TEXTURE CONTROL IN PIPELINE STEELS BY THERMO-MECHANICAL CONTROL PROCESSING

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Abstract: The texture evolution in pipeline steel is studied as a function of the thermo-mechanical processing parameters – rolling temperature and cooling rate after the finish rolling stage carried out with equal rolling reduction per pass. The austenite recrystallization and grain growth during thermo-mechanical control processing (TMCP) of a pipeline steel grade are described and analysed in terms of precipitation state progress. Two stage controlled rolling (roughing and finishing) was carried out on a laboratory rolling mill for a set of completed and interrupted schedules. Subsequent to rolling, two different cooling routes were used water quenching and accelerated water cooling (ACC) together with coiling simulation. From the combination of transmission electron microscopy (TEM) observations, detailed texture analysis and inductively coupled plasma-mass spectrometry (ICPMS) precipitates quantification, consistent correlations between precipitation state and microstructure at every stage of TMCP can be recognized. The formation of specific transformation textures was explained via appropriate texture models that describe both formation of transformation and recrystallization textures in the austenite.

Keywords: PIPELINE STEEL, PROCESSING, MICROSTRUCTURE, TEXTURE

1. Introduction
Many sources of oil and natural gas are situated in regions with extremes of temperatures varying from -80°C to 20°C. This put severe requirements to the pipeline steels from which pipelines are made. The requirements as high fracture toughness at low temperatures and strength in combination with excellent weldability and high thickness of the plates are only part of them which allow them operating under high pressure at low temperatures [1, 2]. Such desired combination of properties is obtained successfully in high strength low alloy (HSLA) steels by combining of appropriate alloying strategy with thermo-mechanical control processing (TMCP). After TMCP the pipeline steel has microstructures containing ferrite and small fractions of pearlite (for the low strength and toughness grades), various types of bainitic ferrite in combinations with martensite/austenite islands (MA) or low carbon martensite and bainite for very high strength grades. The desired combination of high strength and toughness in these steels is obtained mainly via significant grain refinement and precipitation strengthening. Large deformation imposed to the material on its way from slab to plate provides significant microstructural and crystallographic textures and as a result, anisotropy of the majority of mechanical properties. Relationships between plastic anisotropy and crystallographic texture are well-described [3], but at least they should be combined with microstructure morphology data to give sound results for TMCP steels [4,5,6]. The toughness anisotropy is more difficult to be linked with the texture beyond the dependency of the transition temperature on the fraction of cleavage planes parallel to fracture surface in a certain direction of testing [6]. The current work studies the possibilities for texture control during the TMCP of pipeline steels.

2. Experimental
Blocks of 75 mm thickness, 95 mm length and 160 mm width were cut from cast slabs of HSLA steel whose chemical composition is shown in Table 1. The blocks were hot rolled in 2-high laboratory rolling mill (roll diameter, 400mm) in two steps - rough rolling and finish rolling. All specimens were reheated at 1250°C for 2h and hot rolled to 12mm thick plates with a final rough rolling temperature 1180 °C above the calculated austenite non-recrystallization temperature (Tnr). Different schedules were designed and carried out by maintaining identical roughing conditions but varying the process parameters of the finish rolling, namely, the temperature of the first rolling denominated as “start finish rolling temperature” (SFRT), end rolling temperature (ERT) and cooling rates. To study the changes in microstructure and texture part of the samples were quenched after the end of the rolling schedule and the other part were accelerated cooled to 560°C and next coiling simulation was made. The rolling schedules are shown schematically in Fig. 1. All final rolling schedules were executed with a constant reduction per pass.

Table 1. Chemical composition of the steel, (in mass %)

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<tr>
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<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Nb+V+Ti</th>
<th>Mo+Ni+Cu+Cr</th>
<th>Fe</th>
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<tr>
<td></td>
<td>0.06</td>
<td>1.6</td>
<td>0.02</td>
<td>0.003</td>
<td>&lt;0.1</td>
<td>&lt;0.1</td>
<td>balance</td>
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In order to study the evolution of the precipitation state during TMCP, quenching was done at different moments during processing (cf.Fig.1): Quench 0 (Q0) after reheating and before the first roughing pass, Quench 1(Q1) after rough rolling, Quench 2(Q2) before first rolling pass Quench 3 (Q3) after finishing at high FRT, Quench 4 (Q4) after finishing at low FRT. For determining the amount of niobium precipitated small specimens underwent a selectively
dissolved process. An amount of 0.8g of material was dissolved and filtered. Precipitates were trapped with an Anopore (0.02μm) filter while the dissolved iron matrix and the remaining elements in solid solution were filtrated. The Nb content attached to the filter and those in the residue phase were measured by using Inductively Coupled Plasma Mass Spectrometer (ICP-MS). Transmission electron microscope (TEM) operated at 200 KV equipped with Energy Dispersive X-ray Spectroscopy EDX was used to support the precipitates analysis. Optical microscopy and electron backscatter diffraction (EBSD) were used to characterize the microstructure. Crystallographic texture was measured in the 1/2 thickness of the plates using X-ray Diffraction (XRD) and processed employing FHM-MTM software developed by Van Houtte [7].

