.

A second example of fine tuning is related to the total elongation in HSLA steels. Given the numerous tests (approximately 70) we now have done on the HSLA Nb steel, we can make a better prediction of the total elongation in the tensile test by simply fitting the data we have rather than basing the prediction on results from a wide variety of low C steels. Figure 16 shows the measured total elongation plotted as a function of the measured UTS. The two lines represent the predicted elongation for all steel grades and a revised prediction that better fits the data. The suggested refinement to the prediction reduces the error between prediction and measurement over the range of strengths observed in our tests. In either case, there is a large scatter in the measured elongation for coil samples with UTS between 520 and 560 MPa that no simple refinement in the prediction will resolve. We expect that the morphology of the second phase plays an important role that is not included in the transformation model.

A third area for fine tuning is the prediction of UTS for the HSLA Nb-Ti 550 MPa steel, which has the finest grain size. The data from transformation experiments at UBC indicate that there is a change in the grain-size dependence of hardness (estimated UTS) once the grain size gets below 3.5 μm , see Fig. 17. The change in the grain-size dependence for the samples reheated to 1150°C could be related to the influence of a dislocation substructure, which we have no way to characterize at this point. Therefore, a factor is introduced that will add to the predicted UTS in a linear fashion for grain sizes below 3.5 μm ; the difference between the IRSID and modified predictions is zero at 3.5 μm and 50 MPa for a 2 μm grain size.

5.3 Validation of equations

5.3.1 Plain-C steels

For A36 and DQSK, there is a total of 14 cases where we have mill coil samples or heat treated samples that have measured properties and microstructure (α grain size and f_α). A comparison of measured vs. predicted properties is shown in Fig. 18 for the preferred (IRSID) and our alternative choice for equations (Pickering's LYS and Hodgson and Gibbs' UTS). The preferred equations from IRSID show better agreement with the measured values than the predictions from the alternative equations; the average difference between predicted and measured is ± 20 MPa for the preferred equations while the average is ± 38 MPa for the alternative.

An additional data set supplied by National Steel [28] shows quite similar results, Fig. 19. The preferred equations from IRSID show better agreement with the measured values (± 15 MPa) than the predictions from the alternative equations (± 35 MPa). Therefore, the IRSID equations are preferred to predict the strength properties of plain-C steels.

5.3.2 HSLA steels

The prediction of strength properties in HSLA steels is based on the IRSID equations with the additional factor of precipitation. The precipitation increment is a function of chemistry and thermomechanical history. Under certain conditions for the Nb-containing HSLA steels, strain-induced precipitation can occur in the austenite, which would reduce the Nb content available to precipitate in ferrite where the strengthening occurs. Likewise, the coiling conditions and sample

location within the coil can have an influence on precipitation. Both of these factors could lead to a lower precipitation strengthening contribution. Temper rolling will reduce the base strength for YS, but not for UTS. All of these factors lead to greater uncertainty in the prediction.

5.3.2.1 HSLA V grade

The comparison between prediction and measured strength properties for the four coil samples available (the measured LYS of the 3 temper rolled coils has been increased by 50 MPa to account for temper rolling) is shown in Fig. 20. For the preferred base strength equation, the prediction is always greater than the measured values. For the alternate equation, there is slightly better agreement between measured and predicted, particularly for UTS where there is not correction for temper rolling.

5.3.2.2 HSLA Nb-containing grades (no excess Ti)

The comparison between prediction and measured strength properties for the five coil samples available is shown in Fig. 21. The agreement between measured and predicted is excellent for the preferred equations (average difference between predicted and measured is 14 MPa) and slightly worse for the alternate case (average difference between predicted and measured is 20 MPa).

5.3.2.3 HSLA Nb-Ti grade (excess Ti)

The comparison between prediction and measured strength properties for the three coil samples available is shown in Fig. 22. The preferred predictions are comparable to the measured properties for all three coils; the alternate equations result in good predictions for two of the three coils.

