Grain Growth after Intercritical Rolling

R. Petrov, L. Kestens, K. Verbeken and Y. Houbaert

Department of Metallurgy and Materials Science, Ghent University,
Technologiepark 903, 9052 Ghent, Belgium

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Abstract. The distribution of the characteristic texture components between the ferrite grains of different size classes has been studied in a steel with 0.082%C, 1.54% Mn, 0.35% Si, 0.055%Nb and 0.078%V after different rolling schedules with a final rolling temperature above or below A_{r3} . Microstructures and textures were characterized by means of optical microscopy and orientation microscopy. A strong grain refining effect together with a bimodal grain size distribution was observed in the steel both after final rolling in the intercritical region or in the austenite region, close to the A_{r3}^d temperature. The differences in grain size were interpreted on the basis of three potentially acting mechanisms: (i) transformation-induced recrystallization, (ii) increased mobility of specific grain boundaries and (iii) fast nucleation of ferrite grains on specific sites of the parent austenite microstructure. The experimental data clearly favoured the third of these assumptions as the responsible mechanism for the observed bimodal grain size distributions.

Introduction

In some cases the rolling of thin gauge strips can be finished in the intercritical $\gamma+\alpha$ region. This may produce an undesirable coarse-grained microstructure [1,2,3,4] and a texture gradient from the surface to the centre of the strips [2] that can give rise to bad mechanical properties. Vander-schueren et al. [2] observed this phenomenon in plain carbon steels, which were finished in the two-phase region. They explained this behaviour by the existence of austenite grains that transformed to virtually dislocation free ferrite grains, which act as nuclei for the subsequent recrystallization of the strained ferrite. The authors called this phenomenon *transformation induced recrystallization* and mentioned that it did not appear when Ti based precipitates interact with the recrystallization process. Bodin et al [3,4] gave the same explanation for the observed ferritic grain growth which gave rise to a bimodal grain size distribution after intercritical rolling of C-Mn and low carbon steels. In both cases, there were no direct observations of the local texture or specific crystallographic orientations of large and small grains separately.

On the other hand, a strong grain refining effect was observed by Bleck et al [5] after uniaxial compression of a C-Mn-Nb steel in the intercritical region followed by controlled cooling and by Petrov et al [6] after industrial rolling of C-Mn-Nb steel finished close to the A_{r3} temperature.

Although rolling in the two-phase $\alpha+\gamma$ region is of great practical importance for structure and texture formation of thin gauge hot rolled strips, the above examples show that there are contradictory data with regard to grain size and grain size distribution. A satisfactory explanation for the occurrence of a bimodal grain size distribution after intercritical rolling does not exist and there is a deficiency of experimental data on this subject in the scientific literature. The goal of the present work is to obtain a better understanding of the acting mechanism for the formation of the microstructure during intercritical rolling of a carbon – manganese steel, in order to create possibilities for further improvement of their properties (strength, toughness and deep drawability).

Experimental

Rough rolled blocks of which the chemical composition is listed in Table 1, and with dimensions 140x100x35 mm³ were hot rolled in a laboratory rolling mill to a thickness that varied from 4 to 2.75 mm according to the rolling schedules shown Fig.1. Four rolling schedules were designed on

Table 1 Chemical composition of the steel

Chemical composition [mass.%]									
C	Mn	P	S	Si	Al	Nb	V	N	
0.082	1.54	0.01	0.006	0.36	0.033	0.055	0.078	0.005	

the base of the experimentally deduced CCT diagram of the steel, which is reported elsewhere [7]. The first rolling schedule (named 1CS) consists of 3 rolling passes above the austenitic non recrystallization temperature (T_{nr}) and 3 rolling passes below the Ar_3 temperature, i.e. there is no rolling in the region between T_{nr} and Ar_3 , and thus the austenite will be completely recrystallized before the start of the γ/α transformation. The rolling schedule indicated as 2CS is a more classical one in which the final rolling temperature (FRT) is above the γ/α transformation temperature of strained austenite (Ar_3^d). These two rolling schedules were followed by a coiling simulation (CS) at 600°C in a computer-controlled furnace. It should be mentioned that the transformation temperature of the strained austenite Ar_3^d is higher than the γ/α transformation temperature Ar_3 of non-strained austenite because of the changes in the substructure (i.e. the energy) of the strained austenite [7].

