Recrystallization Controlled Rolling of Steels

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Keywords

High strength low alloy steels, microalloying, hot working, thermomechanical processing, microstructure, recrystallization, recovery, precipitation, flow stress, computer models, mechanical properties

Abstract

The aims of recrystallization controlled rolling especially followed by accelerated cooling (RCR+ACC) are to produce hot rolled steel products having high strength, high toughness and good weldability. The concept of RCR+ACC is attractive in that it is a relatively uncomplicated and high productivity process and can be applied on conventional mills.

The means of RCR is to make use of the grain refinement by repeated recrystallization of austenite with an intelligent design of deformation/temperature schedules with the help of computer models to calculate microstructure evolution and, above all, to retain the fine austenite microstructure during inter-pass delays and during cooling down to Ar_3 by a suitable dispersion of second phase particles. The fine austenite structure together with suitably fast cooling leads subsequently to a minimization of the ferrite grain size after transformation. Microalloying with Ti and V together with nitrogen can help to produce HSLA products with very favorable combinations of yield strength and toughness.

The effects of the RCR+ACC process parameters are discussed in relation to the microstructure development, final ferrite grain size and mechanical properties of Ti-V-(Nb)-N steels. A comparison of the results of full scale industrial plate processing using RCR practice with and without application of ACC with the results obtained at low finish rolling temperature CR practice is also made.

1. Introduction

Recrystallization controlled rolling (RCR) followed by accelerated cooling (ACC) has been developed during the last decade as one of the thermomechanical controlled processes for obtaining a good combination of strength, fracture toughness and weldability of HSLA steels in the as hot rolled condition [1-9]. The improvement in properties of RCR-ACC-plates and long products is associated with different strengthening mechanisms of which the most important is grain refinement whereby both strength and toughness are improved at the same time [8, 10].

In order to obtain optimum ferrite refinement, it is necessary to maximize the area of austenite grain boundary per unit volume at the on-set of phase transformation [11, 12]; this may be achieved either by low temperature controlled rolling or by recrystallization controlled rolling when the finish rolling temperature is relatively high, see Figure 1.

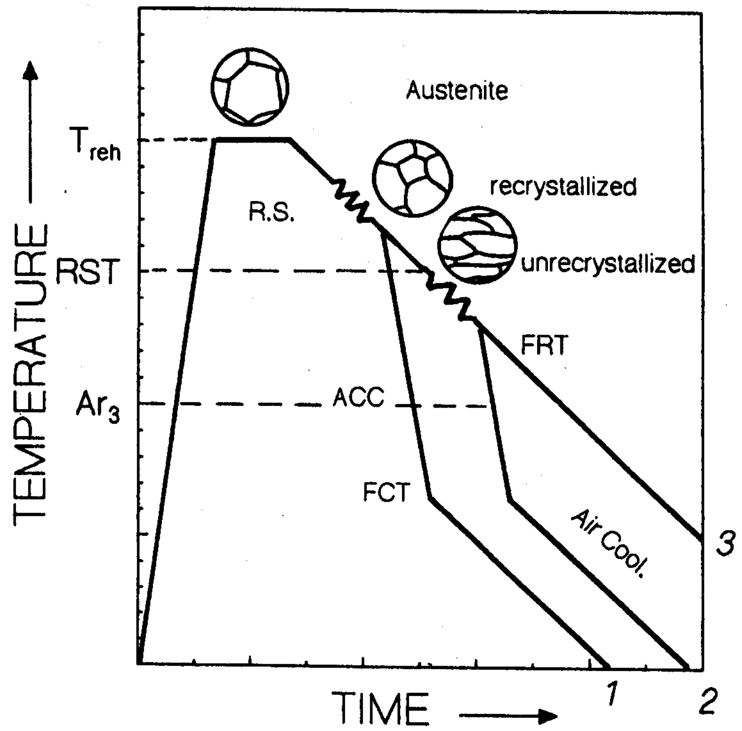

Figure 1. Thermomechanical Controlled Processes.
1. Recrystallization controlled rolling and accelerated cooling,
2. Controlled rolling and accelerated cooling,
3. Recrystallization controlled rolling or controlled rolling with air cooling.

The philosophy of RCR is to make use of the grain refinement occurring from static recrystallization of austenite at higher rolling temperatures, above the recrystallization stop temperature (see Figure 1) and, above all, to retain the fine microstructure during inter-pass delays and during cooling down to Ar_3 by a suitable dispersion of second phase particles.

The most effective way of maintaining a fine initial as-reheated grain size and at the same time controlling the austenite grain size during hot rolling is provided by a dispersion of fine titanium nitrides [5, 8, 13, 14]. An understanding of the role of different microalloying elements can be gained from the solubility product data summarized in Figure 2 from a recent thermochemical evaluation [15]. It is seen that TiN is extremely stable and can withstand dissolution at high reheating temperatures prior to hot rolling. Coarsening of TiN particles by Ostwald ripening is, however, accelerated by an excess of dissolved titanium so to maintain a small particle size it is necessary to make the steel composition under stoichiometric with respect to Ti. When the steel composition is optimised for grain size control by TiN, it is obvious that there will be no Ti available to provide precipitation strengthening. Hence, in practice, to design high strength ferrite-pearlite steel and to capitalize on any available excess nitrogen, it is desirable that Ti be complemented with an additional microalloying element. Both Nb and V can complement Ti as microalloying elements in principle. As seen in Figure 2, however, niobium carbide or carbonitride has a relatively low solubility and may precipitate in the later stages of rolling. This has two disadvantages since strain induced precipitation raises the recrystallization stop temperature and hinders the process of grain refinement by RCR while at the same time reducing the Nb available for precipitation strengthening in ferrite. Niobium also dissolves readily in TiN and reduces the stability of that phase at high temperature [16].

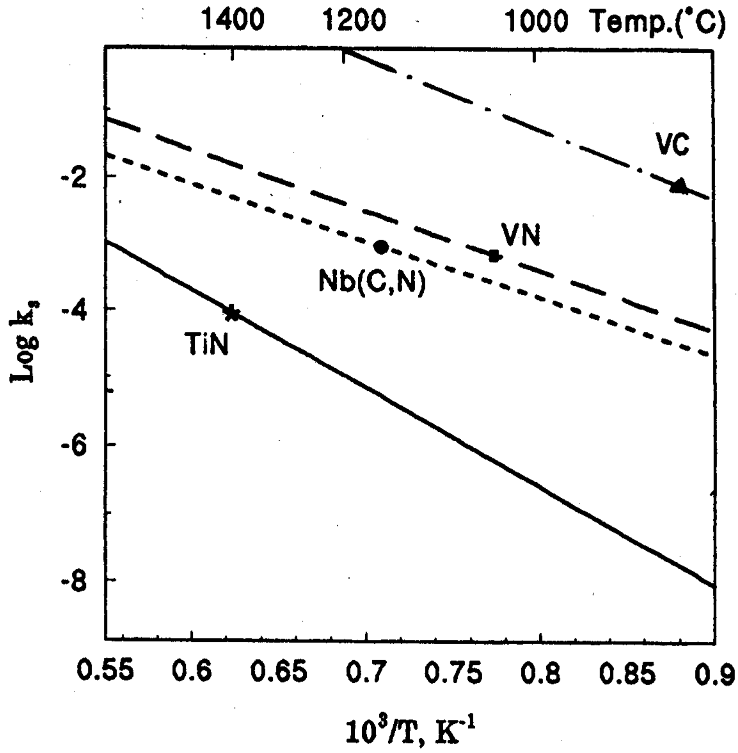

Figure 2. Solubility products of selected carbides and nitrides in the austenite temperature range. Points indicate typical steel compositions.

Vanadium, on the other hand, has little solubility in TiN so that grain boundary pinning can be maintained, but has a rather high solubility in austenite (Figure 2) even at temperatures as low as 1050°C [17]. This is important for the purpose of maximizing the precipitation strengthening, as the microalloying element must be completely dissolved at the reheating temperature and remain in solution until it precipitates in ferrite in the form of finely dispersed particles. Vanadium in solid solution has little effect on the kinetics of recrystallization of austenite and so the RCR process is not adversely affected.

When considering the role of nitrogen in these HSLA steels it may be noted that nitrogen exerts a detrimental effect in the presence of niobium, since niobium carbonitrides are less soluble than carbides and tend to precipitate excessively in austenite [18]. Recrystallization is inhibited and more of the potential precipitation hardening effect in ferrite is lost. In the presence of vanadium, however, excess nitrogen, which is not completely precipitated in austenite, causes not only grain refinement, but also leads to extra strengthening as a result of inter-phase precipitation of vanadium nitride particles. In addition to producing precipitation strengthening, the formation of VN particles is expected to reduce or eliminate any possible embrittling effect associated with 'free' nitrogen in solid solution [19] after accelerated cooling or welding.