For the HSLA Nb-Ti 550 MPa grade, we also have an extensive set of mill data (190 coils with thickness that range from 2.5 to 5 mm) that includes processing temperatures, rolling schedule, roll forces, chemical composition, and final properties (LYS, UTS, and total elongation). With no measured ferrite grain size, we must make some assumptions regarding the microstructure, but some trends are apparent. There is no clear correlation between mechanical properties and coiling temperature, see Fig. 23. There could be a couple of reasons for this; first, the sample for mechanical properties is usually taken at the tail end of the coil where the strength can be reduced by about 30 MPa, depending on the exact position; second, the variation in the measurement can be ± 10 MPa; and finally, there are chemistry variations within this grade. The chemistry variations are the only ones that we can examine in detail.

Figure 24 shows that the YS increases with higher excess Ti contents (similar to trend for UTS). The overall trend in the data has a slightly higher slope than that expected from the literature. The effect of variations in the Nb content within this grade can be considered only if the effect of excess Ti is first subtracted from the measured properties. Figure 25 shows the corrected YS vs. Nb content. The variation of Nb from 0.065 to 0.09 translates into an apparent increase of about 25 MPa, which compares to only a 4-MPa increase predicted from the preferred equation for the maximum precipitation and 24 MPa for the alternate equation.

After accounting for the composition factors, the scatter in data can still be as large as 100 MPa. The variation due to grain size is the other significant factor that needs to be accounted for in the prediction. Unfortunately, physical samples do not exist that would allow for validation of the grain size in such detail. For the limited coil samples that we received and from the testing of crop samples at UBC, we think that the range of grain sizes that will be produced under typical mill conditions is between 3.5 and 2 μm . The strengthening contribution from grain refinement from 3.5 to 2 μm predicted from the preferred equations is about 75 MPa where the alternate equations predict about 80 MPa; both values seem reasonable in light of the measured scatter in data.

In this way, the overall variations in strength can be explained on the basis of chemistry variations (Ti, N, and Nb) within the grade while the scatter for a given composition is largely influenced by the ferrite grain size (transformation characteristics). This prediction scheme appears to be consistent with all of the specific data available in the program.

5.3.3 Validation from the steel companies

The most thorough validation work is done by steel company B. In this work, the four grades with the closest chemistry to those in the program, A36, DQSK, HSLA-Nb, and HSLA-Nb/Ti 550 MPa grade, were studied. In this initial evaluation, the complete program was run on the four grades for various strip thicknesses and finishing and coiling temperatures. A brief summary of the evaluation for the HSLA-Nb 350 MPa grade is shown in Table 24. The predicted temperatures at an intermediate location (50 m from the last rolling stand) on the runout table were generally within ± 50 °C of the measured values. The predicted temperatures at the end of the runout table where the coil enters the downcoiler were often 50 to 100°C lower than the measured values. The YS and UTS predicted from the model were compared to the average values from the grades obtained using similar processing. The agreement between predicted and measured values was typically ± 35 MPa with the predictions equally scattered on either side of the measured values. In general, the model was deemed valuable to test modifications of the current processing procedures.

From our perspective at NIST, the evaluation of the structure-property equations can be revisited in three areas. First, the actual chemistries for the different grades can be input manually to accurately reflect the solid solution strengthening in the base strength and the maximum precipitation strengthening. The second area to be refined in the evaluation concerns the prediction of ferrite grain size. The predicted values need to be compared to the measured values when possible. The final point that needs to be analyzed to complete the evaluation of the structure-property equations is the effect of coiling temperature. In the plain C steels, the coiling temperature effects the contribution from dissolved N on solid solution strengthening; for the microalloyed steels, the coiling conditions will determine what percentage of the precipitation strengthening is obtained.

Figure 26 shows the influence of Nb content on the predicted precipitation strengthening. For the Nb grade 50, steel companies may use anywhere from about 0.015

Table 24. Summary of HSMM predictions for HSLA Nb 350 MPa grade.

