Mechanical Behavior and Microstructure of High Carbon Si–Mn–Cr Steel with Trip Effect

Dagoberto Brandão SANTOS,¹⁾ Ronaldo BARBOSA,¹⁾ Priscila Paulina de OLIVEIRA¹⁾ and Elena V. PERELOMA²⁾

1) Department of Metallurgical and Materials Engineering, Federal University of Minas Gerais, Rua Espírito Santo, 35, 30160-030 Belo Horizonte, MG, Brazil. E-mail: dsantos@demet.ufmg.br

2) BlueScope Steel Metallurgy Centre, School of Mechanical, Materials and Mechatronics Engineering, Faculty of Engineering, University of Wollongong, Northfields Avenue, Wollongong, NSW2522, Australia.

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The influence of the thermal cycle and austempering treatment on the mechanical behavior of a 0.56C–1.43Si–0.58Mn–0.47Cr (wt%) steel has been investigated. The thermal cycle consisted of heating the steel to 800°C or 900°C, in the intercritical and austenitic region respectively, fast cooling down to 600°C or 400°C, followed by 300 s hold. After austempering the materials were cooled at different rates and then submitted to tensile testing. The total elongation of 15–20% and tensile strength of 1 300–1 400 MPa were reached after heating to 900°C and transformation at 400°C. The strain-induced austenite transformation to martensite during the plastic deformation (Transformation Induced Plasticity Effect) is responsible for this combination of high strength and ductility. Silicon acts to stabilize the austenite during austempering. However, the stabilized austenite can be transformed and the microstructure modified, resulting in the formation of others constituents such as bainitic ferrite, upper bainite and martensite.

KEY WORDS: thermal cycle; annealing; austempering; TRIP effect; retained austenite; bainite; work-hardening.

1. Introduction

Production of steels with higher strength without loss of formability has been the subject of many recent investigations. The combination of high strength and high formability of steels can be achieved using Transformation Induced Plasticity (TRIP) phenomenon.

It is well known that a significant amount of retained austenite can be obtained in low carbon steels containing high contents of Mn and Si (1.5–2%).^{1–3)} The silicon inhibits the precipitation of carbides when the steel is isothermally treated in the region of bainite transformation. For this reason, during the bainite transformation, the adjacent austenite is enriched with carbon and it becomes stable at room temperature.^{1,2)} Consequently, the retained austenite can be transformed into martensite during the plastic deformation promoting a tensile strength increase and ductility without losing the formability capacity.^{4,5)} This is the principle of TRIP effect; originally published by Zackay *et al.*¹⁾ and it has been subject of many studies.^{1–5)}

The TRIP-aided steels have been developed as steel sheet for automotive reinforced members. These steels possess not only good formability but also fatigue strength, high impact energy and good hydrogen embrittlement resistance due to the presence of retained austenite in their microstructure.⁵⁾

Studies involving steels with high contents of Si and Mn

report the presence of bainite, retained austenite, martensite and carbides in the microstructure.^{3,5)} The time–temperature–transformation diagrams of C–Mn–Si steels present two different curves in C form, phenomenon that also happens in steels alloyed with elements which are strong carbide formers. According to the studies by Bradley and Aaronson⁶⁾ on steels containing Cr and Mo, the temperature range of the bainite formation gets delimited by a bay. The upper bainite consist of aggregates of ferrite lath separated by regions called residual phases.⁷⁾ Depending on the chemical composition, temperature and time of isothermal treatments in the bainitic transformation region, these residual phases can be composed of non-transformed austenite or of phases that form subsequently to the bainitic ferrite formation: retained austenite is present when bainitic transformation is incomplete, carbides when it is complete, and martensite will form later on cooling.

