Vanadium Microalloyed Bainitic Hot Strip Steels*1

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The aim of the present study was to investigate the possibilities of reaching yield strengths beyond 600 MPa for low carbon bainitic hot strip steels by vanadium microalloying together with suitable base alloying. The processing conditions and levels of carbon and nitrogen chosen in this laboratory investigation correspond to those of a typical 8 mm hot strip steel containing 0.04 mass% carbon and 0.010 mass% nitrogen from electric arc furnace practice processed in conventional or compact strip mills. It was found that a base alloying corresponding to 1.4 mass% Mn, 1.0 mass% Cr and 0.25 mass% Mo is required to form a fully bainitic structure after coiling at 400°C. The decisive factors determining the strength of bainitic hot strip steels are firstly the bainite transformation temperature and secondly the extent to which recovery of the densely dislocated bainitic ferrite can be prevented. The results of this study demonstrate that vanadium microalloying effectively prevents the recovery of the bainitic ferrite and leads to retention of the strength of the virgin bainite after coiling. This is primarily due to retardation of recovery by fine vanadium carbonitrides precipitates on dislocations and only to a lesser extent to true precipitation strengthening. With 0.08 mass% V together with 0.010–0.020 mass% N the yield strength lies in the range of 750–790 MPa compared to 680 MPa for a similar reference steel without vanadium. By raising the chromium content to 2%, yield strengths in the range of 840–880 MPa have been reached. This is attributed to a lowering of the bainite transformation temperature resulting from the higher base alloying.

KEY WORDS: high strength steel; bainite; alloying; hot rolling; cooling; microstructure; mechanical properties.

1. Introduction

The yield strength of low carbon ferrite–pearlite structural steels can be raised to a maximum of about 600 MPa by microalloying. The added strength by micro-alloying is derived from ferrite grain refinement and precipitation strengthening. In order to reach still higher strength levels bainitic or tempered martensitic microstructures have to be utilised. In this case the enhanced strength is due to very fine bainite or martensite laths, down to about 1 μm, and a high dislocation density in the ferrite caused by the displacive nature of the bainite and martensite transformations.

This procedure to reach yield strengths beyond 600 MPa has long been used for structural plate steels, either by a separate quenching and tempering process after hot rolling or by a direct quenching operation in the plate mill. The primary role of alloying is here to generate adequate hardenability to ensure transformation to bainite/martensite during the quenching process. Enhanced manganese contents and additions of chromium and molybdenum are most frequently used to achieve this. Microalloying with niobium and vanadium is also often used, but it appears that their role is mainly to control the austenite grain size and to resist softening of the bainite/martensite during the rapid cooling or during subsequent tempering. In this way, yield strengths of 800–1000 MPa are readily reached for carbon contents of 0.04–0.08%.*1–4,*2

The usefulness of bainite for producing high strength plate steels is well established but its application to hot strip steels with yield strength above 600 MPa is not so obvious. When bainite is formed during cooling to the coiling temperature or in the coil, it is expected to recover and soften significantly during the prolonged holding at elevated temperatures after cooling. However, the ability of vanadium to form dense precipitations of V(C, N) in ferrite, especially with enhanced contents of nitrogen, suggests that V–N-microalloying should be effective in preventing this recovery of bainite and so make it possible to produce hot strip steels with yield strengths beyond 600 MPa directly in the mill. This concept is supported by the well known fact that vanadium is effective in resisting softening and indeed producing secondary hardening during tempering of martensitic microstructures in tool steels.5

Hence, it was decided to perform an exploratory investigation as to what base alloying is required to generate bainite in a low carbon hot strip steel after finish rolling and, furthermore, whether V-microalloying will be capable of adding strength above 600 MPa in such steels.

*1 This paper is dedicated to Michael Korchynsky, who recently retired from Stratcor Inc. after 60 years service to steel and vanadium microalloying. The authors of this paper, as well as former colleagues in Stockholm, are indebted to him for his profound engagement in our research and for countless rewarding discussions over the years.
*2 All steel compositions in this paper are expressed in mass%.