3. Results and Discussion

Fig. 2 shows the microstructure of the steels after reheating at 1250°C 1h (Fig.2a), end of rough rolling at 1180°C (Fig. 2,b) and before the first rolling at 1060°C (Fig. 2c). The shape and size of the prior austenite grains delineated with black lines after Bechet-Beaujard [8] etching show that the austenite recrystallizes completely after rough rolling and in the cooling stage to 1060°C austenite grain growth takes place.

The following rolling passes were carried out below the austenite non-recrystallization temperature (Tnr). Rolling below Tnr causes formation of pancaked grains and significant grain refinement after accelerated cooling as shown in Fig. 3.

Fig. 3: Microstructure of the steel after: (a) quenching from ERT 980°C (Q3); (b) from ERT 800°C (Q4); (c) after accelerated cooling from 980°C and (d) after accelerated cooling by 800°C.

The microstructures shown in Fig. 3(b, d) show also that after rolling at low start finish rolling temperature and accelerated cooling the microstructure is finer than after rolling at high finish rolling temperatures, although there is no significant difference in the prior austenite grain size (cf. Fig.3a and b). The average grain size diameter in samples rolled with low SFRT was 8μm whereas in the samples rolled at high SFRT it was 12μm. Hence the grain refining can be associated to the development of the internal defects in the austenite grain size that play role of potential nuclei for BCC phases.

Fig. 4: Experimentally determined precipitate size distribution at various TMCP stages.

The analysis of the Nb precipitates (Fig. 4) shows that the growth of Nb precipitates start already at the rough rolling stage (Fig. 4b) and it does not change significantly during the finish rolling if the SFRT is high (Fig. 4c). However, if the SFRT is as low as 800°C the precipitates grow and the fraction of precipitates larger than 100nm increases significantly.

Fig. 5: (a) Bright field TEM image showing Ti-Nb precipitates; (b) line scan and (c) EDX spectrum of a precipitate.

Different precipitation activity during controlled rolling has important influence on the crystallographic texture of the plates. The measured transformation textures after different rolling stages are shown in Fig. 6 as orientation distribution functions (ODF's).

Fig. 6: ODFs of plates rolled and quenched from different temperatures.
SFRT are similar and although the maximum intensities on the ODFs change depending on SFRT they remain only on the BCC transformation components originating from deformed austenite with brass and coper{112}<111> orientation.

The appearance of the rotated Kurdjumov-Sachs transformation mechanism. The calculated texture (Fig. 7 bottom left) shows very good agreement with the measured texture (Fig. 7 bottom left). The appearance of the rotated Goss texture component{110}<011> in the modelled texture is due to the fact that the model cannot take into account the valiant selections of the deformation textures of austenite and the transformation of the austenite to BCC phase was assumed to follow the Kurdjumov-Sachs transformation mechanism. The calculated texture components appear but only selected ones which are favoured in the current transformation conditions.

The appearance of different type transformation textures can be followed with the help of Fig. 8[10]. The figure shows the position of the ideal texture components in the Phi2=45° section of the Euler space where the most common BCC texture components appear. The main austenite texture components brass, coper, cube and Goss are plotted in the same section with grey circles. The arrows and the numbers show from which austenite texture component the BCC texture component originates and the number of variants. Austenite with S {123}<634> orientation is not shown because it does not appear in this section of the Euler space.

In real production conditions the plates follow or accelerated cooling route or air cooling route depending on the ancillary equipment of the rolling mill. This gives additional time of the austenite cooling route or air cooling route depending on the ancillary equipment. The time between the start finish rolling temperature (SFRT) they remain only on the BCC components which are typical transformation texture components of the recrystallized austenite like rotated cube {001}<110>, rotated Goss {110}<011> or more rare -Goss{110}<001> texture components. The appearance of recrystallization texture components and especially rotated cube could have negative influence on strength and toughness anisotropy of the plates [10]. Hence, to avoid appearance of this texture component the time between the final rolling and accelerated cooling should be reduced. On the other hand lowering the start finish rolling temperature (SFRT) causes significant growth of the precipitation (cf. Fig. 4d) and they cannot suppress effectively the austenite recrystallization.

4. Conclusions

- Significant grain refining effect was observed in the samples rolled at low SFRT in combination with accelerated cooling;
- Increase of the SFRT and delayed accelerated cooling cause strengthening of the texture component originating from recrystallized austenite;
- Precipitation kinetics plays significant role in controlling the recrystallization of the austenite.

6. References