Thermomechanical processing parameters for low carbon steels, 0.1–0.25% C, have been extensively investigated in order to achieve a high yield strength coupled with a good formability.^{4,5,8–12)} However, there are only few works published on thermomechanical processing of high carbon steels,^{3,5,8,9)} which in addition have used a complex heat treatment route.⁹⁾ On the other hand, high carbon steels can be used in automotive applications as direction members, springs, screws and others structural and forging parts. They exhibit some important mechanical properties, such

as high mechanical strength, high fatigue strength, and toughness, in addition to the reduction of hydrogen embrittlement due to the presence of austenite.¹⁰⁾ Another important aspect in TRIP-assisted steel is the influence of retained austenite volume fraction and their stability on mechanical behavior of steel, mainly the uniform elongation.^{10,11–15)}

Therefore the aim of present work was to investigate the effect of austempering thermal cycle on the microstructure formation and mechanical behavior of high carbon (0.6% C) Si–Mn–Cr steel.

2. Experimental Procedure

The chemical composition of the steel used in this experiment is given in **Table 1**. The transformation temperatures, Ac_1 and Ac_3 , estimated by the empirical formulas of Andrews¹⁶⁾ and by the TTT diagram¹⁷⁾ of the steel. The martensite (M_s) and bainite (B_s) transformation start temperatures were also calculated to be 268°C and 594°C,¹⁷⁾ correspondingly.

The samples were cut from commercial wire rod steel with 8 mm diameter, which main application is the manufacturing of hot coiled spring for automotive industry. These samples were used to machine the cylindrical tensile specimens of reduced size with 4 mm diameter and 25 mm gauge length. These tensile specimens were subjected to the thermal cycle of annealing and austempering, as shown in **Fig. 1**. The samples were austenitized at temperatures of 800°C or 900°C for 300 s in a radiant sealed silica tube furnace with inert argon inert atmosphere, and later they were cooled with forced gas helium blown inside of the silica tube to the austempering temperature, 600°C or 400°C at a cooling rate of about 20°C/s. Such conditions aimed to form bainite and to maintain a specific volume fraction of austenite. A soaking time of 300 s was used for both austenitizing and austempering. After the austempering period the samples were cooled down in the furnace, in the air or in the water promoting three different cooling rates, which correspond to 4, 8 and 380°C/s, respectively.

The retained austenite volume fraction was quantified by X-ray diffraction with $Cu K\alpha$ radiation using direct comparison method. This method uses the integration of the most intensive peaks of austenite, characterized by the planes (111), (200), (220) and (311); and of ferrite, characterized by the planes (110), (200), (211) and (220). The X-ray data were also used to determine the carbon concentration in retained austenite. The austenite lattice parameter extrapolation method, together with the following Eq. (1), was employed to calculate the carbon content.^{8,18)}

$$C_\gamma = (a_\gamma - 3.578) / 0.033 \dots \dots \dots (1)$$

C_γ is the carbon concentration in retained austenite in %wt and a_γ is its lattice parameter in Å.

Following the thermal cycles the samples were submitted to the tensile testing. Tensile tests were performed on an Materials Testing System machine using a crosshead speed of 0.033 mm s⁻¹. After the tensile tests the samples were prepared for the metallographic examination. It was applied separately two etches to the polished surface, the first with nital 2% or the second one with LePera etchant.^{19,20)} A

Table 1. Chemical composition and transformation temperatures of the steel used (mass%).

Steel	C	Mn	Si	Cr	Ac_1 (°C)	Ac_3 (°C)
C-Mn-Si	0.56	0.58	1.43	0.47	768	810

Fig. 1. Heat treatment processing schedule. T_H is heating temperature and T_A is austempering temperature.

special etchant has been used to reveal residual austenite as a light-etching constituent. The heat-tinting technique was used to distinguish austenite and martensite.²¹⁾ An unmounted sample was polished and etched with 2% nital for about 15 s. After that, the sample was heated in a furnace in air at 260°C for 2.5 h, and then cooled to room temperature. With this technique, the various phases appear as different colors under the microscope. Polygonal ferrite and bainitic ferrite are beige, retained austenite is purple, and martensite is dark blue.