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2. Alloy Design and Experimental Steels

Out of the normal alloying elements, manganese, molybdenum and chromium are the most effective for raising the hardenability of steel, defined as their ability to prevent formation of ferrite and pearlite during hardening. Their hardenabilities expressed as the Grossman coefficient are all high but decline somewhat in the order Mn, Mo and Cr, viz. 4.10, 3.14 and 2.83. In the present case, the retardation of bainite formation is particularly significant. Here, the coefficient differs more between these three elements, Mn 4.10, Cr 1.16 and Mo 0.7. Hence, to stimulate bainite formation and at the same time prevent formation of ferrite/pearlite, molybdenum is most effective, followed by Cr and then Mn. In accordance with this, 1.4% Mn, 1.0% Cr and 0.25% Mo were chosen as the standard composition to create an adequate base hardenability. It was estimated that this base composition should be sufficiently alloyed and well balanced to produce a fully bainitic microstructure in 8 mm hot strip steel cooled at a rate of 30°C s\(^{-1}\) after finishing roll when subsequently cooled in the range 600–400°C. Some variations were also made in order to find out what margins exist before ferrite is formed and to what extent the expensive molybdenum can be replaced by less expensive chromium. Since nitrogen is known to promote fine V(C, N) precipitation and, therefore, is expected to effectively retard recovery and softening of the bainite, perhaps even providing secondary hardening, the nitrogen content was varied in the range 0.010–0.038%, where the lower limit corresponds to the content that is typical for electric arc furnace (EAF) steels. The vanadium and carbon contents were selected as 0.08% V and 0.04% C as standard for all the experimental steels. The low level of carbon serves two important objectives. First it should ensure good weldability of the steel. Secondly it moves the steel composition outside the peritectic range so that the castability should be adequate also for manufacture by thin slab casting.

Table 1 shows the compositions of the experimental steels studied in the present investigation. As can be seen in two of the steels, E1 and F1, the aluminium content has been kept low in order to see whether AlN formation possibly could affect the V(C, N) precipitation for normal aluminides.

The vanadium and carbon contents were selected as 0.08% V and 0.04% C as standard for all the experimental steels. The low level of carbon serves two important objectives. First it should ensure good weldability of the steel. Secondly it moves the steel composition outside the peritectic range so that the castability should be adequate also for manufacture by thin slab casting.

<table>
<thead>
<tr>
<th>Steel</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
<th>Mo</th>
<th>Al</th>
<th>N</th>
</tr>
</thead>
<tbody>
<tr>
<td>A5 CrMoV</td>
<td>0.037</td>
<td>0.11</td>
<td>1.44</td>
<td>0.96</td>
<td>0.26</td>
<td>0.061</td>
<td>0.03</td>
</tr>
<tr>
<td>B1 CV</td>
<td>0.044</td>
<td>0.11</td>
<td>1.27</td>
<td>1.04</td>
<td>0.004</td>
<td>0.078</td>
<td>0.037</td>
</tr>
<tr>
<td>G2 CrMo, reference</td>
<td>0.043</td>
<td>0.12</td>
<td>1.40</td>
<td>1.03</td>
<td>0.25</td>
<td>0.001</td>
<td>0.038</td>
</tr>
<tr>
<td>D1 CrMoV1Mn</td>
<td>0.037</td>
<td>0.11</td>
<td>0.98</td>
<td>0.98</td>
<td>0.26</td>
<td>0.088</td>
<td>0.034</td>
</tr>
<tr>
<td>E1 CrMoVNaI-low</td>
<td>0.045</td>
<td>0.11</td>
<td>1.38</td>
<td>1.03</td>
<td>0.26</td>
<td>0.078</td>
<td>0.003</td>
</tr>
<tr>
<td>F1 1.4Cr/1.2MoVNaI-10W</td>
<td>0.045</td>
<td>0.10</td>
<td>1.35</td>
<td>1.38</td>
<td>0.12</td>
<td>0.08</td>
<td>0.003</td>
</tr>
<tr>
<td>H1 1.4Cr/1.2MoVN</td>
<td>0.043</td>
<td>0.10</td>
<td>1.50</td>
<td>1.36</td>
<td>0.12</td>
<td>0.078</td>
<td>0.023</td>
</tr>
<tr>
<td>G1 2.0Cr/1.2MoV0.036N</td>
<td>0.041</td>
<td>0.11</td>
<td>1.31</td>
<td>2.00</td>
<td>0.12</td>
<td>0.073</td>
<td>0.033</td>
</tr>
<tr>
<td>G2 2.0Cr/1.2MoVN</td>
<td>0.048</td>
<td>0.10</td>
<td>1.37</td>
<td>2.00</td>
<td>0.12</td>
<td>0.079</td>
<td>0.026</td>
</tr>
</tbody>
</table>

The contents of P and S were low and in the range of a few thousands of a percent.

