Precipitations in micro-alloyed steel

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Niobium micro-alloyed high strength low alloy (HSLA) steels are widely used in civil construction, automobile and line pipe applications. These steels rely on thermo-mechanical rolling, a technique that simultaneously provides high strength and toughness by grain refinement. Since these steels typically contain low carbon contents, they also have excellent welding and cold forming properties.

The most important role of niobium as a micro-alloying element in thermo-mechanically rolled steel is the retardation of austenite recrystallization, which provides more nuclei for the γ/α transformation and thus a finer grain size. Besides niobium’s role in solid solution by delaying all diffusion controlled processes, its tendency to form carbides provides the dominant effect [3, 4].

HSLA-Steels contain either the elements niobium, vanadium and titanium or a combination of them. Premature precipitation of Nb(C, N) in the secondary cooling section of the continuous caster may trigger hot embrittlement and transverse cracking. Furthermore Niobium has to remain in solid solution in order to make optimum use of its metallurgical potential during thermo-mechanical hot rolling. This is of particular importance in a coupled casting and direct hot rolling process.

Depending on the hot rolling parameters, such as deformation amount and rate, temperature and interpass time, niobium carbide precipitation is generally incomplete with regard to the thermodynamic equilibrium state. As a result, a part of the niobium stays in solid solution after finish rolling in the austenite phase and is effective in retarding the transformation or allowing a strength increase by precipitation hardening in the ferrite phase [4].

The niobium in solid solution at the finish-rolling temperature is available for the formation of fine niobium carbo-nitride precipitates in ferrite. These are of the appropriate size for providing a strength increase via precipitation hardening. In accelerated cooled and tempered steel specimens some investigations did not reveal any evidence that additional precipitation in the ferrite occurred. Precipitation in ferrite was only found after subsequent cold deforming and tempering of the considered samples [3, 4].

Hence, the occurrence of precipitation depends strongly of the deformation, the cooling conditions and the chemical composition of the steel. It is the main intention of these investigations to identify the possible formation process of niobium carbides or other precipitations. It can be assumed that the orientation distribution of the niobium carbide precipitates is not random, because of the strong texture of the Fe-matrix. This could be a reason for anisotropic mechanical behavior.

Samples of different chemical composition were investigated at the high-energy beam line BW5 using an energy of ~100 keV. High-energy synchrotron radiation is needed to get a large diffractive sample volume. As an example the results of a hot rolled micro-alloyed steel (0.11 wt% C, 0.29 wt% Si, 1.66 wt% Mn, 0.144 wt% Nb+Ti+V) are presented here. The diffraction images were taken in the range – 50°≤θ≤50° with Δθ=4°. Fig.1 shows a complete 2-dimentional diffraction image of the investigated HSLA-steel. The strong reflexions of the Fe-matrix suppress the weaker reflexions of the precipitates.

We use a Rietveld algorithm (MAUD) to determine the volume fraction of the phases and – later on- to evaluate the texture [5]. The Rietveld evaluation points out the difficulties to find the right phase-composition, because the solution-precipitation process depends extremely on the thermo-dynamical conditions during hot rolling and heat treatment. Another problem was the relatively small amount of precipitates. In order to get a sufficient count statistics relatively long exposure time is needed that leads on the area detector to an overflow of the strong α-Fe-reflexions. To avoid this we used the secondary slit system available at the BW5 beam line [1, 2]. By means of this slit system the α-Fe-lines were shielded and diffraction rings of the precipitations are now visible. In fig. 2 such a partial 2-dimensional diffraction image (3.90°≤2θ≤4.72°) is shown.

The first and a very difficult step in phase evaluation was to assign the possible phases of precipitations. In the considered sample the best Rietveld fit was received using seven possible precipitations:

- Ti5N, TiN, Ti0.85V0.11, FeTi, MnV, Fe0.3Mn0.7 and Nb0.026Ti0.974
Fig. 3 shows a MAUD-Rietveld-fit based on the integrated diffraction image in fig. 2 with the above mentioned phases. The NbC-phase, often described in literature was not found in this sample. This can only be explained with deviating conditions during heat and plastic deformation treatment. This result shows the necessity to understand the forming process of the precipitations thermodynamically. A next step will be the investigation of samples with different contents of carbon, titanium, niobium and nitrogen treated under different temperature conditions.

References