Once the fine austenite grain size has been obtained after RCR, it should be preserved during cooling and then transformed to the finest possible ferrite grain structure. There are several ways to increase the transformation ratio, i.e., the ratio between the original austenite grain size and the resulting ferritic grain size. The transformation ratio can be increased both by accelerated cooling and by alloying [10]. A further advantage of microalloying with vanadium and nitrogen is that the transformation ratio is greatly increased compared to plain carbon steel. This can be attributed to interphase precipitation of V(C,N) concurrent with transformation which slows down the growth of

38 T. Gladman, D. Dulieu and I.D. McIvor, as ref 11, 32

39 A. Blomqvist and T. Siwecki, 'Optimization of Hot Rolling and Cooling Parameters for Long Products', Swedish Institute for Metals Research , IM-3101, (1994)

ferrite and so contributes to its refinement [17, 20]. However, alloying additions such as V and Nb are expensive and their precipitation-hardening potential should be utilized as effectively as is possible, concomitant with attaining acceptable impact toughness.

Accelerated controlled cooling after hot rolling alters the microstructure of plate from ferrite-pearlite to fine grained ferrite-bainite and consequently increases the strength without a loss in low temperature toughness. The deleterious effect of precipitation hardening of Nb(C,N) or V(C,N) on impact toughness is counteracted by improved grain refinement.

The aims of recrystallization controlled rolling especially followed by accelerated cooling (RCR+ACC) are to produce hot rolled steel products having high strength, high toughness and good weldability. The concept of RCR+ACC is attractive in that it is a relatively uncomplicated and high productivity process and can be applied on conventional mills. This procedure is intrinsically more economical than controlled rolling at low finish rolling temperatures and also lends itself well to use in many mills which are not sufficiently strong for low temperature controlled rolling practice to produce heavy plates and long products.

The key to success with recrystallization rolling is to define rolling schedules combining a maximum degree of microstructural refinement with acceptable low rolling loads, good shape control and high productivity. In this context, computer models for calculation of microstructure evolution during hot rolling [13, 21, 22] are invaluable for design of rolling schedules.

2. Microstructure evolution during RCR-processing

2.1. Grain coarsening behavior of microalloyed austenite

Titanium nitrides are the most suitable phase to inhibit the austenite grain growth between rolling passes. Figure 3 compares a number of different V-microalloyed steels with and without Ti and Nb. It is clear that all the steels with small Ti additions (0.01%) show grain boundary pinning for temperatures below 1200°C and so are suitable for RCR processing. By contrast, the plain V-steel has much larger grain size.

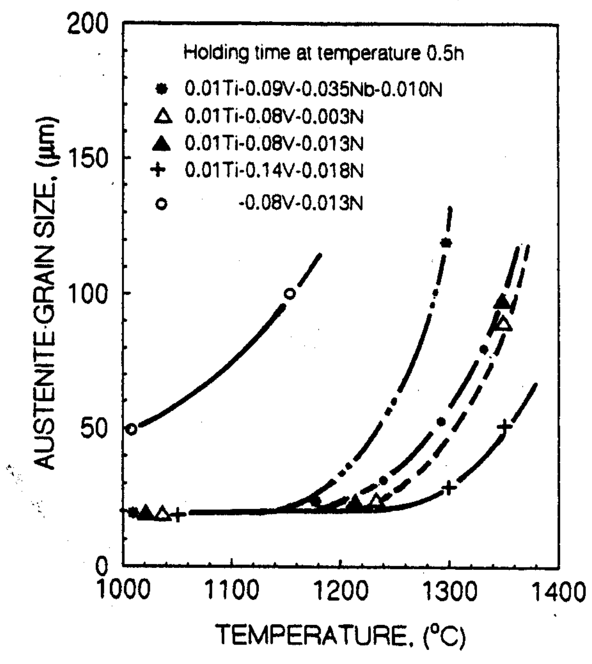

Figure 3. Grain growth of austenite in the as-cast Ti-V-(Nb)-N steels, as a function of a isothermal soaking temperature.

Close control of particle size is vital for successful maintenance of fine grain structures during hot rolling, not least because of the very small volume fraction of TiN which is available for boundary pinning. For a Ti content of 0.01% the volume fraction of TiN is limited to $\sim 2.10^{-4}$. In order to stabilise an as-recrystallized austenite grain size of 20μm, the required particle size according to Hillert's best estimate [23] is <~17nm. The cooling conditions during and immediately after continuous casting are critical for the achievement of such fine TiN precipitate dispersions.

2.2. Modelling of microstructure evolution during hot rolling

The optimization of rolling schedules for processing by recrystallization hot rolling requires detailed knowledge pertaining to the coupling between rolling schedules and development of microstructure. The rolling schedule is characterized by pass temperatures, pass reductions and inter-pass times, whereas the microstructure development is influenced by static recrystallization, static recovery and grain growth of austenite. With intelligent design of the rolling schedules it is possible to obtain a very fine and uniform austenite microstructure. To aid in this work a computer model for calculation of microstructure evolution during hot rolling [13, 21] was developed which is extremely useful for optimising rolling schedules.

2.2.1. Static recrystallization characteristics of the V-C-N-steel

The principal material parameters governing the evolution of austenite microstructure in association with hot rolling are the static recrystallization characteristics: (i) recrystallization kinetics, (ii) recrystallized grain size, (iii) normal grain growth after recrystallization, (iv) static recovery (if recrystallization is not completed between rolling passes).

The progress of static recrystallization after hot deformation was usually studied for each material separately and attention was given to determining empirical relationships suitable for predicting the statically recrystallized fraction and grain size under specific hot deformation conditions. It has been reported previously [21, 22, 13] that the static recrystallization kinetics, $t_{0.5}$ (time for 50% recrystallization), and recrystallized austenite grain size, d^{γ}_{rex}, are dependent upon: strain (ε) (driving force), temperature (T) and pre-existing grain size (d^{γ}_{o}), Zener-Holloman parameter (Z) as well as steel chemistry. The following relationships have been suggested:

$$t_{0.5} = A \cdot d_o^a \cdot \varepsilon^{-b} \cdot Z^{-c} \cdot \exp\left(\frac{Q_{rex}}{RT}\right) \tag{1}$$

and

$$d_{rex} = B \cdot d_o^x \cdot \varepsilon^{-y} \cdot \left[\exp\left(\frac{Q_d}{RT}\right)\right]^{-z} \tag{2}$$

where: A, a, b, c and B, x, y,z are constants that are dependent on the chemistry of the steel.

The above recrystallization equations have been used in computer routines for calculation of austenite microstructure evolution during hot rolling of steel. The models described microstructural changes for particular steels quite accurately. The constants in Equations 1 and 2 are, however, not easily extrapolated to steels of other chemical compositions. For given deformation conditions, the recrystallization kinetics of the C-Mn steel are faster than those of the Ti-V-N and Ti-V-Nb steels. The influence of temperature on the static recrystallization kinetics for different steels after deformation to a true strain of 0.3 is shown in Figure 4. By using $t_{0.5}/d_o^2$ instead of $t_{0.5}$ the effect of grain size on $t_{0.5}$ has been eliminated and so the roles of the microalloying additions on the static recrystallization kinetics are easier to discern. It can be seen that Nb and Ti-V-Nb microalloyed steels show a more complex static recrystallization behavior than the Ti-V and C-Mn steels, especially at temperatures below ~1000°C.

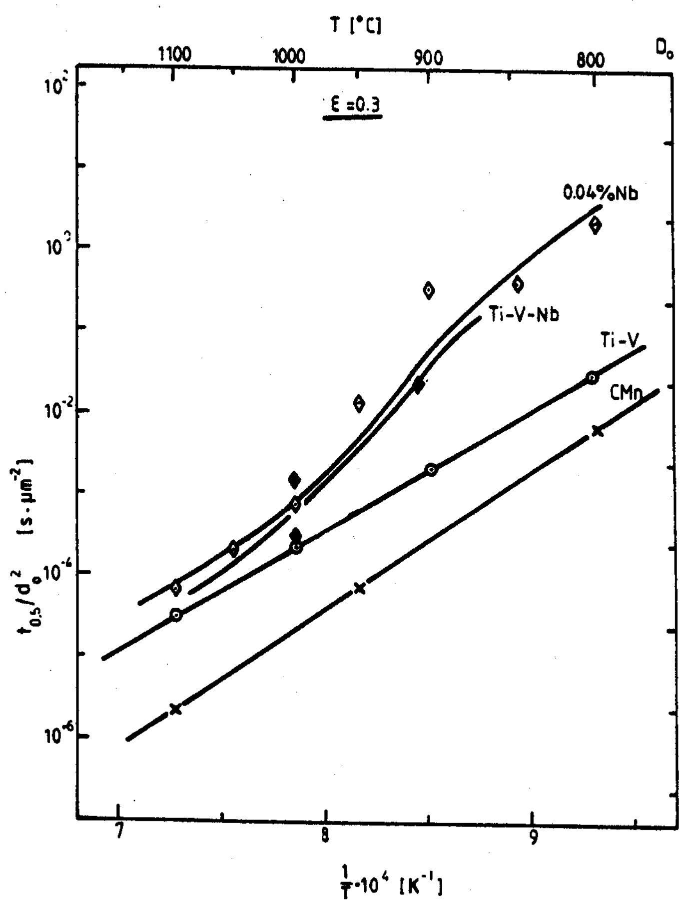

Figure 4. Dependence of compensated $t_{0.5}$ values for recrystallization on temperature for various HSLA steels with base composition (wt%) of (0.11-0.14)C, ~1.5Mn, and ~0.4Si [22].