The microstructural analysis was completed with scanning electron microscopy (SEM) and Transmission Electron Microscopy (TEM). Selected samples were studied using a Philips CM20 Transmission Electron Microscope. Thin foils for TEM study were prepared from mechanically grounded to 0.25–0.3 mm slices. The 3 mm diameter discs were first punched out from the slices and then were electropolished in a solution of 5 wt% perchloric acid in methanol using a twin-jet Tenupol unit, operating at 30 V and 0.2 A. The polishing solution was cooled in the range of –20––30°C using liquid nitrogen. TEM examination was carried out at 200 kV using bright field, dark field and selected area electron diffraction modes.

3. Results and Discussion

3.1. Effect of Thermal Cycle on the Microstructure Formation

The initial as hot rolled microstructure is shown in the **Fig. 2(a)**. It consists of fine pearlite and some islands of pro-eutectoid ferrite, which volume fraction is 6.3%. In the **Fig. 2** there are also scanning electron micrographs of the samples after austenitizing. The microstructure of the samples austenitized at 800°C consists of isolated carbides, intercritical ferrite and martensite, austenite at 800°C, (**Fig. 2(b)**), whereas samples quenched after austenitizing at 900°C contains only a small amount of ferrite and martensite (**Fig. 2(c)**). The presence of ferrite in the microstructure of austenitized samples indicates incomplete austenitizing in spite of the high temperatures used (**Fig. 2(c)**). This is a

(a) As received hot rolled steel; Vv ferrite $6.3 \pm 0.3\%$; HV = 312

(b) Austenitizing at 800°C and quenching in water, Vv ferrite $3.5 \pm 0.15\%$; HV = 830

(c) Austenitizing at 900°C and quenching in water. Vv ferrite $2.5 \pm 0.13\%$; HV = 884

Fig. 2. Scanning electron micrograph of steel as-received (a) *i.e.*, hot rolling, and of the water quenched samples for 300 s austenitizing at 800°C (b) and at 900°C (c). Etched with nital 2%. Fine pearlite and grain boundary polygonal ferrite.

main difference between low and high carbon TRIP steels. For low carbon steels, the initial microstructure is obtained after intercritical annealing, and consists of austenite and a large amount of ferrite ($\sim 77\%$)^{12,22}; whereas the microstructure of medium and high carbon steels is predominantly austenitic one. This difference will be addressed more fully later in the paper.

Figure 3(a) is an example of the multiphase microstructure obtained after reheating to 900°C, austempering at 400°C and then water cooling. This microstructure consists of fine pearlite, bainite and martensite/austenite (MA) constituent. After LePera etching, **Fig. 3(b)** gives the details of the same microstructure shown in **Fig. 3(a)**. The areas with white color refer to the MA constituent, the blue color reveals the laths of bainitic ferrite and polygonal ferrite, the brown color constituent is bainite (bainite containing cementite) and black regions are fine pearlite. It could be seen from **Fig. 3(c)** and **Figs. 4(a), 4(b)** that the bainitic ferrite contains interlath retained austenite films. In the steel with carbon content greater than 0.5% C, the retained austenite is present in two forms (i) wide interlath films, (ii) blocky

(a) Nital 2%

(b) LePera

(c) Heat tinting

Fig. 3. Microstructure of the steel heated at 900°C for 300 s, austempered at 400°C for 300 s, and cooled into water (a–c). The areas with white color refers to the martensite/austenite (MA) constituent, the blue color refers to laths of bainitic ferrite or polygonal ferrite, the brown color constituent is bainite (bainite containing cementite) and black regions is fine pearlite (b). After heat tinting (c), polygonal ferrite and bainitic ferrite are beige, retained austenite is purple, and martensite is dark blue.

islands along packet boundary, block boundary and/or prior austenite grain boundary, as well as inside bainitic ferrite lath (**Figs. 4(a)–4(c)**). When austempered at temperature higher than M_s this retained austenite has been chemically stabilized by high carbon enrichment. The XRD data has shown that average carbon content in retained austenite was 1.4 wt%, which resulted in the M_s temperature below the room temperature. This point will be discussed in the next section.

