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Influence of the austenitic rolling temperature on the microstructure of a TRIP steel before intercritical annealing

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Abstract In the present work, the influence of different thermomechanical treatments in the final microstructure of a multiphase-steel (0.2%C–1.5%Mn–1.5%Si) through of morphological characterization of the transformation products was studied. The goal of this work was to investigate the volumetric fraction variation and phase distribution present in the microstructure. The steel was submitted to intercritic annealing treatment ($\alpha + \gamma$) with and without previous reheating after the hot-rolling. The microstructure was observed by optical microscopy and mechanical properties were measured by Vickers microhardness. The results show a typical microstructure of multiphase steel ("dual-phase") with increase of the martensite volumetric fraction and hardness values when the steel was submitted to intercritic annealing without reheating after the hot-rolling.

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Introduction

High strength steels have found a growing market in the automobile industry due to the need for reducing vehicles weight as well as decreasing their fuel consumption. Among these, multiphase steels stand out as materials for structural components with higher strength. A great deal of research was conduced during the 1970s, mainly focused on austenitic stainless steel. During the 1980s, this was also observed for C–Mn–Si steels based on the development of a high amount of metastable retained-austenite in the microstructure [1]. The contribution of each microconstituent to the enhanced mechanical behavior of these steels is still not totally clear [2]. Therefore, much research effort is still needed on unraveling this contribution as well as on determining the influence of thermomechanical treatments on the amount of each phase.

In 1937, Günther Wassermann in the Kaiser-Wilhelm Metallurgy Institute, in Berlin, made the first observations about the alterations in the mechanical properties of an austenite–martensite transformation. In 1967, Zackay et al. described such transformation in austenitic steels, denominating the effect of "transformation induced by plasticity" [3]. Many researches were accomplished in the 1970s on the use of the effect TRIP, mainly focused on the austenitic stainless steels. Later, in the decade of 80, it was observed that effect also happens in C–Mn–Si steels, since it gets a significant amount of metastable retained-austenite in the microstructure.

The metastability of the retained-austenite can be high for the increased carbon content. This is possible through a thermal treatment constituted by intercritical annealing [4], to obtain a structure composed by pro-eutectoid ferrite and austenite, followed by cooling down to the range of bainitic transformation temperatures, where the austenite becomes

isothermally in ferrite-bainitic, before the bainitic transformation finish, the material is cooled to room temperature. The retained-austenites are located among the ferrite-bainitic laths. In both reactions the partition of carbon produces an enrichment of the austenite turning the metastable at room temperature. Under deformation, the retained-austenite transforms in to martensite. This phase transformation is denominated as TRIP effect or transformation induced by plasticity [5].

The steels assisted by TRIP effect present a microstructure composed by ferrite, bainite, and retained-austenite. The austenite is a phase metastable at room temperature. The multiphase steels present good formability since there is an appropriate microstructure obtained through the control of the thermomechanical process parameters and chemical composition. The forming pieces presented high mechanical resistance due to the high coefficient of hardening proportionate by TRIP effect.

In agreement with Bleck [3], 2002, the manganese has an inhibiting effect in the ferrite formation, during the cooling, once that element minimizes a possible enrichment of carbon of this phase. In that case, the material product of the second rolling sequence, differently of the annealed material, presents a banding in the microstructure. It is believed that this morphology is consequence of manganese segregation.

This work examines the compromise between strength (presented by microhardness results) and the complex microstructure composed by pro-eutectoid ferrite, retained-austenite and transformation products of a 0.2% C–1.5% Mn–1.5% Si–0.029% Nb steel subject to thermomechanical treatments.

Experimental procedure

The material used in this work is a Fe–C–Si–Mn (chemical composition is shown in Table 1) supplied by Companhia Siderúrgica Nacional (CSN), in Brazil, extracted from 23 mm thick hot rolling plates.