The driving force of recrystallization, F_R, is the excess energy that has been accumulated during plastic deformation. The driving force for migration of recrystallization front can be expressed as follows:

$$F_R \approx \left(\frac{\mu \cdot b^2}{2}\right)(\Delta\rho) \tag{3}$$

where: μ - shear modulus for steel, b - Burger's vector , Δρ - change in dislocation density as the recrystallization front moves.

No V(C,N)-precipitation normally occurs in austenite of Ti-V-steel and no effect on recrystallization kinetics is observed. However, for steel microaddition with Nb it is necessary to consider the interaction between recrystallization and Nb(C,N) precipitation in austenite according to the following pattern. At high temperature, recrystallization occurs rapidly as microalloying elements are dissolved in solid solution. Precipitation takes place in fully recrystallized austenite, if at all, and may only influence the growth of recrystallized grains. At low temperature, below the recrystallization stop temperature (RST), precipitation takes place before recrystallization, preferentially on the sub-grain boundaries and deformation bands. The most complicated situation is in the intermediate stage, where precipitation competes with recrystallization. The following equation can be used for calculating the retardation stress caused by particles on the recrystallization front migration:

$$F_P \approx \frac{6\sigma \cdot f_P}{\pi \cdot r} \approx \frac{\sigma \cdot Z}{2 \cdot \pi} \tag{4}$$

where: r is the radius of particle, f_P - volume fraction of particles and σ is the specific grain boundary energy. Z is often referred to as the Zener drag parameter (24).

If $F_R < F_P$ then the recrystallization front migration is completely arrested.

2.2.2. Grain growth following recrystallization

It has been observed [21, 25, 26] that the grain growth of microalloyed austenite following static recrystallization after hot deformation does not conform to a simple parabolic relationship, the growth rate at short times being much too rapid in relation to that at long times. This behavior led some authors [25, 26] to suggest a very large exponent, up to 10, in the grain growth equation. Some others [27] described grain growth behavior by a parabolic law with different proportionality constants for short and long times. These approaches do not give a full explanation of the behavior or a correct description of austenite grain growth, either during the inter-pass periods following recrystallization or after the termination of rolling. The grain growth of austenite following static recrystallization should be in accordance with Hillert's theory [28], where the growth rate of 'spherical' grains with radius *R* is given by:

$$\frac{d\left(R^2\right)}{dt} = \frac{1}{2} \cdot \alpha \cdot M_d \cdot \sigma \cdot \left[1 - \frac{Z \cdot R}{\alpha}\right]^2 \tag{5}$$

where M_d is the mobility of the grain boundary, α is a dimensionless constant, σ is the specific grain boundary energy, R is average grain size and Z is the Zener drag which caused by second phase particles.

Normal grain growth

Austenite grain growth following recrystallization has been studied in C-Mn, Ti- and Nb-microalloyed steels (~0.01Nb - 0.004N - 0.08C). The grain-growth rate after complete static recrystallization was found to be very rapid even at temperatures as low as 1000°C.

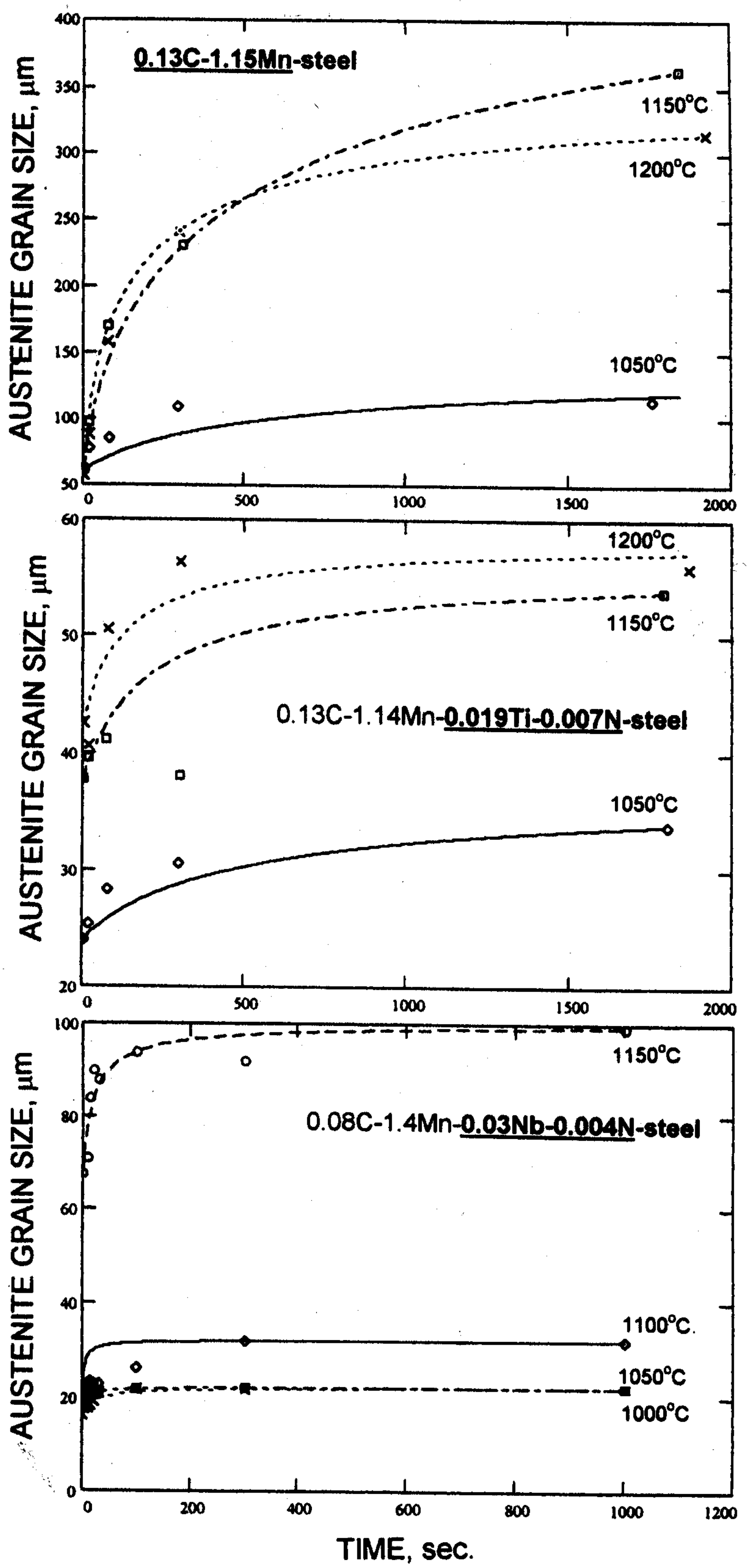

Figure 5. Comparison of the results for grain growth of austenite after recrystallization with the calculated values (based on Equation 5 as a function of holding time and holding (deformation) temperature for: (a) 0.13C-1.15Mn [29], (b) 0.019Ti [29] and (c) 0.03 Nb [30] -steels.

In order to calculate the growth of the austenite grains in the C-Mn-, Ti and Nb-microalloyed steels the above equation has been used. The experimentally measured austenite grain sizes and the calculated curves for normal grain growth are summarized in Figure 5. These results shown that Hillert's equation gave a reasonable agreement with observed results. The Zener parameter determined from the observed results correlates quite well with the volume fraction of pinning particles calculated from the equilibrium solubility data using Thermo-Calc.

The results in Figure 5 demonstrate the increase in grain size with increasing soaking time and temperature following recrystallization. The grain size is strongly dependent upon the content of microalloy in the steel. Very fast grain growth was observed in C-Mn and Nb steels at temperatures higher then 1100°C, whereas, in Ti-steel the grain growth was slower. The explanation of this fact is that the Ti-steel contains very stable TiN precipitates which retard the austenite grain growth.

Abnormal Grain Growth

In alloys containing small precipitates, the possibility of abnormal grain growth in association with hot working cannot be discounted. The kinetics of this process are usually slow and it is only expected under certain conditions such as when a small deformation at high temperature initiates strain induced abnormal grain growth (e.g. following a final sizing pass).