3. Experimental Procedure

The steels of Table 1 were melted and cast to 40×40×200 mm ingots and after reheating for 1 h at 1 150°C, were hot rolled to 20 mm plates. Specimens were machined from the plates, φ5.0×10.0 mm, both for dilatometry to determine the CCT curves and for deformation dilatometry to simulate the last stages of hot strip rolling and coiling.

Dilatometer experiments were carried out in a high speed dilatometer to determine the temperature interval for the bainite transformation at varying cooling rates to provide data for construction of CCT curves and to simulate hot strip rolling. All dilatometer samples were first reheated in a furnace at 1 150°C in vacuum for 30 min and quenched to room temperature. For CCT determination they were then reheated at 1 150°C for 3 min, cooled to 1 000°C and deformed 25% to refine the austenite grain structure, followed by cooling at 2°C s\(^{-1}\) to 900°C and thereafter cooled at 5, 10, 30, 60 and 100°C s\(^{-1}\) to room temperature.

The thermomechanical controlled processing of 8 mm hot strips was simulated by hot deformation dilatometry by first reheating specimens in the dilatometer in vacuum at a rate of 5°C s\(^{-1}\) to 1 150°C and holding for 3 min before subjecting them to one of the following hot deformation schedules:

1. Cooling 2°C s\(^{-1}\)→1 000°C/def.25%→cooling 2°C s\(^{-1}\)→930°C/def.25%→cooling 2°C s\(^{-1}\)→900°C→cooling 30°C s\(^{-1}\)→coiling temp. (400–600°C)→cooling 10°C h\(^{-1}\) to RT.

The purpose of this was to simulate as closely as possible an industrial strip production process, including the final coiling treatment.

2. Cooling 2°C s\(^{-1}\)→1 000°C/def.25%→cooling 2°C s\(^{-1}\)→930°C/def.25%→cooling 2°C s\(^{-1}\)→900°C→cooling 30°C s\(^{-1}\) to RT.

These treatments aimed to study the condition of the steels after simulated processing but without the final cooling operation.

3. Cooling 2°C s\(^{-1}\)→1 000°C/def.25%→cooling 2°C s\(^{-1}\)→930°C/def.25%→cooling 2°C s\(^{-1}\)→900°C→cooling 30°C s\(^{-1}\)→RT→tempering at 350–600°C for 5 h→cooled to RT.

These treatments were carried out to compare the behaviour of the present materials with the known effect that vanadium has in quenched and tempered steels. The enhanced hardenability of the 2% Cr steels appears to allow a fully bainitic structure to be formed at lower cooling rates after finish rolling. Accordingly samples of steel G2 were run through route 2 with a cooling rate of 20°C/s which would correspond to processing of a 12 mm hot strip. All deformations were made at a strain rate of 2 s\(^{-1}\).

Compression cylinders, φ4.0×4.0 mm, were machined from the thermo-mechanically processed samples. Com-
pression tests were performed in an Instron electro-mechanical testing machine for evaluation of the offset yield strengths, \( R_{p0.2} \) and \( R_{p1.0} \) at plastic strains of 0.2 and 1.0%. The deformation speed was 0.10 mm min\(^{-1}\) corresponding to a strain rate of approximately \( 4 \times 10^{-4} \) s\(^{-1}\). The experimental scatter of the \( R_{p1.0} \) data is less than that of \( R_{p0.2} \). The main purpose of evaluating \( R_{p1.0} \) has therefore been to secure a safer curve drawing of the dependence of \( R_{p0.2} \) on coiling or tempering temperature.

Optical microscopy was carried out on all the samples after simulated processing in the deformation dilatometer. Grinding and polishing to 0.25 mm diamond paste was followed by etching in 2% nital solution. Thin foils were made from a selection of the samples by electropolishing and these were examined in a Jeol 2100F microscope.