Coil ID	thickness	α grain size	α fraction	coiling temperature/ accuracy*	$\Delta\sigma_{pptn}$, MPa	UTS accuracy*
E87559	2.36 mm	4.2 μm		661°C/-53°C	128	
E28793	6.49	4.6	0.93	636 / -32	144	+46 MPa
E36066	4.87	4.3	0.92	657 / +8	130	
E57506	3.13	3.6		598 / +87	153	
J23282	12.95	10.1	0.98	755 / -74	56	
J30386	1.92	5.1	0.90	727 / -47	74	
E61412	4.28	3.9		606 / +48	153	+41
E87334	9.67	6.6	0.94	641 / -2	141	+36
J24291	3.10	3.4	0.94	567 / +87	143	+30
J31903	2.13	4.6	0.93	694 / -38	100	-3
E79057	4.77	4.4		443 / +125	50	-58
E83630	5.31	4.1	0.94	430 / +139	47	-56
J24762	12.77	7.7	0.94	567 / -11	143	+31

* accuracy is defined as the measured value - HSMM predicted value.

to 0.04% Nb to make the required properties, which translates into a variation in precipitation of 50 (alternate) or 60 MPa (preferred) for the precipitation increment. For the evaluation referenced in 5.3.3, the model uses 0.036% Nb while the sheets had only 0.023% Nb. In other words, if the model used the actual Nb content of the B sheets instead of 0.036% Nb (used in the model) then the predicted strengths would be about 25 MPa lower if the maximum precipitation is predicted from the coiling model.

Figure 27 plots the measured ferrite grain sizes for samples of this grade received from a steel company as a function of thickness along with the HSMM predictions. The interesting feature here is that the model predicts a minimum ferrite grain size for the sheet at a thickness of about 3 mm. The measured grain sizes seem to indicate that there is a gradual grain refinement with decreasing thickness down to the 1.9 mm thick sample we examined.

Figure 28 shows the model predictions for free N in plain C steels and precipitation strengthening in HSLA steels as a function of coiling temperature. The interesting features here are the distinctive shape for each curve; in the plain C steels, the free N predicted from the model (Fig. 28a) is a strong function of coiling temperature between about 550 and 660°C (the model predictions are considered valid for the majority of the length of the coil, but may not for the head and tail as well as for the extreme edges). For temperatures above and below this range, the prediction is not a function of CT. For example, if the actual CT were 600°C and the model predicted a CT of 650°C, the model performance would be considered good for the temperature aspect; however, the influence of the inaccuracy of the temperature prediction would be that the predicted free N would be about 10% where the actual free N would be about 65%. The 55% difference in free N translates into the property predictions as a difference in the predicted properties of about 22 MPa for a steel with a total of 80 ppm of N.

A similar exercise can be performed for the precipitation strengthening contribution to the strength in HSLA steels containing Nb. From Fig. 28b, the precipitation peaks at 600°C and drops off to low values on either side of the peak. If the actual CT was 650°C, and if the model predicted a CT of 700°C, the model performance would be considered good for the temperature aspect; however, the influence of the inaccuracy of the temperature prediction would be that the predicted precipitation would be 42% where if the temperature prediction for CT were perfect, then the precipitation would be 69%. The 27% difference would translate into the property predictions as a difference in the predicted properties of about 41 MPa (preferred equation) or 30 MPa (alternate equation) for a steel with a total of 0.036 % Nb.

5.4 Additional applications

The proposal anticipated that the model could be applied to a variety of applications other than the hot-strip mill. The best example of this may be the prediction of properties for steel products made on a Steckel mill where the chemical compositions for the steel and processing conditions, temperature variations and interpass times, are quite different from that of the hot-strip mill. The preferred equations are used in the program, but are not validated for Steckel mill products.