The influence of *sizing passes* on the final microstructure has been studied on Ti-V-N-steel [22, 31]. The RCR hot deformation sequence consisted of four deformations (20-25%) in

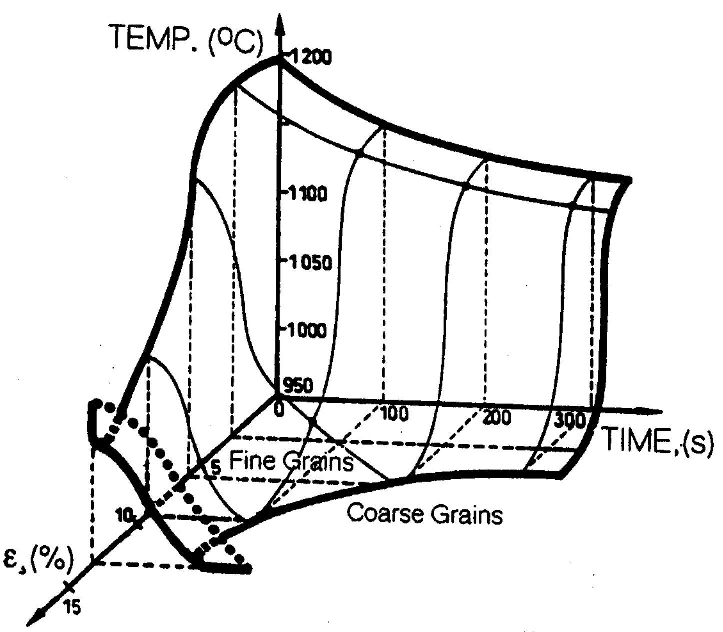

Figure 6. ***Processing window for avoiding coarse grained austenite structures in Ti-V-N steel after the final sizing pass. The surface in ε-T-t space separates fine and coarse grained regions defined at a transitional grain size of 30μm***

the temperature range of 1180°-1030°C followed by final deformation of 3, 6 and 9% at the temperatures 970°, 1030° and 1090°C. Measurements from these experiments were complemented with grain coarsening data for the same steel heated to high temperatures without intervening deformation.

No grain coarsening was observed following a compressive strain of 3% for any of the combinations of holding time and temperature. On the other hand, with 9% strain, coarsened grain structures were visible in all samples excepting that quenched directly after deformation at 970°C. Samples deformed by 6% contained no coarse grains for short holding times at the lower temperatures. However, duplex structures developed at higher temperatures or with longer holding times. Grain structure coarsening of the austenite gives rise to a larger ferrite grain size and in some cases, undesirable Widmanstätten structures after cooling to room temperature. Figure 6 shows the result in ε-T-t space where a surface is defined which separates the safe working region for the final processing from that region where undesirable coarse grain structures can be expected (arbitrarily defined at an austenite grain size of 30 μm). A practical conclusion is that accelerated cooling applied rapidly after the final pass can be of great benefit in reducing the danger of structure coarsening which is a sensitive function of temperature.

2.2.3. Strain hardening and recovery

The evolution of the flow stress of austenite during rolling is determined by generation of dislocations, recovery of dislocations and in some cases dynamic recrystallization. As the driving force for recrystallization is proportional to the dislocation density a good description of dislocation generation and recovery is essential. A good description is also vital for the description of strain induced precipitation.

The flow stress, σ, is well described by

$$\sigma = \sigma_0 + m \cdot \alpha \cdot G \cdot b \cdot \sqrt{\rho} \tag{6}$$

where σ_0 is due primarily to precipitation and to a lesser extent to solid solution and grain boundary strengthening. The Taylor factor, m, will vary with deformation due to the development of a deformation texture. The shear modulus, G, and Burgers vector, b, have a slight temperature dependence. The strength of the barriers, α, is assumed to be independent of temperature and strain.

The evolution of the dislocation density, ρ, with time, t, is considered to be comprised of two parts, generation and recovery, and is described by [26, 32, 33]

$$\frac{d\rho}{dt} = \frac{m}{bL} \cdot \frac{d\dot{\varepsilon}}{dt} - M.\rho^2. \; 1 - \left(\frac{3f_p}{4r\sqrt{\rho}} \right) \tag{7}$$

where ε is the plastic strain, L the mean free distance of dislocation slip, M the mobility for recovery and f_p the volume fraction of particles with radius r. L is described [33] by a constant set of obstacles to slip at a distance L_0 combined with the distance to other dislocations, which is proportional to $1/\sqrt{\rho}$

$$\frac{1}{L} = \frac{1}{L_0} + \frac{\sqrt{\rho}}{k_1} \tag{8}$$

Recovery is assumed to be controlled by climb of dislocations for which M is given by [26, 32]

$$M = M_0 \cdot \frac{D_m \cdot c \cdot b^3 \cdot G}{T \cdot \frac{d\varepsilon}{dt}} \tag{9}$$

where $D_m c$ is the diffusion coefficient expressed as the product of vacancy migration, D_m, and concentration, c. At thermal equilibrium c is given by an Arrhenius expression with an activation energy approximately of the same size as the activation energy for migration.

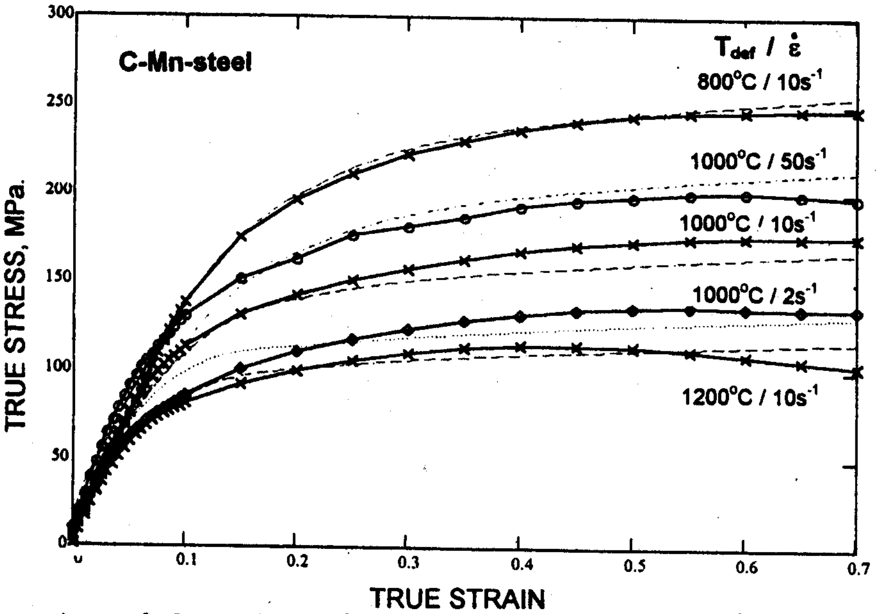

Figure 7. Stress-strain curves for C-Mn-steel at the deformation temperatures of 800, 1000 and 1200 °C. Solid lines are experimental and dashed or dotted lines calculated curves. The strain rates are 10 s^{-1}, x and dashed lines, 2 s^{-1}, ◊ and dotted line, and 50 s^{-1}, o and dash-dotted line. The strain rate is varied at 1000°C.

During deformation substantial amounts of vacancies are created due to dislocation interactions [25] while at the same time vacancies diffuse towards sinks (dislocations). The generation of vacancies during deformation can thus be written as

$$\frac{dc}{dt} = c_1 \cdot b \cdot \frac{d\varepsilon}{dt} \cdot \sqrt{\rho} - c_2 \cdot D_m \cdot c \cdot (c - c_0) \cdot \sqrt{\rho} \tag{10}$$

where the first term describes the generation and the second the diffusion to dislocations, assuming a diffusion distance proportional to the distance between dislocations. c_1 and c_2 are material constants.

The ability of the model to describe material behavior is seen in the two following figures. The first is a fit to compression test data for a plain C-Mn-steel (Figure 7) and the second concerns single and two-step deformation of a niobium steel (Figure 8).

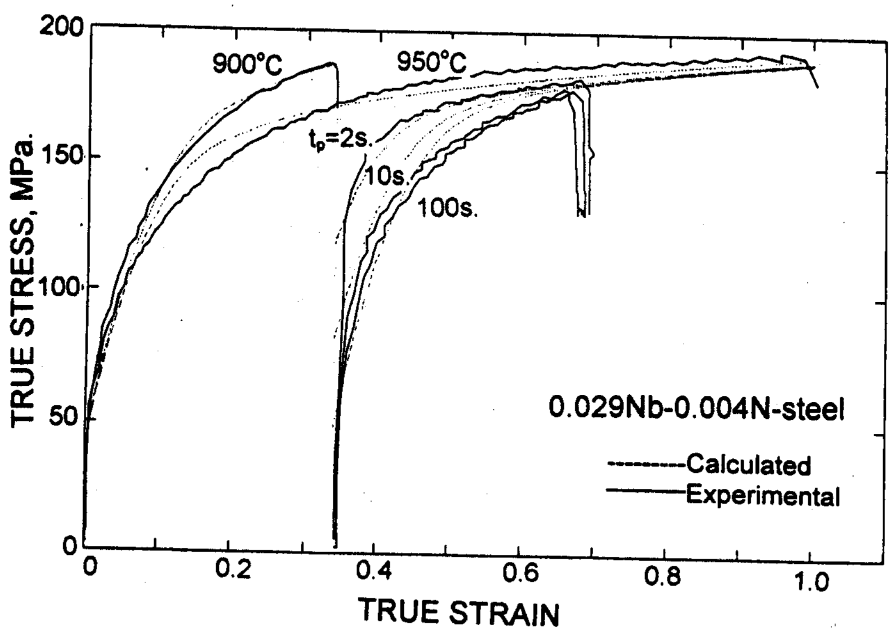

Figure 8. Two step true stress-strain curves for the 0.029Nb-0.004N-steel tested at 950°C with holding times between deformation steps of 2, 10 and 100 seconds. Solid lines are experimental and dashed lines calculated curves.