4. Results and Discussion

4.1. Transformation Characteristics

In order to determine the window for transformation of austenite to bainite in the steels, dilatation experiments were carried out for cooling rates in the range 5–100°C s\(^{-1}\). The start and finish temperatures were determined for the steels and are displayed as curves in temperature–time graphs, Figs. 1 and 2. By comparing the transformation curves of the reference steel, C2, CrMo, with the three CrMoV steels, A5, E1 and F1, it can be seen that they are all similar and that the observed variations fall within the experimental scatter, Fig. 1a. The higher observed finish temperature of the reference steel at 100°C s\(^{-1}\) appears real, but these rates are outside practical interest and furthermore the transformation products are here martensitic.

Without molybdenum, steel B1 CrV, the transformation temperature shifts clearly to higher temperatures and polygonal ferrite is observed for cooling rates up to at least 30°C s\(^{-1}\), Fig. 1b. Similarly, reduction of the manganese content to 1.0%, in steel D1, lowers the bainite hardenability, although not to the same extent, Fig. 1b. Hence, it can be concluded that the base composition of 1.4% Mn, 1.0% Cr and 0.25% Mo produces sufficient bainite hardenability at cooling rates of about 30°C s\(^{-1}\).

Comparison between the results of CCT curves for steels A5 and F1 shows, furthermore, that the bainite hardenability can be retained by compensating a reduction of the molybdenum content from 0.25 to 0.12% by an increase of the Cr content from 1.0 to 1.25%, Figs. 1a and 2. The start and finish temperatures for two 2% Cr steels, G1 and G2, are inserted in Fig. 2 and, as expected, the transformation temperatures are considerably lower for those. Note also in Fig. 2 the inserted transformation temperature of steel G2 cooled at 20°C s\(^{-1}\). This cooling rate corresponds to the processing of a 12 mm hot strip. As expected the bainite transformation then starts at a higher temperature, increased by around 40°C.

Optical microscopy showed that bainite is formed in all steels excepting B1 and D1 for cooling rates above 10°C s\(^{-1}\). Examples of steels F1 and C2 with different cooling conditions are shown in Fig. 3. In optical micrographs it is not possible to distinguish any difference between the bainitic structures in specimens cooled at 400°C and those directly cooled to room temperature although the mechanical properties differ as discussed below. For the slowest cooling rate of 5°C s\(^{-1}\) some pro-eutectoid ferrite is visible in the microstructures. At the highest rates, the content of martensite increases and at 100°C s\(^{-1}\) the microstructures are practically fully martensitic.

Transmission electron microscopy (TEM) showed that the virgin bainite formed directly on cooling has a very dense dislocation structure, Fig. 4(a). The substructure consists of laths with mainly low angle boundaries. Average slip distances between the lath boundaries were estimated to lie between 0.5 and 1 μm. Similar lath structures were also seen in the bainite after the simulated cooling treatments. Many dislocations existed inside the laths and their
densities were so high that they could not meaningfully be quantified from the images. This was the case for both the directly cooled and the simulated coiled samples.

Recordings of electron diffraction from these structures in the steel A5 with simulated coiling at 450°C showed faint spots that are compatible with V(C, N) in addition to those from the ferrite matrix. When dark field imaging was carried out using these extra spots (Fig. 4(c)), there appeared to be small precipitates within the dislocation substructure.
4.2. Strength of the Bainitic Steels

In Figs. 5a and 5b the yield strengths, \( R_{p0.2} \), for all the V-microalloyed hot strip simulated steels are displayed for coiling temperatures varying between 400 and 600°C, together with the corresponding figures for the reference steel, C2, without vanadium. To avoid unnecessary repetition in the following, the bainitic hardenabilities of C2, A5, E1, F1 and H1 are denoted as standard. The results for steels B1 and D1, without Mo and with a lower Mn content of 1% respectively, have been left out since they have inadequate bainite hardenability, see the above Sec. 4.1. In line with this, their strengths were much inferior to those of all the other steels, with values of \( R_{p0.2} \) of about 600 MPa for coiling temperatures between 450°C and 550°C.

In order to assist in a correct analysis and interpretation of the results of the hot strip simulated steels, their strength levels have also been evaluated as:

1. Fast cooled at a rate of 30°C s\(^{-1}\) to room temperature (RT),
2. Fast cooled at 30°C s\(^{-1}\) to RT and then tempered at temperatures between 350°C and 600°C.