The pilot hot-rolling was accomplished in the CSN Research Center. Before pilot hot-rolling, the plate was pre-heated at 750 °C and heated up to 1,250 °C in a furnace (muffle), followed by pilot hot-rolling (below the T_{NR}) in a FENN-135 rolling mill, with three step rolling performing 65% thickness reduction, in way to obtain sheet with approximately 3 mm thick. The subsequent treatment was divided in two sequences: (i) cooling down in air at room temperature after the pilot hot-rolling, with subsequent intercritical annealing at 800 °C for 4 min followed by tempering in ice brine at −15 °C (Fig. 1); and (ii) isotherm (intercritical annealing) at 800 °C for 4 min soon after the pilot hot-rolling, followed by tempering in ice brine at −15 °C (Fig. 2). The acquisition of the temperature during pilot hot-rolling was made using a thermocouple fixed in the plates.

The morphology study, after quench in both thermomechanical treatment sequences (Figs. 1 and 2), microconstituent identification, and their respective volumetric fractions were made by optical microscopy. The samples in longitudinal section (parallel to rolling direction) fixed in bakelite were prepared for optical microscopy in the following sequence: wet sanding in 180, 220, 400, 600, and 1,200 mesh sandpaper and polished in alumina of 1, 0.3, and 0.05 μm. The phases were revealed, initially, immersing the samples in 2% Nital solution for 3 s in order to measure the ferrite grain size and the ferrite volumetric fraction. In the second step, the samples were immersed in

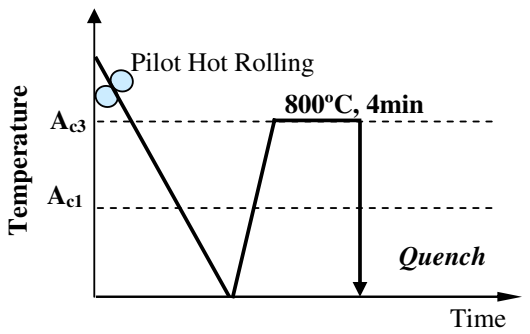

Fig. 1 First scheme of thermomechanical treatment

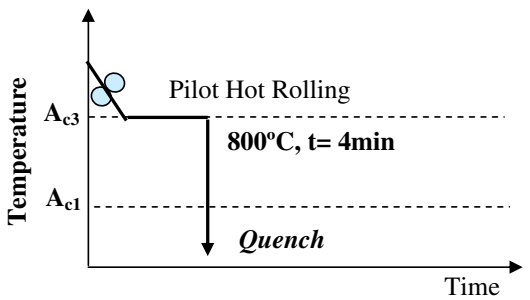

Fig. 2 Second scheme of thermomechanical treatment

Table 1 Chemical composition of the steel

Element	C	Mn	Si	P	S	Al	Cr	Ni	Nb	N
(wt)%	0.20	1.5	1.5	0.020	0.006	0.026	0.020	0.006	0.029	0.0039

Table 2 Component volume fractions and microhardness (HV) values

Samples	V_{α}(%)	V_{MA} (%)	(HV)
Treatment one	60 ± 1.8	32.7 ± 1.7	346 ± 2.8
Treatment two	17.3 ± 1.0	82.7 ± 1.0	428 ± 5.0

a 1% metabisulfite solution in order to measure the amount of representative M.A. (martensite plus retained-austenite).

In order to determine a ferrite volumetric fraction, a net with 117 points on the image was used. The Leica image analyzer Quantimet Q600 was used to measure the M.A. volumetric fraction, with 20 images analyzed by sample.

The Vickers hardness test (load of 30 kgf and 5 point by samples) was performed to evaluate the strength behavior in thermomechanical processes in study (Table 2).

Results and discussion

Figure 3 shows the optical micrograph of the as-received condition. The microstructure is composed by ferrite (white regions) and perlite (dark regions).