The model is thus able to describe both strain hardening and dynamic recovery as well as static recovery during hold times. The same activation energies are used for both steels. As expected the mobilities differ slightly as do the parameters for dislocation generation. In order to obtain a reasonable description of strain hardening at strain above ~0.2 the Taylor factor has been calculated as shown in Figure 9, based on an initially random texture and slip on octahedral planes [34].

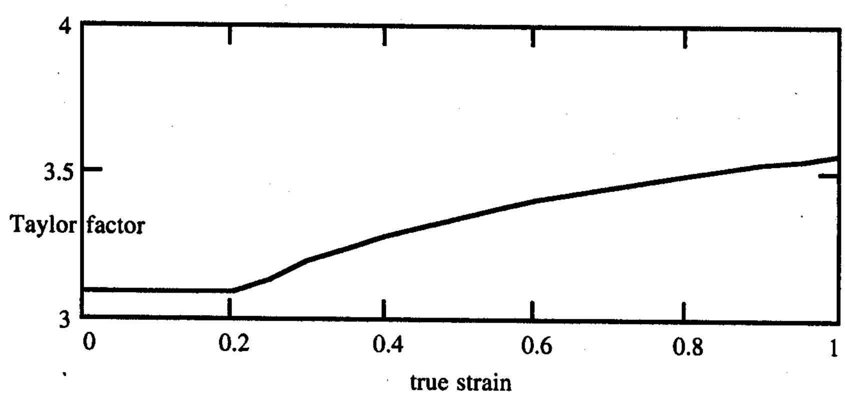

Figure 9. Taylor factor as a function of true strain

2.3. Prediction of microstructure evolution during recrystallization hot rolling

The computer models of austenite microstructure development have been extended to many different steel compositions and permit comparison to be made of the effect of various material parameters. Figure 10 shows examples of predicted microstructure evolution during realistic (full scale) hot rolling of 0.01Ti-V-N [5] and 0.01Ti-V-Nb [35] steels having different levels of vanadium, niobium and nitrogen when subjected to the same rolling schedule. These data refer to rolling of

25 mm plate in 11 passes starting from 1100°C, the initial grain sizes (d_o^γ) after reheating were taken to be 20 μm for both 0.01 Ti-V-N steels and also 500μm in the case of the lower alloyed steel, whereas, d_o^γ for the Ti-V-Nb steel was taken to be 55 μm. As is clear from Figure 10 the steel with high nitrogen is characterized by the most effective microstructural refinement.

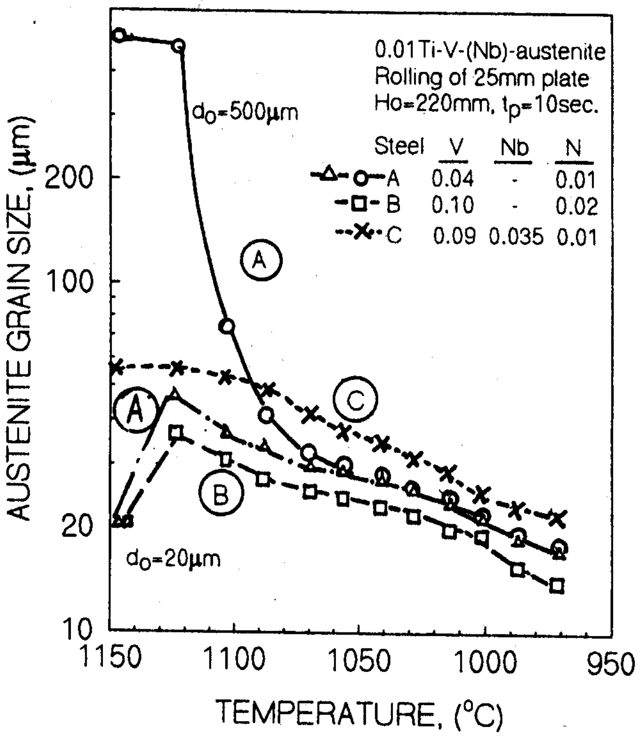

Figure 10. Development of austenite microstructure for Ti-V-(Nb)-N-steels during simulation of an industrial plate hot rolling schedule.

Figure 10 illustrates also the effect of varying the initial austenite grain size on microstructure evolution during an otherwise identical rolling schedule. The large d_o^γ (500 μm) is intended to simulate the situation where abnormal grain growth has occurred during reheating of a Ti-V steel. The widely differing d_o^γ values have no bearing whatsoever on the final austenite grain size. Indeed the grain size is virtually identical in both cases after as few as four passes. This means that for rolling schedules comprising more than 4 -5 passes, the final austenite grain size is predicted to be insensitive to the extent of austenite grain growth during reheating [5].

It can also be seen that the Ti-V-Nb-steel usually gives larger austenite grains after static recrystallization than the Ti-V-steels. This may reflect a smaller inhibiting effect (Zener term) for the larger (Ti,Nb,V)N particles in this steel which permits more grain growth to occur during or immediately after recrystallization. Alternatively it may be due to an effect of Nb(C, N) precipitates or complexes hindering nucleation of recrystallization and so increasing the final as - recrystallized grain size [35].

In industrial practice there are, however, other factors to be considered which restrict the range of parameters in hot working. Another way of presenting such data is to plot contours of equal recrystallized grain size, d_{rex}, in a map having axes of temperature and strain (T-ε) as shown in Figure 11 [36].

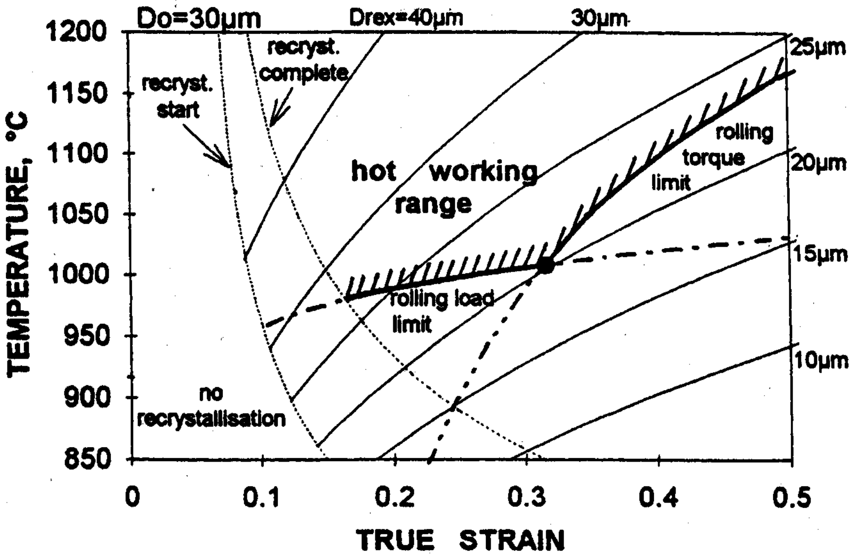

Figure 11. Effect of temperature and strain in a single stage deformation on as-recrystallized grain size, plastic work of deformation, mean flow stress, and the possible optimum conditions for hot rolling [36]

During the initial stages of rolling much mechanical energy has to be applied to deform the thick work piece and the rolling torque sets a practical limit on the deformation which can be achieved in each pass. The work needed to deform a given volume of metal can be found from integrating the flow stress - strain curve of the steel at the appropriate temperature. Contour lines showing equal levels of plastic work can be plotted on a T-ε map in Figure 11.

In the later stages of rolling, however, when the metal becomes harder at lower temperature, the practical limit in rolling is more often governed by the rolling loads which can be applied. It is principally the mean flow stress which determines the rolling load and this can also be deduced from the flow stress-strain curve. Contour lines of mean flow stress for the same steel can also be mapped as functions of T and ε in Figure 11.