The thermal cycle involved austenitizing at 800°C, austempering at 400°C and water cooling, also produced a multiphase microstructure. The main parameter of interest is the bainite transformation temperature. Previous research has shown that low bainite transformation temperatures re-

Fig. 4. Microstructure of the steel heated at 900°C for 300 s, austempered at 400°C for 300 s, and cooled into water (a) or cooled at furnace (b). Heated at 800°C for 300 s austempered at 400°C for 300 s, and cooled into water (c). SEM micrographs.

sult in finer bainite laths, a greater degree of carbon partitioning from bainitic ferrite into the austenite, and subsequently greater austenite stability.^{8,12,21)} Although blocky RA was observed in ferrite in Fig. 4(c). From the micrographs in Fig. 4, it is found that the microstructure of the steel mainly consists of bainitic ferrite lath matrix and interlath retained austenite films (Figs. 4(a), 4(b)). If austempered at temperatures higher than martensite-start temperature of the steel quasi-ferrite and blocky martensite phases coexist with coarsened retained austenite films and bainitic ferrite matrix (Fig. 4(c)).

On the other hand, when the transformation took place at 600°C, the microstructure consisted mainly of fine pearlite (Figs. 5(a), 5(b)) in spite of the carbides that were already present during the austenitizing process (Figs. 2(b), 2(c)). Zone axis in Fig. 5(b) is $[111]_{\alpha}/[100]_c$, and α denotes ferrite and C denotes Fe_3C . P indicates pearlite and F-proeutectoid ferrite). In addition, proeutectoid ferrite was located at prior austenite grain boundaries. In many colonies pearlite has degenerated (see Figs. 5(a), 5(b)). These

Fig. 5. Representative TEM micrographs of sample soaked at 800°C for 300 s, transformed at 600°C for 300 s and water cooled. Zone axis is shown in (b).

pearlite colonies were much coarser in the sample water cooled after 800/600°C processing compared to the water cooled sample after 900/600°C treatment (Figs. 6(a), 6(b)). Zone axis in Fig. 6(b) is $[110]_{\alpha}/[11\bar{1}]_c$. Arrows indicate carbides at pearlite/ferrite interface. Presence of coarse carbides was also observed at pearlite/proeutectoid ferrite interfaces, as clearly visible in Figs. 6(a), 6(b).

Formation of bainite starts from the austenite, which volume fraction and carbon content were determined by the previous polygonal ferrite transformation during cooling from austenitizing to austempering temperature, Fig. 7(a). Together with bainitic ferrite there were some islands of martensite, Fig. 7(b), as pointed out before in Figs. 4(a), 4(b). Zone axis in selected area diffraction pattern (inset in Fig. 7(a)) is $[112]_{\gamma}$. Zone axis (inset in Fig. 7(b)) is $[110]_{\alpha}$. α denotes martensite and γ denotes austenite, M-martensite, F-polygonal ferrite and RA is retained austenite. Twin system is $(\bar{1}12)[\bar{1}\bar{1}0]_{\alpha}$.

Even when steel contains substitutional alloying elements, such as Si and Al, which are known to retard the formation of cementite, the untransformed austenite could decompose to cementite and ferrite during extended holding at bainite transformation temperatures. The first step during isothermal holding corresponds to the formation of bainitic ferrite, and the onset of the second step is recognized as the start of cementite precipitation from carbon-supersaturated austenite. In the case of bainite containing cementite, upper bainite, Figs. 4(a), 4(b) and, the carbide precipitation consumes the carbon enrichment at the reaction front and the volume fraction of retained austenite is reduced when observed at room temperature.

(a)

(b)

Fig. 6. Thin foil TEM micrographs of sample heated at 900°C for 300 s, transformed at 600°C for 300 s and cooled into water. Zone axis is shown in (b).

(a)

(b)

Fig. 7. Representative TEM micrographs of the steel heated at 900°C and austempered at 400°C. (a) and (b)—bright field images. Zone axis in selected area diffraction pattern (inset in (a)) is $[112]_{\gamma}$. Zone axis (inset in (b)) is $[110]_{\alpha}$.

