Microtexture of Thin Gauge Hot Rolled Steel Strip

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A series of trials were conducted on a compact rolling mill to evaluate the properties and characteristics of carbon steel strip hot rolled to less than 2 mm in thickness from thin cast slabs. Steels of two different compositions were studied, the first one, a low carbon steel, was rolled to a total reduction ranging from 94 to 98%, final thickness ranging from 1.06 to 2.69 mm, whereas the second one was a Nb bearing microalloyed steel rolled to a total reduction of around 96%. The rolling trials were complemented by means of computer modelling to get a deeper understanding of the process. It was found that the ferritic grain size of the low carbon strips varied from 7 to 10 μm, with the finer sizes found in the thinner strips, the grain size of the microalloyed steel was found to be 3.6 μm. Analysis of the texture of the hot rolled strips indicated that the ferrite in the low carbon resulted from the transformation of recrystallized austenite, in comparison, low intensity transformation texture from unrecrystallized austenite was found in the Nb bearing steel. The observed texture data correlate with the R-values measured.

KEY WORDS: hot rolling; steel; texture; microstructure.

1. Introduction

The standard processing route for steel strips thinner than 2 mm with narrow dimensional tolerances and high surface quality includes, as a rule, hot rolling, pickling, cold rolling and annealing. The main functional properties for cold rolled and annealed steel strips are their strength, toughness and formability. The latter is often characterized by the deep drawability of the material, which strongly depends on the texture of the steel sheet. The texture of cold rolled and annealed strips usually displays two main texture components: the {111} fibre (ND fibres) that has a beneficial effect on the deep drawability and the {001} texture fibre, which has a detrimental effect.1)

In some cases when the requirements for high surface quality are not very strong, hot rolled thin strips can be a viable economical alternative for cold rolled steel strips. This replacement will be more effective when the mechanical properties of the hot rolled thin strips, i.e. strength and toughness, are high enough and are combined with low values of plastic anisotropy. Such a combination of properties may be achieved by means of thermo-mechanical treating that results in a fine-grained microstructure with a specific texture. Significant grain refinement is possible to obtain by lowering the transformation temperature in the steel, for instance by the use of accelerated cooling after hot rolling, controlled rolling in combination with microalloying or by solid solution strengthening. A widely accepted opinion is that the development of a {332}\{113\} transformation texture component in hot rolled sheets enhances both toughness and formability in steel strips,2) whereas the {113}\{110\} texture component has detrimental effects on both characteristics. A number of studies has proven that the {332}\{113\} transformation texture component sharpens when the substitutional solutes, like Mn and Ni, are added to the steel together with Nb,2,3) when the cooling rate increases5–8) and when the austenite grain size decreases.3)

Improvements in thin slab casting, hot direct charging together with hot rolling techniques and practices have resulted in the production at an industrial scale, of hot rolled strips of gauges thinner than 2 mm. Such processing implies that the reductions of thickness in the originally cast slab is higher that 96%. Furthermore, hot direct charging of thin cast slab does not permit the common isotropic transformations that take place in conventionally processed material.6)

The goal of the present work is to study the influence of processing parameters on the microstructure and texture of two carbon steels, one plain and the other with Nb addition, which were cast into thin slabs and hot rolled in a compact strip mill to different final thickness.