For a given mill and size of plate, both torque and load will set limits for operation as indicated schematically by the boundary lines in Figure 11 where hot working can only be carried out on the shaded side of the boundary. The finest as-recrystallized grain structure which can be achieved is found at the crossing point of the two boundary lines. This combination of strain and temperature therefore represents the best conditions for hot rolling in order to obtain the optimum microstructure for a given steel plate and rolling mill. Although Figure 11 refers to a single deformation step and a single initial austenite structure, the as-recrystallized grain size d_{rex} always varies monotonically with the initial grain size d_o so that exactly the same conclusions will apply in the case of multiple RCR deformations.

The present treatment is greatly idealized but it does lead to an important conclusion, namely that RCR processing should not be carried out at unduly low temperatures. An optimum temperature exists which may be in the region 950^o to 1050^{o}C depending on the relevant conditions of material and the equipment being used.

As described above, the optimum finish rolling conditions involve large reductions at relatively high temperatures. In industrial practice, however, the final pass cannot normally be so large since there are conflicting requirements of flatness and dimensional tolerance. There is, accordingly, a danger that *the final sizing pass* may bring about an unfavorable coarsening of the austenite structure, as was discussed in the previous section.

2.4. Phase transformation of recrystallization controlled rolled austenite

The austenite condition has a profound influence on the final ferrite grain size. It has been shown [11] that for the same austenite grain boundary area per unit volume, Sv, of Nb-microalloyed austenite the ferrite develops a smaller grain size when transformed from an unrecrystallized austenite rather than from a equiaxed (recrystallized) one. For V- and Ti-V-steels, however, [5, 12] it has been observed that elongated austenite grains (CR) transform to ferrite with virtually the same grain size as that produced from equiaxed grains (RCR) of equivalent size (same Sv). It is clear from Figure 12 that excellent ferrite grain refinement can be achieved, even from recrystallized austenite, provided that Sv is sufficiently high (>100mm^{-1}) and especially if accelerated cooling is applied. The ferrite grain size, d^{α}, is independent of grain shape and processing method, as shown in Figure 12, where the V-steels occupy an intermediate position between the two lines reported for Nb microalloyed steels.

Different combinations of steel composition and processing parameters have been studied [8, 35] as regards their effect on the $\gamma \rightarrow \alpha$ transformation and ferrite refinement in Ti-V-(Nb)-microalloyed steels. A beneficial effect of nitrogen and cooling rate on the grain size transformation ratio, d^{γ}/d^{α}, has been observed. The transformation ratio increases moderately with increasing nitrogen content and strongly with cooling rate.

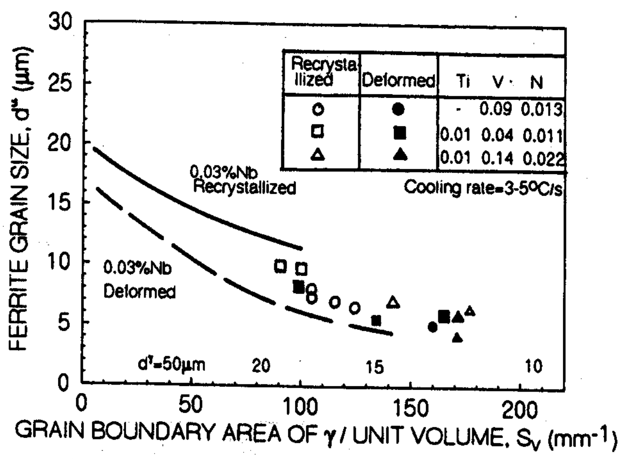

Figure 12. Dependence of ferrite grain size, (d^{α}), on the total area of austenite grain boundary per unit volume, (Sv). Data points are for Ti-V and V microalloyed steels [5, 12]. The curves refer to Nb-microalloyed austenite [11].

On the basis of the phase transformation data obtained from CCT dilatometer studies, it was deduced that the cooling rate subsequent to RCR (during the $\gamma \rightarrow \alpha$ transformation) should be, in principle, restricted to a maximum of 10-12^{o}C/s and the finish cooling temperature should not be lower than 500^{o}C if the formation of martensite or bainite is to be avoided.

3. Mechanical properties and microstructure of RCR+ACC material

The mechanical properties and the final microstructure of Ti-V-N-microalloyed steels are dependent on the following (process) parameters:

- reheating temperature (Treh),
- rolling schedules: reduction (Red) and finish rolling temperature (FRT),
- cooling parameters: accelerated cooling rate (ACC) and finish accelerated cooling temperature (FCT),
- steel chemistry.

3.1. Effect of reheating temperature

The slab reheating temperature has a strong influence on the strength, toughness and microstructure of microalloyed steel in the as thermo-mechanical controlled processing condition. A low slab reheating temperature gives finer austenite grains, refines the final microstructure of the materials and as a consequence improves the low temperature toughness for steels processed in a similar manner. This is mainly attributed to the more profuse fine precipitation remaining after low temperature reheating which more effectively resists grain growth of austenite following recrystallization.

However, the yield stress and tensile strength decrease because lower reheating temperature reduces the amount of dissolved vanadium (and niobium) in the austenite (see Figure 2) and accordingly the potential for precipitation hardening after cooling. For a Ti-V-N microalloyed steel it has been found that a reduction in reheating temperature from 1250° to 1100°C reduces the yield stress by about 40MPa while decreasing the ductile-brittle transition temperature by about 15°C [8].

3.2. Effect of rolling schedules

The finish-rolling temperature and the deformation in the final pass are usually important TMCP-parameters, affecting strength and toughness. The effect of the finish rolling temperature on the

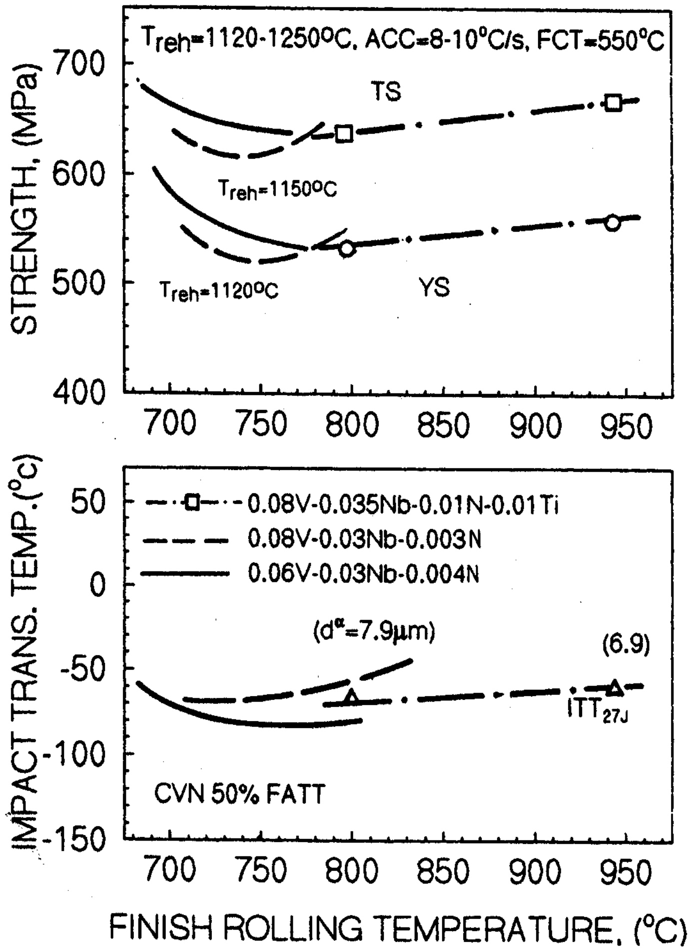

Figure 13. Effect of finish rolling temperature (FRT) on the strength, impact toughness and ferrite grain size of Ti-V-(Nb)-N -steels processed in a similar manner

microstructure and mechanical properties for various microalloyed steels is summarized in Figure 13 using data from various sources [8, 19, 20, 21]. Figure 13 shows that the best combinations of low temperature toughness and tensile strength are obtained in Ti-V-(Nb)-N steels for finish rolling temperatures close to Ar_3 (i.e. conventional CR practice). However, an almost equally good combination of strength and toughness was obtained for a FRT of 950°C by recrystallization controlled rolling.