Table 2. Retained austenite volume fraction and carbon content (between parentheses—%wt) as function of the cooling rate for the samples treated at 400°C and 600°C. *****—Not determined.

Volume Fraction and Carbon Content in			
Retained Austenite (%wt)			
Cooling Mode			
Sample	Water	Air	Furnace
800/400°C	16.8 (1.2)	2.0	2.2
900/400°C	19.3 (1.36)	18.5 (0.90)	13.9 (0.72)
800/600°C	*****	*****	*****
900/600°C	*****	*****	*****

These results indicate that carbon is rejected to the austenite during the transformation and that the diffusion away from the interface is likely to be a limiting step in the development of the laths.¹⁴⁾ Moreover, it is strongly suggested that cementite may provide extra sinks to absorb carbon. In this way, part of austenite will be prone to transform to martensite during cooling to room temperature, *i.e.*, furnace cooling, otherwise if the samples were water quenched, the bainite reaction is interrupted and more austenite could be retained until room temperature is reached.

3.2. Effect of Processing Schedules on the Retained Austenite Amount and Carbon Concentration

Application of X-ray diffraction to determine the volume fraction of retained austenite in the microstructure was only possible for the samples undergone austempering at 400°C. The data are presented in **Table 2**. The carbon concentration in the retained austenite is also shown in Table 2. It is clear that the highest volume fraction of austenite was obtained in water quenched samples after austenitizing at both 800 and 900°C. At the same time, the lowest volume fraction of retained austenite was obtained after furnace cool, which might be associated with decomposition of carbon-saturated retained austenite into carbides and ferrite (upper bainite) and with transformation after cooling to room temperature to martensite.

The retained austenite was not present when the austempering was realized at 600°C. TEM micrographs show in Figs. 6 and 7 that the microstructure consists predominantly of fine pearlite and polygonal ferrite; the austenite present at initial heating, at 800°C or 900°C, when was cooled to 600°C has been completely transformed to fine pearlite and some proeutectoid ferrite.

Figure 8 shows a characteristic XRD spectra for a specimen heated to 900°C, austempered at 400°C and then water quenched. It is clear the peaks from retained austenite, and this example represents a great amount of nearly 20% volume fraction.

The stability of austenite in Si–Mn TRIP steels is controlled by the bainite transformation, during which the formation of bainitic ferrite laths enriches the interlath austenite with carbon. If the austenite carbon level is sufficiently high, and the formation of Fe_3C is suppressed, the M_s tem-

Fig. 8. Characteristic X-ray diffraction pattern, illustrating the presence of retained austenite ($V_v = 19.3\%$) in the sample heated at 900°C for 300 s, isothermally transformed at 400°C for 300 s and then cooled into water.

perature will be below ambient temperature. As a result, the austenite survives the final cooling to ambient temperature without transformation to martensite. Thus, the degree to which austenite is retained depends upon the volume fraction and carbon content of the austenite at the start of the bainite transformation and the extent of carbon partitioning into the austenite during the transformation.^{21–26)}

As could be seen from Table 2, with increasing cooling rate after austempering at 400°C , the retained austenite fraction increases together with an increase in its carbon content. On the other hand, slow cooling after austempering allows extra time for transformation to take place, which results in the decrease of retained austenite volume fraction. This may be due to sluggish bainite transformation in this higher-carbon austenite.^{4,27)} Since the austenite carbon content in this steel is already high, further C-enrichment during the cooling stage stops the bainite transformation as the carbon content in the austenite reaches the T_0 -concentration. At T_0 , ferrite and austenite of identical compositions have the same free energy. Austenite enrichment to the level at T_0 is believed to represent completion of the bainite transformation by a displacive mechanism.¹⁷⁾ The microstructure of the furnace cooled sample (Fig. 4(b) and Table 2), showed less volume fraction of bainitic ferrite than of water quenched samples, and the presence of upper bainite supports the above hypothesis. Similar effect was observed during extended times for isothermal holding at 450°C in C–Mn–Si TRIP steel.^{4,27–30)} The fastest cooling rate inhibits the atoms diffusion, not only of the substitutionals but the interstitials also. In this way the retained austenite will sustain the carbon enrichment in water cooled samples, whereas in slowly cooled samples the decomposition of retained austenite is more feasible. Other researches revealed that the increase in the carbon content of retained austenite improves its stability leading to an increase in the mechanical strength and ductility, produced by TRIP effect.^{1,4,5,8,9)}