For the plain C steels in the Steckel mill, little will change in the structure-property equations as long as the matrix structure is predominantly polygonal ferrite. However, in the microalloyed grades, often the compositions are designed to produce additional strength from transformation to acicular microstructures. Also, the maximum precipitation strengthening will depend on the soluble microalloy content in austenite prior to the phase transformation, which will vary along the length of the strip from zero to the total microalloy content of the steel. The Shercliff-Ashby process model can be used precisely the same as for the hot-strip model except the maximum strength will have to be modified to reflect the soluble microalloy content.

6.0 LIMITATIONS

The major limitation in the preferred structure-property equations for general applications appears to be the composition of the steel to be studied. The program has been setup to look at specific steel compositions and the equations work in the range of microstructures typically produced. However, for different combinations of composition and grain size, the equations may give unrealistic predictions. An example can be shown with the HSLA Nb composition; Fig. 29 shows the prediction of strength from the preferred equations for grain sizes between 2 and 14 μm . For grain sizes less than 3 μm , the predicted YS is greater than the predicted UTS. With the current processing capability of a hot strip mill, the grain size will probably never be less than 3.5 μm and the predictions for base strength are OK. But with improvements in the equipment and better control of the process, it is possible in the future that finer grain sizes will be achieved. In general, it appears that the preferred equation for YS represents a source of potential problems because the grain size dependence is a function of both C and Mn content while the grain size dependence of the UTS equation is not. Figure 30 shows predictions from the alternate set of equations for the HSLA Nb as a function of grain size; now the predictions for the base strength are in agreement with the properties typically found.

7.0 SUMMARY and RECOMMENDATIONS

Equations have been developed and validated for plain C and HSLA steels produced on the hot-strip mill. The lower yield strength, ultimate strength, and % total elongation are the main outputs from the equations. For thicker, plate-type products, the ductile-to-brittle transition temperature can also be predicted. The accuracy of the HSMM predictions for YS and UTS are about ± 35 MPa, which is not as good as the accuracy reported for a similar models [29], but the HSMM allows for more flexibility and detail in the setup for the hot-strip mill.

The equations developed for the structure-property relationships are applicable to a wider range in chemistry than just the eight grades in the program, but the alternate equations may give better predictions than the preferred equations in some cases. Base strength equations are considered applicable to grades where the microstructure is predominantly polygonal ferrite; prediction of precipitation strengthening is applicable to steels with Nb contents from 0.02 to 0.09, 0.08 V, and excess Ti levels up to 0.025.

Two areas of additional work appear to be needed to improve the applicability of the model. The first is related to the base strength; a systematic approach to correlate base strength to retained austenite and acicular constituents in the microstructure is needed. The second area is related to the Shercliff-Ashby model and its application in the coiling module. We need more data on precipitate coarsening (activation energy as a function of microalloy content) and we need measurements of the maximum precipitation strengthening that can be achieved for a given microalloy content. With the additional work, it should be possible to extend the model to predict the properties for all low C steels.

Any future work in this area should consider two additional experimental techniques. First, the use of hardness as a measure of strength could be utilized more extensively than we have in the program. The correlation between hardness and tensile properties is well known and reasonably accurate. In this program, equations that relate Rockwell B hardness R_B to tensile properties have been developed from data supplied by USS for steels with less than 0.12% C and are shown below:

$$YS = 386 - 3.23R_B + 0.104R_B^2$$

$$UTS = 586 - 11.3R_B + 0.1255R_B^2 .$$

The equations were used to predict the strength of hot-rolled coils and then compared to the measured properties; the results are shown in Fig. 31. Over the full range of strength properties, there is a good agreement between the predicted and measured properties.