The results of 1 are inserted in Figs. 5a and 5b while those of 2 are shown in Fig. 7.

Technically, the most important result, revealed in Figs. 5a and 5b, is that V-microalloying produces higher yield strengths than the reference steel C2 for coiling temperatures in the range 400–550°C. At 400°C the effect is significant with strength levels between 740 MPa and 790 MPa for the Cr–Mo–V steels having standard hardenability as compared to the reference steel with yield strength of 680 MPa. For the vanadium steels with 2% Cr, G1 and G2, having enhanced bainite hardenability, the yield strengths are still higher at 840 and 880 MPa, respectively. For the steels with standard hardenability, A5, E1, F1 and H1, this enhancement of strength over the reference steel C2 must be attributed to V-microalloying. It is noticeable that this effect diminishes with increasing coiling temperature and more so for the high nitrogen (\( \approx 200 \) ppm) steels. In fact, for steel A5 with \( \approx 100 \) ppm nitrogen the effect is rather marginal and this steel maintains its strength best with increasing coiling temperatures, Fig. 5a. The reason for this difference is unclear. The higher levels of yield stress for the 2% Cr steels can be attributed primarily to their lower bainite transformation temperatures.

During coiling the bainite is tempered and its dense dislocation structure will recover, causing softening, unless hampered by existing carbides or nitrides in the bainite or by precipitates formed during coiling. For all steels with standard hardenability the bainite transformation is practically completed at 400°C. Hence, the initial bainite at coiling is the same as the bainite formed in steels directly cooled at 30°C s\(^{-1}\) to room temperature. Accordingly, the difference between the yield strength of a steel directly quenched to room temperature and the yield strength of the same cooled steel is a direct measure of the temper softening during coiling. From Fig. 5a we can deduce that the reference steel without vanadium, C2, undergoes a substantial softening from 780 to 680 MPa. For the V-microalloyed steels, on the other hand, softening is prevented. In A5 and E1 (high N), the strength levels of 760 MPa and 780 MPa are retained and even a slight strengthening from 750 to 790 MPa is obtained in the case of steel F1 (high N).

Figure 6 shows how a reduction of the cooling rate from 30 to 20°C s\(^{-1}\) after finish rolling affects the yield strength for coiling temperatures in the range of 400–500°C. This change in cooling rate corresponds to an increase of strip thickness from 8 to 12 mm for typical strip mill conditions. The yield stress levels for 12 mm strips are still impressive, 780 MPa at a coiling temperature of 400°C and 768 MPa at 450°C, but are, as expected, lower than for 8 mm strips. This decline in strength can be attributed to the higher bainite transformation temperature, by about 40°C, for the lower cooling rate, Fig. 2. Furthermore, the loss of strength is less sensitive to the cooling temperature. This is believed to be a consequence of less recovery of the somewhat coarser bainite formed during cooling, having a lower dislo-
cation density due to the higher transformation temperature.

These results show clearly that vanadium, especially in combination with nitrogen, is capable of eliminating the softening during cooling that occurs in steels without those elements. In the present case this resistance to softening may result from precipitation strengthening or be due to prevention of recovery of the bainite dislocation structure, or a combination of the two. In both cases this is most probably caused by V(C, N) precipitation.

A comparison of the observed yield strengths for hot strip simulated steels with quenched and tempered steels as well as their variations with coiling and tempering temperatures is instructive, Figs. 5a, 5b and Fig. 7. Firstly, comparing the strengths of all steels for the two types of treatments at a temperature of 400°C we find that they are just the same within the experimental scatter. This is logical and an expected result because at 400°C the bainite transformation is completed and the response from V(C, N) precipitation on the recovery of the dislocation structure and on particle strengthening should be similar during tempering and during cooling. At higher temperatures, however, the tempered steels retain their strengths generally better than the cooled steels. The probable explanation for this difference is that successively larger portions of the bainite had then formed at higher temperature in the cooled steels, and these were accordingly softer in their initial state.