2. Experimental

Two low carbon steels, designated as A (plain carbon) and B (with higher Mn and Nb contents), were hot rolled in an industrial six-stand continuous rolling mill to different final thickness. All the samples were hot rolled directly after being cast into slabs of around 50 mm in thickness. Figure 1 shows a schematic diagram of the mill from which the samples were taken. Full details of the mill can be found elsewhere.6) Three different rolling schedules were...
used with steel of composition A (identified as A1, A2 and A3), whereas only one schedule, similar to that of A2, was applied with steel B. All the samples from either type of materials were cut at room temperature from the coiled product. Coiling of the strips take place at the end of the run out table, which is equipped with a cooling system. The cooling system consists of a series of low pressure laminar water headers located on the top of the strip and low pressure water jet headers from the bottom side. The number of headers being used was controlled by the operators to assure cooling temperature below 650°C. The chemical composition of the steels is shown in Table 1 and the parameters of the rolling schedules together with the measured $R$ and $\Delta r$-values are shown in Table 2. The value of critical temperatures $A_1=718^\circ$C and $A_3=860^\circ$C were calculated with the equations proposed by Andrews. The temperature changes in the strips during rolling were predicted by computer modelling, as this approach allows to have an idea whether the final rolling temperature (FRT) stays within the austenite phase field or within the two phase, $\alpha+\gamma$, region. The values shown in Table 2 indicate that the final rolling temperatures predicted by the model were above $A_3$, without entering the intercritical region for the rolling schedules A1, A2 and B, but not for the thinnest sample (A3), on which the final pass may have been imparted in the intercritical region.

Evaluation of the forming characteristics of the hot rolled steels was carried out by testing tensile samples cut parallel, perpendicular and at 45° with respect of the rolling direction. Individual measurements of the instantaneous width and thickness to the samples were made to obtain the strain values as referred to the width and thickness ($\varepsilon_w$ and $\varepsilon_t$ respectively) of the samples. The plastic strain ratio $r$ was determined by:

$$ r = \frac{\varepsilon_w}{\varepsilon_t} \quad \text{(1)} $$

Following the procedure described in a previously published work.

The average value of $r$ was determined from the data collected by testing samples cut along the different directions:

$$ R = \frac{r_0 + r_90 + 2r_{45}}{4} \quad \text{(2)} $$

where $R$ represents the average value of $r$ and the sub-indexes 0, 90 and 45 indicate the angle with respect to the rolling direction. The plastic anisotropy of the sheet ($\Delta r$) was calculated by:

$$ \Delta r = \frac{r_0 + r_90 - 2r_{45}}{4} \quad \text{(3)} $$

The full details of the experimental procedure can be found
elsewhere.\textsuperscript{10) The microstructure of the samples was examined in the cross section of the samples after standard sample preparation procedure and 2\% nital etching. The plane for observation was perpendicular to the transverse direction as it is schematically shown in Fig. 2. A, B, C and D indicate the position of points located in the centre of the cross section of the strip (A), the centre of the upper surface (B), the centre of the side wall (C) and the corner (D) (cf. Fig. 2). In the centre of the plane that includes points A and B a local microtexture measurements by means of orientation imaging microscopy (OIM) were performed. The samples for OIM were electrolytically polished after a final step of mechanical polishing with 1\(\mu\)m diamond paste and etched in 2\% nital. The OIM attachment was installed on a Philips XL30 ESEM with an LaB\(_6\) filament and the electron backscattering diffraction (EBSD) patterns were acquired and analysed by means of the commercial TSL OIM\textsuperscript{*} software.\textsuperscript{11) The orientation data of at least 3 local measurements of each sample were summarized and further post processed by means of FHM-MTM software developed by Van Houte\textsuperscript{12)} in order to represent the texture of the strips by means of the orientation distribution functions (ODFs). Every ODF created in this way represents statistically reliable texture data.

3. Results

The temperature evolution in different zones of the hot rolled strips predicted by the computer model\textsuperscript{8)} is shown in Fig. 3. The temperature at the end of rolling was measured in point B and it is marked on the diagram. The average grain size of austenite, represented by the average grain diameter, was also modeled. The predictions for the austenite grain size, together with the ferrite grain diameters that were measured in the hot rolled thin strips, are shown in Fig. 4.