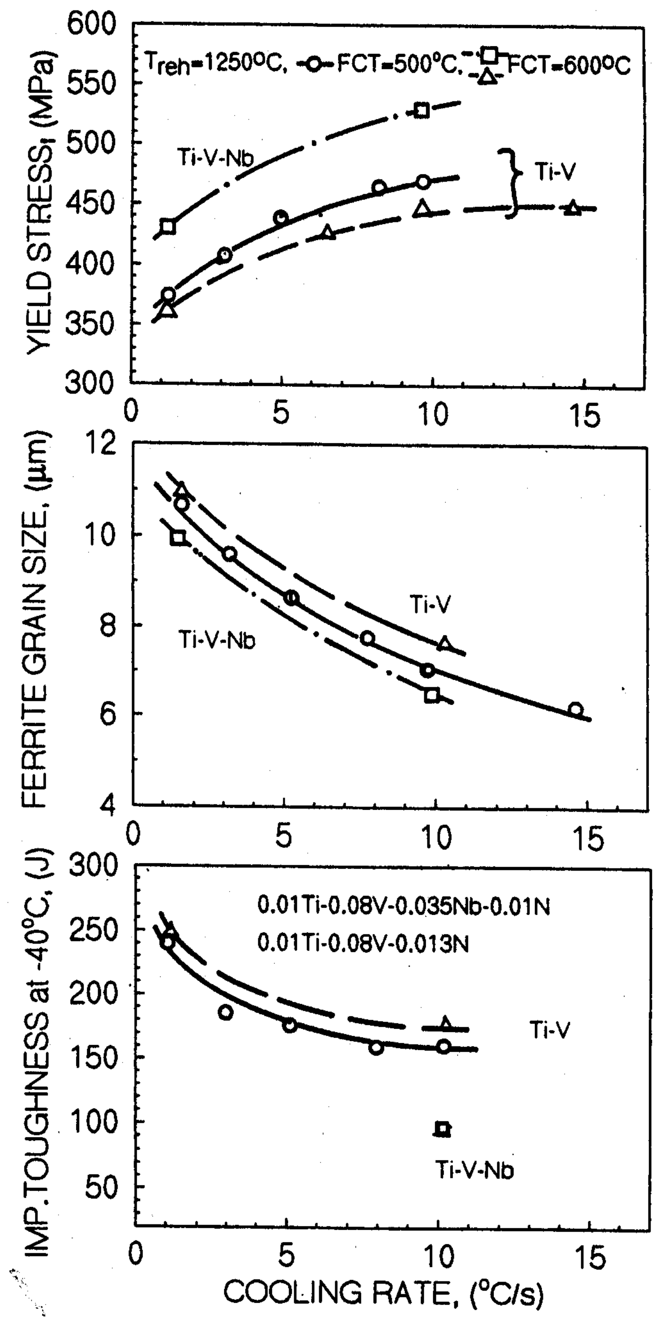

Figure 14. Effect of cooling rate from finish-rolling temperature of 1030°C to finish-cooling temperature (FCT) on the ferrite grain size, yield stress and toughness of Ti-V-(Nb)-N steels. Final reduction was 25%

3.3. Effect of accelerated cooling parameters

The effect of cooling rate and finish cooling temperature on the final microstructure, yield strength and toughness of Ti-V-(Nb)-N-steels after RCR is illustrated in Figure 14. It is apparent from the figure that the final microstructure and mechanical properties are greatly influenced by cooling rate.

The yield stress of Ti-V-(Nb)-N-steels rises as the cooling rate increases, although the variation of strength with cooling rate is smaller at higher rates than lower (<7°C/s). At very high cooling rates (15°C/s) the austenite transforms to a ferrite-bainite microstructure. Impact toughness decreased with higher cooling rates and this effect has been found to be greater in high N-steels.

The final ferrite grain size depends only weakly on the finish-cooling temperature in the range of 400-600°C although the second phase consists of bainite when FCT is below 500°C. Yield strength on the other hand, depends on the finish cooling temperature and increases with decreasing FCT down to 500°C. The appearance of bainite in the structure of Ti-V-N and Ti-V-Nb- steels changes the dependence of yield strength on FCT.

Impact transition temperatures of RCR+ACC material increase as the cooling rate becomes higher despite the fact that the ferrite grain size decreases. This is a result of the precipitation strengthening of V(C,N) or (V,Nb)(C,N) and at low FCTs the increasing volume fraction of bainite in the microstructure. Cooling rate is a principal factor influencing the size distribution of precipitates. A greater proportion of smaller particles and reduced distance between rows has been observed when the cooling rate increased [8, 19], so producing more efficient precipitation strengthening.

3.4. Effect of nitrogen and carbon content

Nitrogen's effect in connection with recrystallization controlled rolling and accelerated cooling is demonstrated in Figure 15. Material with higher N content is significantly stronger than that with low N processed in the same manner, although with some sacrifice of toughness. The strength is enhanced much more by accelerated cooling in the higher N steel and this effect is more pronounced in the case of yield stress than tensile strength.

It may be seen that the strengthening potential of vanadium can be effectively utilized only at higher levels of nitrogen and that increased cooling rate has a profound effect especially at these higher levels. Ferrite grain sizes of low nitrogen steel are significantly larger than those of the higher N steel, in particular for slow cooling rates. The effectiveness of nitrogen in the Ti-V steel in the as controlled rolled condition is much smaller than in the RCR and ACC condition as shown previously [8, 37]. For low N steel (0.01Ti-0.08V-0.003N) the yield stress after low temperature CR was almost identical with that following RCR+ACC processing. However, for higher N steel (0.013%N) the yield stress was ~50MPa less after the CR treatment, which is attributed to the loss of precipitation hardening effect due to premature formation of coarse VN particles in austenite.

Some results illustrating the effect of processing method by recrystallization hot rolling, controlled rolling and normalizing (the processes were followed by air cooling at 0.9 °C/s) on precipitation strengthening contribution to yield strength, ΔRp, for V-N steels are presented in Figure 16 [12] (in evaluating ΔRp the regression formula recommended by Gladman [38] has been adopted).

For RCR steels, ΔRp increases linearly with nitrogen content, for normalized material, however, the precipitation hardening contribution to yield stress saturates at high nitrogen levels because of the increasing amount of V(C,N) remaining undissolved at the normalizing temperature.

Figure 16 also demonstrates that ΔRp is lower after processing by CR with FRT=800°C than after RCR with FRT=950°C. This difference in ΔRp can be reconciled with the loss of precipitation hardening potential when deformation-induced precipitation of V(C,N) accompanies rolling at low finishing temperatures.

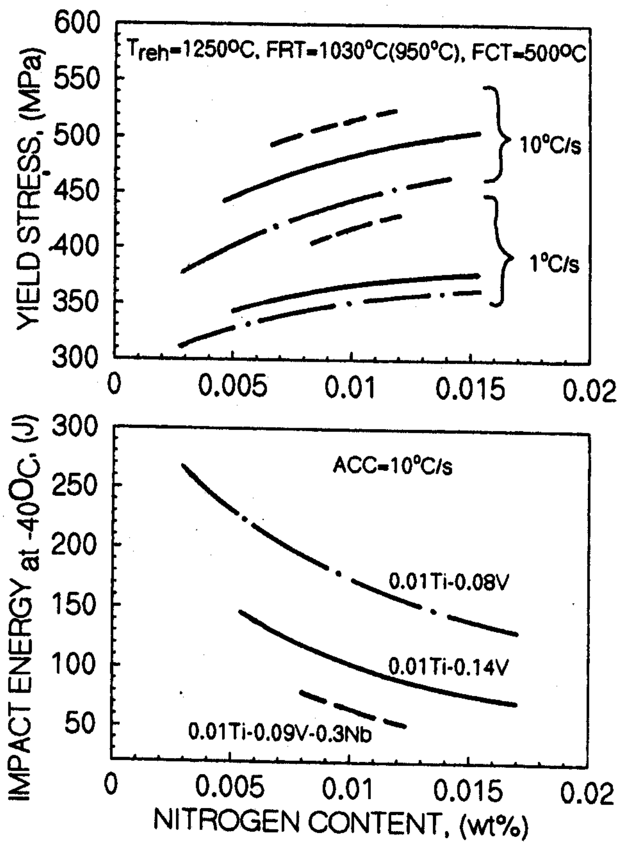

Figure 15. The effect of nitrogen, vanadium (niobium) and cooling rate on the yield stress and impact toughness of Ti-V-(Nb)-N-steels after recrystallization controlled rolling.

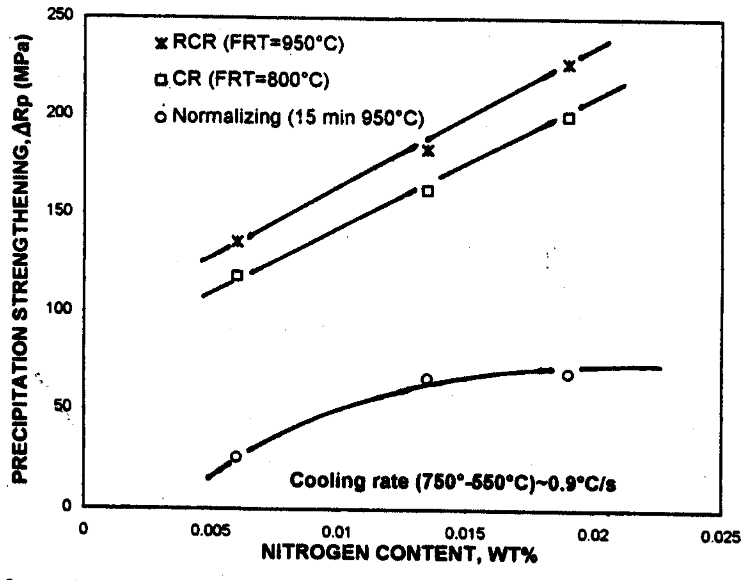

Figure 16. Effect of processing method on the precipitation strengthening derived from V(C,N). Base composition of steel (wt%): 0.12C, 1.35Mn, 0.35Si, 0.02Al and 0.09V [12].