Fig.1: (a) A record of cooling during rolling schedules 1CS and 2CS; (b) A detailed view displaying the exact value of the rolling temperatures with respect to the Ar_3^d and Ar_3 temperature.

The same rolling schedules were also performed, followed by a water quench after the last rolling pass at different temperatures between the last rolling pass and the coiling temperature (600°C). This allowed to study the ferrite grain size evolution in the intercritical region. The rolling temperature together with the rolling passes for schedules 1CS and 2CS were recorded vs. time and plotted as shown in Fig.1.

The samples for metallographic observations were cut from the middle of the rolled sheet and the microstructure was studied in the plane perpendicular to the transverse direction after an appropriate sample preparation. This included mechanical grinding and polishing up to 1 μ m diamond paste, followed by 2% nital etching for observation in the optical microscope or the SEM. An additional electrolytic polishing stage after the mechanical one followed by 2 % nital etching is required for the OIM samples. The OIM attachment was installed on a Philips XL30 ESEM with an LaB₆ filament and the electron backscattering diffraction (EBSD) patterns were acquired and analyzed by means of the commercial TSL[®] OIM software [8].

Results and discussion

Fig.2. (a) and (c) Middle thickness microstructure of strips 1CS and 2CS; (b) Surface microstructure of strip 1CS. (d) Grain size distribution in the surface and center of the strips 1CS and (e) in the center of the strips 1CS and 2CS .

the midsections of these strips. In both strips the following features were observed (i) a *strong grain refining effect* and (ii) a *bimodal grain size distribution*. The grains larger than 10 μm and smaller than 3 μm were selected as representatives for the large and the small grains (cf. Fig.2c) and their textures are displayed in Fig.3. Fig 4 displays the ideal BCC product orientations in the $\phi_2=45^\circ$ section together with the FCC parent orientations from which they originate according to the Kurdjumov–Sachs relations. Fig. 4 should be used as a key to interpret the experimental data presented in Figs. 3a-d.

The texture of the ferrite grains smaller than 3 μm (cf. Fig.3a and c) is the transformation texture that could be obtained when a strained parent austenite phase (mainly with Brass $\{110\}\langle 112 \rangle$ and Copper $\{112\}\langle 111 \rangle$ orientations) transforms to ferrite according to K-S transformation law. The product ferrite orientations are usually situated on the so-called “transformation” or β -fibre (dashed lines in Fig.3.a-c). As it was expected the grains

Three different possibilities for grain growth were explored in order to explain the preferential growth of the large grains: (i) *transformation induced recrystallization*; (ii) *selective growth of specific orientations* and (iii) *growth advantage based on the fact that first nuclei grow first*. The orientation data were partitioned between the large and small grains. The textures of the large and the small grains were compared with the theoretically predicted ferrite transformation textures produced from strained and non-strained parent austenite. If transformation induced recrystallization is the acting mechanism for preferential growth, the large grains must display a transformation texture originating from strained austenite, whereas the texture of the small grains should be typical for warm rolled or recrystallized ferrite grains. Fig. 2 displays the Image Quality (IQ) maps of strips 1CS and 2CS that were finished below and above the A_{r3} temperature respectively. There were no significant differences with regard to microstructure and grain size in

Fig.3 A $\phi_2=45^\circ$ sections of ferrite ODFs of small (a, c) and large (b, d) grains in strips 2CS (a, b) finished above A_{r3} and 1CS (c, d) finished below A_{r3} temperatures.