The other area for refinement in any future test program is related to microstructure. Grain size is a critical factor in the model's predictions of properties. An ultrasonic measurement that could be correlated to grain size would be of great value in the work. Much of our effort has been in metallographic examination of ferrite grain size and there remain differences in the measurements made at NIST and those at UBC, particularly for the HSLA Nb-Ti 550 MPa grade that could account for 30 MPa in the prediction. An ultrasonic measurement should provide more consistent values within a given grade. One of the AISI/DOE Process Control Projects, Online Mechanical Properties Measurement, illustrates clearly what an ultrasonic measurement can contribute in the way of characterizing the microstructure.

8.0 SPECIAL NOTE:

In this report, the actual equations and the coefficients, both for the preferred and the alternate case, have not been published, as these are considered proprietary data of AISI.

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Fig. 1. Typical microstructure for A 36. Bar in picture represents 10 μ m.

Fig. 2. Hardness profiles through the two thickest A36 coils.

Fig. 3. Quantitative metallography from the thickest A36 coil as a function of location through the thickness of the sheet.

Fig. 4. Ratio of tempered hardness to as-quenched hardness for 1016 steel as a function of tempering temperature.

Fig. 5. Charpy V-notch test results from A36 after TMP.

Fig. 6. Comparison between equation predictions and measured values for the DBTT in A36 after TMP.

Fig. 7. Typical microstructure for DQSK. Bar in picture represents 10 μ m.

Fig. 8. Typical microstructure for HSLA V Grade 350 Mpa. Bar in picture represents 10 μ m.

Fig. 9. Typical microstructure for HSLA Nb Grade 350 Mpa. Bar in picture represents 10μm.

Fig. 10. Typical microstructure for the HSLA Nb-Ti grade 550 MPa. Sample was prepared with a color etch. Bar in picture represents 20 μm .

Fig. 11. Hardness as a function of aging time for the Nb-Ti HSLA 550 MPa grade steel.

Fig. 12. Predictions for DQSK YS and UTS from preferred set of equations as a function of grain size.

Fig. 13. Calculated precipitation strengthening as a function of estimated volume fraction of particles for NbC [16] and VN [27].

Fig. 14. Aging curves for HSLA V 50 grade steel as a function of time and temperature.

Fig. 15. Aging curves for HSLA Nb steel as a function of time after TMP at NIST and data from literature [17].

Fig. 16. Measured elongation in tensile test as a function of UTS for HSLA Nb steel.

Fig. 17. Estimated UTS from transformation experiments conducted at UBC on Nb-Ti 550 MPa grade.

Fig. 18. Comparison between A36 and DQSK measured and predicted properties from both the preferred and alternate equations.

Fig. 19. Comparison between measured data from one manufacturer [28] and predicted properties from both the preferred and alternate equations.

Fig. 20. Comparison between HSLA V 50 grade measured and predicted properties from both the preferred and alternate equations.

Fig. 21. Comparison between HSLA Nb, no excess Ti, measured and predicted properties from both the preferred and alternate equations.

Fig. 22. Comparison between HSLA Nb-Ti measured and predicted properties from both the preferred and alternate equations.

Fig. 23. Measured YS and UTS as a function of coiling temperature for the HSLA Nb-Ti 550 MPa grade.

Fig. 24. Measured YS as a function of excess Ti in the HSLA Nb-Ti 550 MPa grade.

Fig. 25.

Corrected YS as a function of Nb content.

Fig. 26.

Maximum precipitation strengthening as a function of Nb content.

Fig. 27. (a) Measured and predicted ferrite grain size and (b) predicted ferrite fractions for HSLA Nb 350 grade vs. thickness.

Fig. 28. Microstructure predictions from HSMM: (a) fraction of free N vs. coiling temperature and (b) fraction of maximum precipitation strengthening vs. coiling temperature.

Fig. 29. Predictions for HSLA Nb YS and UTS from preferred set of equations as a function of grain size.

Fig. 30. Predictions for HSLA Nb YS and UTS from alternate set of equations as a function of grain size.

Fig. 31. Comparison of measured to predicted tensile properties as a function of hardness for DQSK, HSLA 350 grade, and HSLA 550 grade sheet steels for automotive applications.