Carbon content and alloying elements such as Mn and Ni stabilize the austenite and lower the Ar_3-temperature. This suppression of the $\gamma \rightarrow \alpha$ transformation temperature leads to a refinement of the final structure by decreasing the growth rate of the ferrite grains.

The principal difference between the steels which are intended for long products and plate steels lies in the higher carbon contents (~0.2%C) of these former steels. The amount of pearlite is greater and they show a stronger tendency to develop bainitic or other acicular microstructures. Since the processing of sections and other long products necessitates relatively high finishing temperatures, the RCR processing route together with appropriate microalloying should lend itself well for obtaining improved mechanical properties. Figure 17 shows a plot of impact transition temperature against yield strength for long product steels and comparably treated plate steels (FRT=1000°-1050°C, ACC=8°C/s and FCT=500°-550°C). The impact transition temperature is seen to be strongly dependent on the carbon content of the steel.

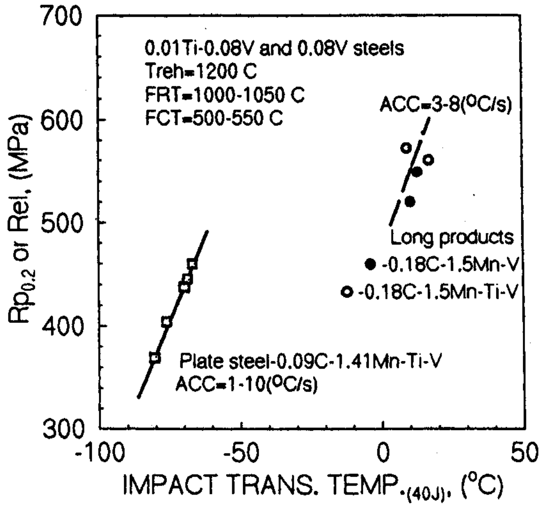

Figure 17. Effect of carbon content in the Ti-V and V microalloyed steels [39] on the strength vs. toughness of the specimens processed in the similar manner.

4. Some results of full scale industrial processing

The Ti-V-N steels described in the previous section have also been rolled to 20 mm or 25 mm plate on a commercial mill using the experience previously gained from laboratory simulations [8]. The steels were rolled using RCR practice with and without application of accelerated cooling. They were also processed with a low finish rolling temperature CR practice such that the austenite was in a substantially deformed condition prior to transformation. Table 1 summarizes the important parameters relevant to their processing.

Microstructures and mechanical properties of the resulting plates for Ti-V-N-steel have been determined. The effect of accelerated cooling on grain size is very evident and pearlite banding is almost completely eliminated. The finest grain structures are, however, obtained from the CR route and these should be capable of some further refinement with application of accelerated cooling.

Despite the finer grain size of the CR plates these have slightly lower strength than the RCR+ACC ones as a result of premature precipitation of VN in austenite during the final low temperature rolling passes. Mechanical properties of the as rolled plates of 0.01Ti-0.08V-0.013N-steel are summarized in Figure 18. Toughness is in all cases excellent (ITT_{40J} <-80°C)but best for the CR treatment because of the very fine ferrite grain size and smaller precipitation hardening contribution.

Process	Treh (^oC)	FRT (^oC)	N^o of passes	Time (min)	Thickness (mm)	Cooling rate ($^oC/s$)	FCT (^oC)
RCR	1250	1050	10	3 min.	25	0.4	-
RCR+ACC	1250	1050	10	3 min.	25	7	600
CR	1250	800	13	8 min.	20	0.5	-

Table 1. Process parameters for full scale production.

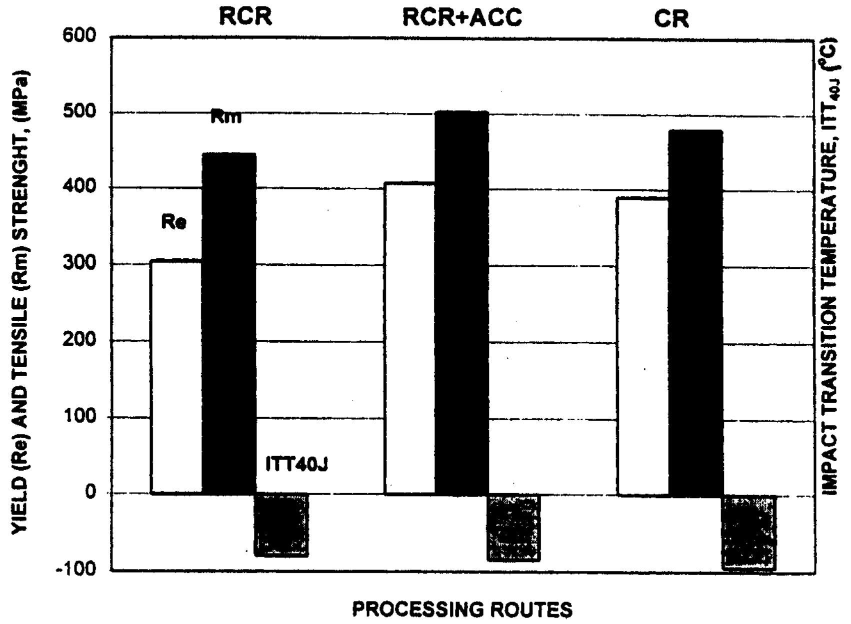

Figure 18. The mechanical properties of the commercially processed 0.01Ti-0.08V-0.013N steel plates produced by different full scale processing routes.

The measured rolling loads for the two types of processing are compared in Figure 19. It is evident that the maximum loads are ~ 25% higher for the CR process due to the lower temperatures and the accumulated deformation below the recrystallization stop temperature. The longer time necessary for CR is mainly due to the delay for cooling from 1100°C to 920°C at the 40 mm gauge but partly also due to the greater number of finishing passes.

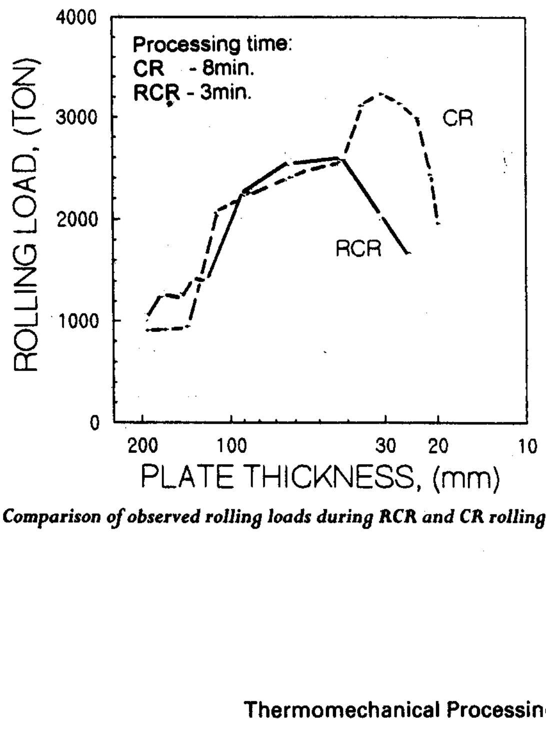

Figure 19. Comparison of observed rolling loads during RCR and CR rolling of Ti-V-N-steel [22].

5. Conclusions

Recrystallization controlled rolling together with accelerated cooling is an alternative to low temperature controlled rolling as a process for manufacturing steel plate with high strength and toughness. Advantages of the RCR process are its high productivity and lower mill loading as compared to CR. In general there is an optimum finish rolling temperature which depends on the steel composition and the power of the rolling mill being used in the RCR processing.

Grain growth of austenite after recrystallization has been studied in HSLA-steels and Hillert's equation for grain growth has been adopted. This should be included in computer routines for calculation of austenite microstructure evolution during RCR processing.

The flow stress as well as static recovery in HSLA steels seem to be well described by a unified model with recovery based on climb of dislocations and vacancies creation by plastic deformation

Prevention of grain growth between rolling passes and after the final pass is essential for austenite structure refinement. This can be achieved by a small addition (~ 0.01%) of titanium which generates stable TiN particles. The small austenite grain size means that a fine ferrite grain structure can be obtained in the as-rolled condition, which confers both strength and toughness.

Accelerated cooling increases the strength of the product both by refinement of ferrite grain size and by generating smaller and more effective microalloy precipitates. Vanadium is required together with nitrogen to increase the strength level via precipitation of VN. The influence of cooling rate is greatest in the range 1 - 5^{o}C/s, but further improvement is found at higher cooling rates.

RCR+ACC processing lends itself well to long products as well as plate production. However, at higher carbon contents there is a significant loss of toughness.

Acknowledgments

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