Figure 4 shows the microstructures of the specimens, after etching with Nital and metabisulfide, respectively, for each of the thermomechanical schemes. The first thermomechanical treatment, called "treatment one" (Fig. 1) consisted of isothermally soaking the specimen in the two-phase region ($\alpha + \gamma$), at 800 °C, for 240 s, in order to control the fraction of pro-eutectoid ferrite formed at the austenite grain boundaries [6]; the second one (Fig. 2) was aimed at producing a two-phase steel straight from a deformed austenite microstructure. To this end the last three passes of a controlled rolling scheme were conduced below the T_{NR}, the specimens being immediately subject to treatment one. This allowed the evaluation of the effect of the NbC precipitation and of the strain in the unrecrystallized austenite grains on the volume fraction of retained-austenite. In both cases the samples were quenched in icy brine (−15 °C) to develop the final two-phase microstructures.

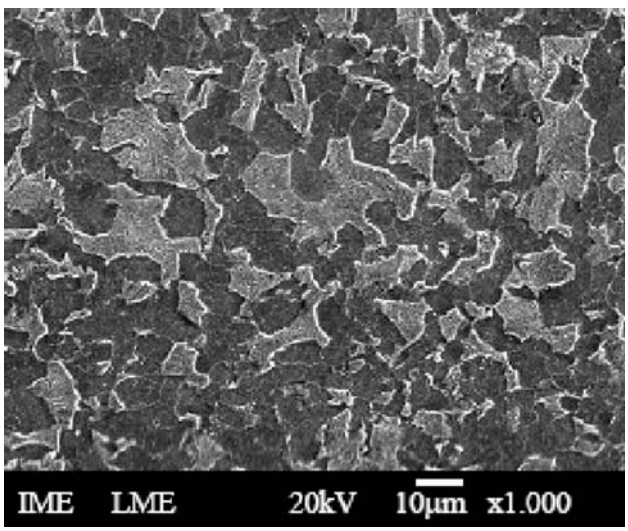

Fig. 3 Optical micrograph of the as-received condition

The photomicrographs in Fig. 4 confirm that a two-phase steel was obtained (Fig. 4a,c)—ferrite plus martensite—in which the light areas are ferrite grains and the darker ones correspond to quenched areas rich in carbon (martensite). According to Davies [7], the observed grain refinement is a consequence of the previous hot-rolling above the T_{NR}, which allowed the austenite to recrystallize more than once.

The samples resulting from the second thermomechanical treatment (Fig. 4b,d) show the characteristics of the controlled rolling, where the austenite was prevented from recrystallizing by the NbC precipitation. Ferrite grains are seen to nucleate at the boundaries of previously elongated austenite grains now turned into martensitic (darker) areas by the final quench. The products from both thermomechanical schemes display banded microstructures, probably as a consequence of Mn segregation, due to the high content of this element in the steel.

However, distinguishing retained-austenite from martensite is more difficult [8]. The volume fraction of the two components can be seen in Table 2, obtained by quantitative metallography. There, the effect of treatment two on ferrite nucleation in the two-phase region can be clearly observed. It is clear that treatment one leads to a larger fraction of ferrite. At first, one would expect it to be just the opposite but it seems that treatment one is much more effective in refining the austenite microstructure, leading to a larger ferrite nucleation rate at the austenite grain boundaries. On the other hand, preventing austenite recrystallization also leads to larger austenite grains chemically stabilized in the two-phase treatment. These in turn transform to larger volumes of MA during quenching.

Since it is mainly ferrite that is responsible for the strength of these materials, it is clear that controlled rolling is certainly a way of increasing the material's yield strength, but at the cost of drastically reducing its ductility, to prevent austenite recrystallization. This results from the increased volume fraction of transformed austenite in treatment two.

The samples hardness values are related with the material mechanical resistance. High hardness values in the treatment two implicate in the increased of the material mechanical resistance.