The microstructure of hot rolled strips obtained by means of optical metallography (OM) in the middle thickness of the strips is shown in Figs. 5(a)–5(d). The average ferrite grain diameters calculated from the EBSD data are shown in Fig. 6(a), whereas Fig. 6(b) displays the grain diameter in the central zone and in the surface of the strip A2. The definition of grains in the OIM post processing software differs from the conventional one commonly used in metallography and needs further explanation. The determination of grains by OIM is based on an algorithm that groups neighboring similarly oriented points into grains. The software identifies the orientation of every point, and checks whether such a particular orientation falls within the tolerance allowed for by the operator to decide whether that point, together with neighboring points is a part of one single grain. The minimum number of points required to decide if a group of closely oriented points should be considered as a grain (minimum grain size) is defined by the operator, as well as the minimum misorientation angle that determines if the measured neighboring points belong to the same grain or not. The procedure is repeated until the boundaries between connected points can be established. As a consequence of this method it is possible to observe one single grain with a large intragranular orientation gradient accumulated over a certain distance by low point-to-point misorientations. This approach allows a more precise determination of the grains with different level of misorientation than the classical metallographic methods, but the re-

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Results are strongly dependant on the definitions, ranges and values set by the operator. The grains data in this work were obtained using a grain tolerance angle of 5° and the minimum grain size was chosen to be 2 measuring points, which were at a distance of 1 \( \mu m \) apart. These parameters are commonly used in this type of analysis. All data points with a confidence index (CI) lower than 0.05 were excluded from the analysis as dubious. The CI quantifies the reliability of the indexed pattern. The data displayed on Figs. 6(a), 6(b) are taken from one representative measurement of every sample, which includes more than 1000 grains. Afterwards, the average grain diameter for each sample was calculated from at least 3 to 6 different measurements and the averaged data for every strip were plotted vs. the total strain together with the predicted one by the model austenite grain size (cf. Fig. 7). The average grain size calculated by means of the linear interception method was presented also in this figure (white dots) and it coincides very well with the data derived from the OIM measurement.

Figures 8(a)–8(d) display the ODF in the \( \phi_2=45^\circ \) section of Euler space (Bunge notation) measured in the middle thickness of each strip and in Figs. 9(a) and 9(b) some most important ferrite texture components are represented in \( \phi_2=45^\circ \) section of the Euler space (Fig. 9(a)) together with the \( \gamma \)-phase orientations from which they might emerge in accordance with the Kurdjumov–Sachs orienta-

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Fig. 4. Evolution of the average austenite grains in different zones of the strip predicted by the model of Zambrano et al. and measured average grain size of the ferrite at room temperature (white dots).

Fig. 5. Microstructure of the middle thickness of the strips; (a) steel B, \( t=2.18 \text{ mm} \); (b) steel A3, \( t=1.06 \text{ mm} \); (c) steel A2, \( t=1.92 \text{ mm} \); (d) steel A1, \( t=2.69 \text{ mm} \).
The ODF of steel B (with 0.023% Nb), Fig. 8(a), shows a maximum of $2/H_{1100}$, which represents the random texture level, at $\{001\}/H_{110}$, which corresponds to the rotated cube orientation and it is a characteristic ferrite component transforming from a recrystallized cube austenite orientation. The texture components $\{113\}/H_{110}$ with an intensity of $2.5/H_{110}$ and $\{332\}/H_{110}$ with $3/H_{110}$ are also present in the strips of steel B, and are associated with the transformation of ferrite

Fig. 6. (a) Grain size distribution in terms of area fraction for steels B ($D_{av}=3.6 \mu m$), A3 ($D_{av}=7.2 \mu m$), A2 ($D_{av}=9 \mu m$) and A1 ($D_{av}=10 \mu m$); (b) grain diameter measured in the center of the strip and in the edge zone (strip A2 $D_{av}$ center $=9 \mu m$, $D_{av}$ edge $=7 \mu m$); (c) number of grains and area fraction of the grains plotted vs. grain diameter for steel B.

Fig. 7. Changes in average austenite grain diameter $D_{av}$, average ferrite grain diameter $D_F$ and $D_{av}/D_F$ ratio as a function of the rolling reduction in the thin strip from the plain carbon steel. The white dots display the data for the average ferrite diameter measured by the linear intersection method.