Fig.3d displays the texture of the grains larger than 10 μm in the strip 1CS (finished below the Ar_3 temperature). The rotated cube $\{001\}\langle 110\rangle$ texture component in the ODF can be attributed to the transformation product of strained austenite with Brass $\{110\}\langle 112\rangle$ orientation or to the transformation product of the recrystallized austenite parent phase, while the observed strong rotated Goss $\{110\}\langle 001\rangle$ texture component can *only* be obtained after transformation of the recrystallized austenite parent phase with cube $\{001\}\langle 100\rangle$ orientation. If the transformation induced recrystallization mechanism acts, the large grains should display the transformation texture originating from strained austenite, because they are the

Fig.4 Selected BCC product orientations in the $\varphi_2=45^\circ$ section shown together with FCC parent orientations from which they originate according to the K-S relations [9]

random misorientation distribution. According to some authors middle orientation boundaries (with misorientation angles between 20° and 30°), rather than CSL boundaries are responsible for selective growth. The fractions of various misorientation classes (large, middle and small) are shown in Fig.6 for the samples 1CS and 2CS. For each sample a comparison is made between the misorientation profile observed for the entire sample and the profile obtained from local misorientation data between large and small ferrite grains. For all misorientation classes the difference between the

Fig.5. OIM map, which displays rotation angle grain boundaries with misorientation of 5-15°; 15-20°; 20-40° and 40-65° (From light to dark). The local misorientation between the large grains (grain 1) and the neighboring grains (2, 3 etc.) is measured between the marked

Fig.6 Fraction of the grain boundaries with different misorientation for the overall scan and in the neighborhood of the large grains

global and local misorientation profiles could be ignored, except for the sample 2CS, which displayed a slight dominance of middle orientation boundaries around the large (growing) grains. It remains to be verified, though, on sound statistical grounds whether or not this is a significant result.

The third assumption, surmising that the first transformation nuclei grow first (*i.e. the oriented nucleation hypothesis*) was explored by means of a detailed microstructural observation of the specially prepared samples, finish hot rolled at 820°C, which is above the Ar_3^d temperature (but below T_{nr}) and water quenched from different temperatures in the intercritical range. In this way the microstructural changes were traced in the temperature range between the final rolling pass and the coiling tem-

Fig. 7 (a) Microstructure of strip finished at 820°C and next water quenched from 780°C (~30% ferrite phase) (b) water quenched from 685°C (~50% ferrite phase) . 1- nuclei on the triple junction points; 2-nuclei on the grain boundaries; 3- nuclei on the shear bands. At 685°C large ferritic grains already exist in the microstructure without intercritical rolling.

Fig. 8 Grain size distribution in terms of area fraction observed after quenching at 685 and 780°C, respectively.

perature. Fig. 7 displays the microstructure of strips, which were hot rolled under the conditions of rolling schedule 2CS but instead of being slowly cooled in a coiling simulation they were air cooled to different intercritical temperatures and next water quenched. The first ferrite nuclei appear mainly on the former austenite grain boundaries (Fig.7a, points 2), but the largest nuclei were observed on the triple junction points (cf. points 1 in Fig. 7a). These two groups of nuclei seem to have a preferential growth after the further decrease of the temperature. At lower temperatures small amounts of new nuclei were observed inside the austenite grains in the shear bands or twins (Fig.7b, point 3) as it was observed also by Khlestov et al [11]. With decreasing

quenching temperature the ferrite growth is clearly observed in the microstructure as it is shown in Fig. 7a and b. The black line in Fig. 8 shows the grain size distribution of the ferritic grains at temperatures close to Ar_3 and it is easy to be seen that the area fraction of the large grains is relatively low in comparison to the one corresponding to the small grains. The decrease of the quench temperature, however, leads to an increase of the fraction of large grains and the appearance of the bimodal shape of the curve (cf. Fig.8). Hence, these data strongly suggest that the growth advantage of the first ferritic nuclei has contributed to the development of a bimodal grain size distribution as observed in the present experiment.

Conclusions

A strong grain refining effect and a bimodal grain size distribution is observed both after final rolling above and below the Ar_3 temperature. The observed bimodal grain size distribution after intercritical rolling of the studied Nb-V alloyed carbon-manganese steel cannot be explained by the transformation induced recrystallization hypothesis, whereas a more detailed statistical analysis is required to clarify the role of selective growth. It could be established, however, that preferential nucleation of ferrite grains in the vicinity of austenite grain boundaries and triple junctions has led to a significant size advantage to these grains in the ensuing growth stage, which has largely contributed to the formation of a bimodal grain size distribution.

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