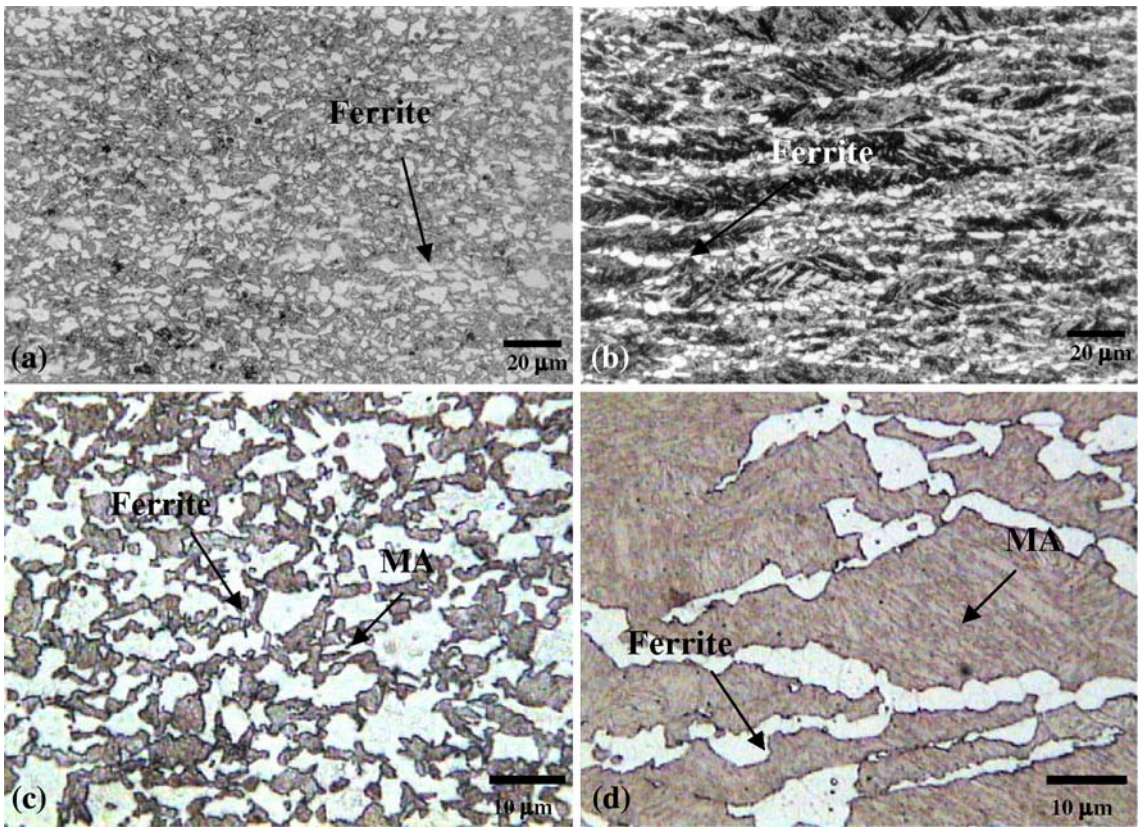

Fig. 4 Optical micrographs for **a** and **c** treatment one; **b** and **d** treatment two. **a** and **b** Nital 3%; **c** and **d** Sodium metabisulfite 1%

Conclusion

It was possible to characterize the effect of controlled rolling (hot-rolling below the T_{NR}) on the volume fractions of the multi-phase (ferrite and MA) steel produced by an isothermal soaking treatment plus quenching straight after the rolling scheme. In this material, it seems that preventing austenite recrystallization leads to larger austenite grains chemically stabilized in the two-phase treatment that, in turn, transform to large volumes of MA (martensite plus retained austenite) during quenching. The same intercritical treatment, using reheated specimens stemming from a previous hot-rolling above the T_{NR}, leads to larger ferrite volume fraction since this rolling scheme is more effective in refining the austenite microstructure. The high hardness values in the sample submitted to treatment two improve a better mechanical resistance suitable to the application of this kind of steel than the treatment one, what is also due the largest amount of martensite.

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