Fig. 8. Middle thickness texture of the strips; (a) B, $t=2.18 \ mm$; (b) A3, $t=1.06 \ mm$; (c) A2, $t=1.92 \ mm$; (d) A1, $t=2.69 \ mm$. Iso-intensity levels 0.8–1.0–1.3–1.6–2.0–2.5–3.2–4.0–5.0–6.4.
from deformed austenite, which is assumed to display a characteristic β-fibre texture of the deformed FCC metal. The textures of the strips from steel A were found to vary with the processing parameters. Figure 8(b) displays the texture of the sample A3 with a thickness of 1.06 mm. The maximum intensity, 8×, is observed at the {001}〈110〉 rotated cube component and 2× at the component with Euler angles \( \varphi_1 = 10^\circ \) \( \Phi = 80^\circ \) and \( \varphi_2 = 45^\circ \), which differs approximately \( \sim 10^\circ \) from the ideal position of the rotated Goss (〈110〉〈111〉) component. Rotated cube, Goss and rotated Goss are characteristic texture components of the ferrite transformed from cube austenite component (cf. Fig. 9(b)). The {332}〈113〉 and {558}〈311.5〉 transformation components emerging from the deformed austenite are also presented in the ODF but with low intensity (1.6×) though.

Steel A2 with a final thickness of 1.9 mm, which was finished at 885°C displays almost the same type of texture as the sample A3, i.e. the typical texture of ferrite transformed from a cubic austenite component (Fig. 8(c)). The texture is characterized by a strong rotated cube {001}〈110〉 (6.4×) combined with a Goss and a rotated Goss component with intensities of 2.5× and 3.2× respectively. The relatively weak fibre texture component with maximums of 1.6× on {332}〈113〉 and {558}〈311.5〉 can be identified as transformation products from unrecrystallized austenite.

The texture of the strip A1 (Fig. 8(d)) with a thickness of 2.69 mm is the weakest one among the plain carbon steel sheets. It displays a maximum of 2.5× on RD fibre {113 to 114}〈110〉, which spreads towards the {111}〈112〉 component forming a continuous series of orientations which constitutes the ferrite β-fibre. The ferrite β-fibre is the transformation product from the austenite β-fibre, which represents the characteristic texture components of the deformed FCC material with high stacking fault energy (SFE). A rotated cube recrystallization texture component {001}〈110〉 is also presented in the ODF with the intensity of 2×.

4. Discussion

4.1. Microstructure

The grain size is an important structural parameter, which strongly influences the toughness of the steels as well as their forming characteristics, and therefore it was used for the quantitative characterization of the microstructure. The average grain diameter \( D_{av} \), calculated on the basis of EBSD measurements is used to define the grain size. This parameter changes significantly as a function of the chemical composition of the steel and the processing parameters. A general observation is that the measured ferritic grain diameter corresponds sufficiently well to the predicted austenite grain diameter on the final rolling stage (cf. Fig. 4) and the microstructures in the cross section of the hot rolled strips consist of homogeneously distributed ferritic grains, i.e. no differences between the microstructure and grain size in the middle thickness and in the surface layers of the strips were observed (cf. Fig. 6(b)).

Strong grain refinement is observed in steel B, which contains 0.023%Nb. The average grain diameter \( D_{av} \) measured in this steel (Fig. 6(a)) is 3.6 μm. It is 2.5 times smaller than the average grain diameter measured in strips A1, A2 and A3 (compare the data in Fig. 6(a)). Most of the microstructure of steel B consist of grains smaller than 5 μm, which include between 75 and 95% of all number of grains (see Fig. 6(c)), but cover only 30% of the area.