The microstructure of the TRIP steel could be characterized in terms of a number of morphological features, such as the size and morphology of the bainitic packets, the volume fraction and distribution of retained austenite and martensite, and the volume fraction and morphology of ferrite. These features should be related to the mechanical properties of TRIP steel.^{10–13,26)}

Fig. 9. Effect of annealing temperatures and different cooling rates on the tensile strength after austempering at 400°C (a) and 600°C (b).

Fig. 10. Effect of annealing temperatures and different cooling rates after austempering at 400°C (a) and 600°C (b) on the yield strength.

3.3. Effect of the Thermal Cycle upon the Mechanical Properties

The effect of processing schedules on tensile and yield strength are shown, in Figs. 9 and 10, respectively. Figure 11 provides three examples of stress–strain curves. The data on total elongation of steel is summarized in Table 3.

The strain hardening analysis behavior of metals can be performed using several methods. One such method em-

Fig. 11. Three examples of stress-strain curves of the steel after different processing conditions.

Table 3. Effect of the cooling rate in the total elongation (%) for the samples treated at 400°C and 600°C.

Sample	Total Elongation (%)		
	Cooling Mode		
	Water	Air	Furnace
800°C/400°C	10.5	10.1	25.3
900°C/400°C	26.3	22.7	28.5
800°C/600°C	11.1	13.5	7.2
900°C/600°C	17.7	10.5	18.7

employs the Hollomon equation associated with the Considère necking criterion and Jaoul-Crussard analysis,²⁹⁾ Eqs. (2) and (3).

$$\sigma = K\varepsilon^n \quad (\text{Hollomon}) \dots\dots\dots(2)$$

$$n = \varepsilon_u, \text{ when } \delta\sigma/\delta\varepsilon = \sigma \quad (\text{necking criterion}) \dots\dots\dots(3)$$

Where, σ is the true stress, ε is the true strain, K is a constant, n is the strain hardening exponent of the material, and ε_u is uniform elongation. It can be demonstrated that if the material follows such a relationship, the strain hardening exponent ($n = \delta \ln \sigma / \delta \ln \varepsilon$) should be constant and equal to the strain at the onset of necking.

The strain hardening exponent increased when the temperature of austempering change from 600 to 400°C, as described in Table 4.

Considering the transformation at 400°C, the water cooling promoted higher tensile strength for samples austenitized at both 800°C and 900°C temperatures, 1 071 and 1 350 MPa respectively, compared to the samples which were air cooled or furnace cooled, 1 039 and 1 168 MPa, for 800°C or 1 335 and 1 239 MPa for 900°C. The same behavior can be observed for transformation at 600°C, but the values are reduced to around 1 100 MPa. When the intercritical temperature is 800°C, these values are reduced by 10% (Fig. 9). Figure 11 shows an increase in work hardening and tensile strength as the annealing temperature increased and austempering temperature decreased, that is, 900°C and 400°C, respectively. An important result from Fig. 11 and Table 4 is that both samples that were austenitized at 800 or 900°C and transformed at 400°C showed the general TRIP effect.^{1,10-12)} The material tensile-tested, when transformed at 600°C, displayed the same level of tensile

Table 4. Effect of thermal cycle on strain hardening exponent of the samples treated at 400°C and 600°C. They were calculated from the 0.2% yield strength to the onset of necking.