It is possible that the aluminium content of the steel could affect the amount of free nitrogen at the end of finish cooling prior to coiling, resulting from more or less of the nitrogen being tied up as AlN at the reheating temperature of 1150°C. In turn, this could affect the precipitation of V(C, N) and accordingly the recovery of the bainite dislocation structure during cooling. The steels F1 and H1 are chemically the same except that the former is very low in aluminium, 0.003% Al, whereas the latter has a more normal content, 0.023% Al, and are therefore suited to examine this possibility. Figure 5b shows that the yield strength is somewhat higher for the F1 steel below a coiling temperature of about 500°C, thus indicating a stronger reduction of the recovery in the absence of aluminium.

4.3. Principles of Bainite Strengthening in Hot Strip Steels

Low carbon bainite provides high strength to steel by a combination of strengthening mechanisms:

1. A fine ferrite lath width.
2. A high dislocation density resulting from the displacement taking place during the bainite transformation.
3. Ferrite matrix strength, including solid solution hardening by the alloying elements.

For the present low levels of carbon and nitrogen it is unlikely that sufficient fine iron carbides and nitrides are formed in the quenched condition to produce significant strengthening.

According to Pickering⁵¹ tensile strengths up to 980 MPa can be generated in low carbon bainitic steels rapidly cooled to room temperature. The decisive parameter for the strength of bainite is its transformation temperature. Pickering found for a series of low C bainitic steels that the tensile strength increased linearly at rate of 3 MPa for every 1 degree C reduction of the 50% transformation temperature in a temperature range from 650 to 430°C. The most commonly used alloying elements for enhancing hardenability, manganese, chromium, and molybdenum, are all very effective in depressing the bainite transformation temperature and do so at approximately the same rate.⁵¹ For reasons of cost, the first two of those will be the most attractive choice.

For the present steels, yield stresses of between 750 MPa and 780 MPa were achieved in the as quenched condition, at 30°C/s, with a mean value of 770 MPa, Figs. 5a and 5b. Pickering and Gladman⁵⁵ found in their studies on low carbon bainites that the matrix strength including solid solution strengthening from normal alloying contents amounts to about 170 MPa for these types of steels. In order to estimate the strength contribution of the fine bainitic ferrite with a mean slip distance in the ferrite of about 1 μm (cf. previous section), it is necessary to know the Petch–Hall factor for ferrite having a very high dislocation density, as in the present case. It is well established that the grain size dependence of the lower yield stress is far larger than that of the flow stress after completion of the Lüders elongation. The Petch–Hall factor for the upper yield point, the lower yield stress and the flow stress during homogeneous plastic deformation represent a series of decreasing values; typical values for these three stages are: 1–1.5, 0.6–0.8 and ~0, 1 MPa m⁰.⁵¹ Lagneborg et al.¹⁰¹,¹⁰² have analysed and modelled this feature further. They described quantitatively how the Petch–Hall factor for a 0.03% C mild steel decreased through these three stages and reached a value of 0.19 MPa m¹/² in the region of homogeneous deformation that remained constant with increasing strain. This value is considered to be relevant for estimating the strength contribution of the fine bainitic ferrite with its very high dislocation density, Fig. 4. Hence, with the grain size of 1 μm this gives a strength contribution of 190 MPa. It may be noted that Pickering and Gladman used a Petch–Hall factor of 0.54 MPa m¹/² in their analysis of the strength of low carbon bainite. However, this is a value typical for the lower yield stress and should not be applied to a highly dislocated ferrite as in the present case.

Strain hardening is the remaining strengthening contribution and can now be obtained by subtracting the other two components from the total strength. The resulting contribution to the yield stress is 410 MPa which corresponds to a dislocation density (ρ) of 1.2×10¹⁵ m⁻², as calculated from the standard expression

\[
\sigma = \frac{2G\rho}{1 + 2\nu}
\]

where \(G\) is the shear modulus, \(\rho\) the dislocation density and \(\nu\) the Poisson's ratio.
The question of how much precipitation strengthening the V(C, N) particles can contribute to the total strength is not straightforward. This is because the precipitates and the dislocation structure represent two different populations of point obstacles that a gliding dislocation has to bypass as it moves in its slip plane. In such a case, the two strengthening contributions, as they are known when operating alone, are not simply additive. Koppenaal and Kuhlmann-Wilsdorf showed already in 1964 that the strength of two different dispersions of point obstacles is given by

$$\sigma = \alpha Gb \rho^{1/2} \quad \text{................................(1)}$$

with $\alpha = 0.6$, $G = 80,000$ MPa, $b = 2.5 \times 10^{-10}$ m.