OIM allows to discriminate grain orientations with regard to the grain size. Such an exercise was made with the microstructure of steel B. Figure 10 shows the microstructures for the grains that have a size smaller than 3 μm, Fig. 10(a), where it can be noticed that the small grains show the tendency to be distributed in chains, which may correspond to some of the previous austenite grain boundaries, or are concentrated in regions with elongated shapes along the rolling direction, which may correspond to previous austenite grains. The grains that have a diameter higher than 10 μm are represented in Fig. 10(b). It may be possible that such large grains of ferrite emerged from highly strained austenite, and were the first to be transformed. Hence, after transformation, these grains will have a growth advantage in comparison to the ones transformed later on. The grain size distribution observed in the final ferritic structure may bear a legacy to the retardation of recrystallization of the parent austenite phase, as a consequence of...
microalloying with Nb, as well as for finish rolling in the intercritical \(\alpha+\gamma\) region.\(^{13}\)

The ferrite grains in plain carbon steel are larger than those in the Nb containing steel B, their average diameter vary from 7.2 \(\mu\)m, for strip A3, to 10 \(\mu\)m, for strip A1. The metallographic observations did not show significant differences between the microstructure (grain size and shape) in the surface and in the middle thickness of the strips (Fig. 6(b)), but somehow, the grains close to the surface are slightly smaller in diameter than those at mid thickness. A relatively weak grain refining effect was observed in the plain carbon strips as a function of the rolling reduction, which can be associated with the increased dislocation density in the parent austenite phase prior to the transformation. An indirect proof of this assumption is the difference in the parent austenite phase prior to the transformation, which can be associated with the increased dislocation density before the transformation predicted by the model\(^{13}\) and \(D_f\) is the size of the ferrite. It can be considered that with increase of this ratio more ferrite grains (with a small diameter) are nucleated from a single austenite grain, assuming that there is no grain growth during coiling. Because coiling was executed at temperatures below 650°C, such assumption seems reasonable. Therefore, the number of ferrite grains emerging from a single austenite grain will be a function only of the deformed substructure of the austenite, and can be used as a qualitative character-

istic of the accumulated plastic deformation in the parent austenite prior to transformation. The data in Fig. 7 show that austenite grains of a given size produce ferrite grains of smaller size as the total accumulated strain increases, i.e. the \(D_A/D_f\) ratio increases with an increase of the rolling reduction.

### 4.2. Texture

The ODF of steel B, which was calculated on the basis of 5 different local measurements, is shown in Fig. 8(a) and displays two important components of ferrite that were generated from deformed austenite. The \(\{332\}\{113\}\) ferrite texture component, which is considered as a favorable orientation from the viewpoint of drawability\(^{12}\) (high \(R\) value) is presented with an intensity of \(2.5\times\). The \(\{112\}\{131\}\) component, which also represents a transformation product from deformed austenite, appears with a maximum intensity of \(3.2\times\). This component, together with the rotated cube that was observed with an intensity of \(2\times\), are considered to have a detrimental effect on the \(R\)-value.\(^{21}\)

The analysis of the distribution of the texture components among grains of different size was done by calculating the ODF from grains smaller than 3 \(\mu\)m and larger than 10 \(\mu\)m, as they can be clearly discriminated in steel B. The ODF plots shown in Figs. 10(c) and 10(d) correspond to the smaller and larger grains shown in Figs. 10(a) and 10(b), respectively. The transformation texture components of the deformed austenite \(\{332\}\{113\}\) (shifted to \(\{554\}\{225\}\)), \(\{111\}\{112\}\) displaying an intensity of \(6\times\) and \(\{001\}\{010\}\) with an intensity of \(4\times\) are the strongest components in the ODF of the grains larger than 10 \(\mu\)m, whereas the \(\{001\}\{110\}\) and \(\{110\}\{110\}\) components appear with intensities of \(3\times\), respectively (Fig. 10(d)). According to some authors\(^{2,13,15}\) these texture components could emerge from either deformed or recrystallized austenite (see Fig. 9(b)), but the later is less probable, taking into account the Nb content of steel B, the rolling parameters and the model predictions.