Sample	Strain Hardening Exponent, n		
	(Obtained by Considère Criteria)		
	Cooling Mode		
	Water	Air	Furnace
800/400°C	0.16	0.10	0.25
900/400°C	0.25	0.27	0.27
800/600°C	0.16	0.17	0.15
900/600°C	0.15	0.09	0.10

strength, but with inferior levels of yield strength, total and uniform elongation, because there is no retained austenite. Annealing at 900°C promoted better mechanical results because at this temperature the steel was almost completely austenitized, resulting in a great amount of retained austenite than the specimens intercritically annealed that is after the isothermal treatment. Although, part of austenite, existing at austempering treatment, that contains less amount of carbon, transformed to martensite at cooled stage to room temperature, Figs. 4(b), 4(c).

The highest total elongation values were obtained after 900/400°C treatment for the three modes of cooling, Table 3. This could be explained by the occurrence of small austenite films dispersed (Figs. 4(a), 4(b)) in a relative ductile ferrite matrix that has a large influence on the work hardening rate. Due to the volume and shape change associated with the transformation of austenite to martensite, local plasticity is generated in the neighboring ferrite grains. Another important aspect is the presence of bainite with high toughness, which contributes mostly to strength. It has a good toughness and thus improves strength-ductility balance.

This tendency was observed for either annealing temperatures, 800 or 900°C (Table 3). Lower strain hardening exponent is a characteristic of the steels with high strength. Whereas austempering at 400°C resulted in the material with the highest yield strength, Fig. 10. It has the strain hardening exponent compared to 600°C isothermal treatment. Although, it still presents high values.

The microstructure is characterized by thin films of austenite between the bainitic ferrite laths, Figs. 4(a), 4(b) and 7. This behavior was found also at works of Sugimoto *et al.*^{5,13,28)} and recently at work of Lee.³⁰⁾

However, when the steel was cooled in the air yield strength was lower in the samples initially annealed at 800°C compared to the samples annealing at 900°C. On the other hand, the total elongation had its value considerably increased with annealing at 900°C, austempering at 400°C and water cooling as shown in the Table 3. The highest values of ductility and mechanical strength are related to the higher amount of retained austenite in the microstructure, which were obtained from the heating to 900°C, but also due to the multiphase microstructure, which contains in addition martensite and carbide-containing bainite.

As the amount of martensite increased the volume fraction of retained austenite decrease, and its carbon content too, *i.e.*, for furnace cooled samples heated at 800 or 900°C and transformed at 400°C, Table 2, although the total elongation are equivalent, Table 3. The explanation for this behavior is given by Sugimoto *et al.*³¹⁾

In general, monotonic deformation of the TDP steel is controlled by the following two items:

- (1) long-range internal stress resulting from a difference of plastic strain behavior between the ferrite matrix and the second phase; and
- (2) strain-induced transformation (SIT) of the retained austenite resulting in both the increases in strain-induced martensite content and the relaxation of the long-range internal stress.

These items are influenced by retained austenite stability. If the retained austenite is stable against straining, then the retained austenite particle, as a hard phase, effectively enhances the internal stress during early stages. The stable retained austenite tends to transform to martensite over a large strain range, thereby resulting in large elongations by TRIP effects. On the other hand, if the retained austenite is unstable, most of the retained austenite island transform to martensite at an early stage of deformation. The resultant strain-induced martensite considerably increases the flow stress with relatively low elongations and relaxed internal stress (Tables 2 and 3).^{5,8)}

Figure 4(c) shows a significant volume fraction of polygonal ferrite in the microstructure, martensite and retained austenite in blocky morphology, which contribute to maintain a high total elongation despite of lesser retained austenite amount. Such the retained austenite morphology is effective for transformation induced plasticity compared to retained austenite film along bainitic-ferrite lath⁸⁾ (Figs. 4(a), 4(b)) and according to previous research of the authors,³¹⁾ ductility of the low alloy TRIP-aided steels considerably increased when the matrix structure of bainitic ferrite was replaced in part by annealed martensite structure (Fig. 4(c)), which increase the compressive internal stress and led to equivalent elongation of 900°C heated samples.