Although the transmission electron microscopy reported above did not lend itself to any meaningful measurements of the dislocation density, a judgement, based on the authors' previous experiences of such measurements, is that the observed dislocation structure for the present steels in the quenched condition indicates a density clearly above $5 \times 10^{14}$ m$^{-2}$. We therefore think that the estimated strain hardening of 410 MPa is consistent with the observed dislocation structures. Hence, in summary the strength of the quenched steels is composed of the following estimated strength components:

1. **Matrix strength**, including solid solution strengthening of alloying elements: ~170 MPa.
2. **Fine bainitic ferrite**, mean slip distance: 1 $\mu$m: about 190 MPa.
3. **Strain hardening from the dislocation substructure**: about 410 MPa.

The prolonged holding during coiling in the hot strip process is expected to produce a fine V(C, N) precipitation in the bainitic ferrite. These precipitates will have two important effects on the final strength of the hot strip:

- Strength will be added to the finished steel by precipitation hardening.
- The fine precipitates will retard and may even prevent the recovery of the dislocation substructure of bainitic ferrite during coiling, thereby effectively retaining the strength of the virgin bainite.

Table 2. Demonstration of how precipitation strengthening compensates for reductions of an original strain hardening of 410 MPa caused by recovery.

| Strain hardening, MPa | Precipitation strengthening, MPa | Resulting strength, MPa \
<table>
<thead>
<tr>
<th></th>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>operating alone</td>
<td>operating alone</td>
<td>according to Eq. (2)</td>
</tr>
<tr>
<td>410</td>
<td>0</td>
<td>410</td>
</tr>
<tr>
<td>320</td>
<td>256</td>
<td>410</td>
</tr>
<tr>
<td>320</td>
<td>90</td>
<td>332</td>
</tr>
</tbody>
</table>

For the combinations of other strengthening mechanisms—such as solid solution hardening + particle strengthening, or fine grain strengthening + solid solution hardening or particle strengthening—the situation is simpler. In these cases a plain additive rule appears to be a good approximation and can easily be justified. In the first case the particles are normally at least two orders of magnitude more widely spread than the dissolved atoms and hence solid solution hardening can be regarded as a friction stress acting uniformly over the gliding dislocation as opposed to the point forces of the particles acting on the dislocation. Clearly, under these circumstances the friction stress plus the stress needed to move the dislocation past the particles in a frictionless slip plane must be overcome for slip to occur in the real material. A similar reasoning can be applied for the combination of fine grain strengthening + solid solution hardening or particle strengthening, as is commonly done in the case of ferrite–pearlite HSLA steels.

The reduction in yield stress of 90 MPa for the reference steel C2, from a level of 780 MPa when quenched to room temperature to about 690 MPa when quenched at 400°C and 450°C, Fig. 5a, can be taken as a measure of the recovery of the dislocation structure during cooling in the absence of V–N microalloying. The implication of Eq. (2) applied on the combination of strain hardening and precipitation strengthening is that much more precipitation strengthening is necessary to compensate for an observed loss of strain hardening by recovery than the actual number of the loss. This effect is illustrated for the case above in Table 2. As can be seen, it requires a very large precipitation strengthening, 256 MPa, to fully compensate for the observed loss of strain hardening of 90 MPa. Similarly a pure precipitation strengthening of 90 MPa would produce only a marginal increase of strength, from 320 to 332 MPa in the presence of the dislocation substructure.

Hence, to fully compensate the loss of strain hardening of 90 MPa would require a precipitation strengthening of 256 MPa when operating alone, Table 2. This level of precipitation strengthening is, however, more than can be expected in the present case. From previous work a maximum precipitation strengthening of about 90 MPa can be estimated in a ferrite–pearlite steel aged at 650°C and with the standard levels of carbon, vanadium used in the present case, 0.04% C, 0.08% V and 0.010% N. Below about 600°C the precipitation strengthening has been found to fall due to incomplete precipitation and is therefore expected to be lower than 90 MPa at the lower end of the present cooling temperature range. However, even a high strengthening of 90 MPa will add only 12 MPa to the reduced strain hardening of 320 MPa, as revealed in Table 2. In turn, this corresponds to a decrease of the yield stress of the virgin bainite (quenched steel) of 78 MPa (410–332 MPa). As opposed to this, steel A5 fully retains its strength after cooling at 400°C at a level of 760 MPa, Fig. 5a. Accordingly, it must be concluded that the V-microalloyed steels maintain their strength during cooling primarily by an effective retardation of dislocation recovery rather
than by precipitation strengthening.