The ODF of the grains smaller then 3 \(\mu\)m (Fig. 10(c)) displays almost equal intensity of both ferrite texture components that could emerge from deformed austenite (\(\{223\}\{110\}\) and \(\{332\}\{113\}\)) with intensity of \(2-3\times\) and from recrystallized austenite (\(\{001\}\{110\}\) and \(\{110\}\{110\}\) with intensity of \(2-3\times\). Considering the data of the grain size diameter and how the texture components are distributed among them it can be concluded that the strong deformed grains are among the first to transform into ferrite and hence, they have a growth advantage with respect to later transformed grains that might have emerged from the partially recrystallized austenite.

The textures of plain carbon steel strips are shown in Figs. 8(b), 8(c) and 8(d). In general the textures of the strips A3 and A2 differ significantly from the texture of the Nb steel, whereas the texture of strip A1 looks similar to it. The intensity of the texture increases with the increase of the total rolling reduction and the thinnest strip A3 displays the strongest texture. Two main groups of texture components are presented in the ODFs of the plain carbon steel. The first group corresponds mainly to transformation products emerging from recrystallized austenite (i.e. rotated cube,
rotated Goss and Goss). The intensity of the rotated cube component is stronger in the thinner strips A3 and A2 displaying a value of 8 and 6.4× respectively and only 2× in the thickest strip A1. The rotated Goss and Goss components are concurrently present only in the texture of strip A2, whereas the rotated Goss component with a certain deviation from the exact position is observed also in the texture of the thinnest strip A3. The transformation components from deformed austenite are weak in the textures of the strips A3 and A2 but they are the strongest components in the texture of the strip A1 with a thickness of 2.69 mm, which displays the highest R-value. Hence, the improvement of the R-value in the thickest strip is due to the two main reasons: (i) a general decrease of the intensity of unfavorable components and (ii) the development of the \{111\}\{112\} texture component which has a positive effect on the deep drawability.\(^2\) For better understanding the texture formation in hot rolling of thin strips, it is necessary to know also the strain distribution among the different passes. Figure 11 displays the evolution in the reduction of the strips thickness presented by the true strain for every rolling pass. Strips A2 and A3 were subjected to higher reductions at elevated temperatures than strip A1 but in the final pass all strips received an almost identical rolling strain of 0.25. If the reduction at elevated temperatures is higher the intensity of the ferrite transformation texture from recrystallized austenite is stronger (cf. Figs. 11 and 8(b), 8(c), 8(d)), because the high temperature deformation intensifies the recrystallization of the austenite. Hence, taking into account the texture data in Figs. 8(b), 8(c) and 8(d) in Fig. 11 it is possible to assume that recrystallization in the thickest strip A3 develops weaker than in the thinnest strips, as far as it was subjected to the lower strain. In a later stage it will be possible to employ the present data to extend the \(\gamma\)-recrystallization model\(^3\) to a fully quantitative texture dependant description.

5. Conclusions

The microstructure and texture of thin strips produced by hot rolling of hot charged-thin cast slabs is studied in this work. Two types of steels were studied, plain carbon steel, subjected to total hot rolling reduction of 94 to 98%, and Nb containing steel finish rolled to a total reduction of 94%. It was found that the final grain size on hot rolled plain carbon steels ranged from 7 to 10 \(\mu\)m, was dependant on the total amount of reduction during hot rolling. Significant grain refining was observed to occur in the Nb containing steel in which the average grain size was 3.6 \(\mu\)m.

A low intensity transformation texture from deformed austenite was observed in the Nb containing steel, whereas the type and intensity of ferrite texture in the plain carbon steels depends on the thermo-mechanical history of the steel strip. When the reduction at elevated temperatures is higher the ferrite texture intensifies and possesses all typical transformation components from the recrystallized austenite phase. When the strain at elevated temperature is low the texture is weak and additional transformation components emerging from deformed austenite are dominant. The observed texture data of plain carbon steel are in good correlation with the results from R-value measurements.

The data confirm that by proper control of the processing parameters it is possible to obtain thin hot rolled strips with fine grains and appropriate texture providing satisfactory deep drawability.

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