The present analysis provides important clues for the development of hot strip bainitic steels:

1. The base alloying, manganese, chromium and molybdenum, should provide the lowest possible bainite transformation temperature that is compatible with the lowest technically feasible cooling temperature while, of course, exceeding the martensite start temperature. This should result in the highest strength of the original bainite. It is to be noted that it serves no purpose to lower the 50% transformation temperature much below the lowest cooling temperature because, then, more than 50% of the bainite would anyway form at the cooling temperature.

2. It is crucial to prevent the recovery of the dislocation structure of the bainite as effectively as possible so as to retain as much as possible of the strength of the virgin bainite. The primary means to accomplish this is generation of a fine dispersion of precipitates during cooling that anchors the dislocation network. Vanadium–nitrogen microalloying, leading to a fine V(C,N) precipitation during cooling, appears to be the most potent way to achieve this. A low cooling temperature (cf. point 1) will also assist in reducing the recovery.

3. In these bainitic steels, having a high dislocation hardening, neither precipitation strengthening nor grain size strengthening can be as potent as in ferrite–pearlite steels.

In summary, it can be stated that the essential effect of V–N microalloying is to effectively prevent the recovery of the structure of the bainite through precipitation of V(C,N) on dislocations and that the retained strength of the bainite after cooling is primarily due to this effect. Precipitation strengthening by the fine V(C,N) only makes a much smaller contribution to the observed strength.

5. Conclusions

(1) Generating bainitic microstructures during the rolling process of hot strip steels is a potent method to reach yield strength levels above 600 MPa for low carbon hot strip steels. A base alloying of 1.4% Mn, 1.0% Cr and 0.25% Mo is required to form a fully bainitic structure during the cooling after finish rolling of a 8 mm 0.04% C hot strip steel. It is also demonstrated that the bainite hardness can be maintained by compensating a reduction of the molybdenum content by an adequate addition of chromium. With these compositions the bainite transformation is virtually complete at around 400°C. Thus, when cooling at 400°C the bainite transformation is completed during the prior cooling whereas with higher cooling temperatures an increasing fraction of the bainite forms after cooling.

(2) With vanadium microalloying, the strength of the virgin bainite of the present steels is effectively retained during cooling, in some cases even slightly raised, whereas a substantial softening occurs in the reference steel without vanadium. With a base alloying of 1.4% Mn, 1.0% Cr and 0.25% Mo or 1.4% Mn, 1.4% Cr and 0.12% Mo together with 0.08% V and 0.010–0.020% N the yield strength for simulated 8 mm strip lies the range of 740–790 MPa after cooling at 400°C, whereas it is 680 MPa for a similar steel without vanadium. By raising the chromium content to 2%, yield strengths in the range of 840–880 MPa can be reached. The enhanced hardenability of the 2% Cr steels allows also yield stresses up to 780 MPa to be achieved at lower cooling rates, corresponding to thicker strips up to 12 mm.

(3) The principles of bainite strengthening in hot strip steels have been outlined. The added strength of low carbon bainite is derived from a fine microstructure with lath widths below about 1 µm and a dense dislocation structure in the bainitic ferrite. Both strength contributions are enhanced by reduction in the bainite transformation temperature due to a finer lath size and denser dislocation structure. During the prolonged holding at elevated temperatures after cooling, the dislocation structures recover and lose strength. Precipitation of V(C,N) during cooling alleviates the effects of this annealing, both by preventing the dislocation recovery and by precipitation strengthening. However, these contributions are not simply additive and an analysis shows that the loss of strength cannot be compensated by predicted levels of precipitation strengthening when strain hardening and precipitation strengthening are operating jointly. Accordingly, a main conclusion of this work is that the vanadium microalloyed bainitic steels retain their strength primarily as a result of retardation of the recovery of dislocation structure by fine V(C,N) precipitates. Precipitation strengthening contributes only to a lesser degree.

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