When the cooling is fast enough so that the austenite does not transform itself during cooling. In present work, it was verified that the faster cooling promoted a higher amount of bainitic ferrite in the microstructure, comparing Figs. 4(a)–4(c) and 7, resulting in higher values of mechanical strength. Amar *et al.*²⁵⁾ and others researchers^{28,30)} investigated the influence of accelerated cooling in the microstructures and mechanical properties of low carbon TRIP steels and reached similar conclusions. For the samples transformed at 600°C, all conditions led to inferior mechanical strength, Figs. 9(b) and 10(b). The stable retained austenite within the bainitic structure affect the yield strength if one compares the results in Figs. 9 and 10. In this case, the volume fraction, morphology (films or coarse blocks), and location of martensite within the ferrite matrix were more significant in determining the structure-property dependence (Figs. 4(a), 4(c)). The distribution of different phases within the microstructure is important, because when a large fraction of a phase harder than bainite, such as martensite, is included in a soft matrix, plastic deformation at first occurs in the softer phase. On the other hand, the

Fig. 12. Two examples of instantaneous strain hardening exponent curves of the steel after different processing conditions.

hard phase only begins to deform when the softer phase has work hardened sufficiently to transfer load. The hard martensite phase under straining can promote the formation of new mobile dislocations in a ferrite matrix and, by means of this, increase the uniform elongation. Furthermore, the martensite was not uniformly distributed within the ferrite, so giving rise to a gradual deviation from elastic deformation.²¹⁾

The high capacity of the material strain hardening is quantified in Table 4, which means it, is coupled in one unique material of high mechanical strength, tensile strength of 1350 MPa, total elongation of 26% and a high strain hardening potential, n , which value is equal 0.25. This behavior can be understood considering the volume fraction of retained austenite (around 20%).

The results of flow curves provide interesting indications of the stability of austenite. Indeed, as shown in Fig. 11 and Fig. 12, a good stress-strain combination corresponds to a particularly favorable development of work hardening during plastic deformation. These curves show the flow and work hardening behaviors. Considering Fig. 12(b) it is possible to claim that for the material heated at 900°C, transformed at 400°C and water cooled there is a plateau on the middle of the curve which value is closed to 0.30. This is due to initial austenite stability and then its transformation to martensite after deformation in accordance with others researchers.^{3,21,22,32–34)} On the contrary, for the sample with pearlitic microstructure, obtained after heating at 900°C, transformed at 600°C, and air cooling, the strain hardening exponent curve *versus* strain shows an initial increasing in n values and a sharply drop, reaching a zero value for small true strain, 0.10 (Fig. 12(a)).

Although the n , strain hardening exponent, provides only a phenomenological description of work hardening, it is in-

fluenced by the active plastic deformation mechanisms, *i.e.*, dislocation glide and/or mechanically-induced martensitic transformation.

Dispersed bainite and martensite constituents strengthen the ductile ferrite matrix and delay plastic instabilities. Both the TRIP effect and the composite strengthening effect,^{14,22)} thus, contribute to a reasonable enhancement of mechanical properties as proposed by Jacques *et al.*¹¹⁾ The formation of the upper-bainite structure with coarse carbides between parallel bainitic ferrite laths (Fig. 4(b)), led to a reduction in the volume fraction of retained austenite (Table 2) and at the same time increase in the yield strength (Fig. 9). Carbides influence the ability of the lath boundaries to impede dislocation motion, thereby confining slip to within the laths and increasing the lath-size contribution to the flow stress.^{21,26)}

The mechanical properties of the steel achieved in this work, strength, ductility (considering gage length of 25 mm), and work hardening are superior to the third generation of Advanced High Strength Steel (AHSS), which tensile strength falls in the range of 1 200 MPa and 20% of total elongation (gage length of 50 mm).³⁵⁾

4. Summary

The effect of processing parameters, such as austenitizing and austempering temperatures, and cooling rates, on the microstructure and mechanical properties of Si-Mn-Cr high carbon steel has been studied. It could be concluded that:

The thermal cycle, which included annealing at 900°C, austempering at 400°C and water quench, has produced the highest values of yield strength, tensile strength (1 350 MPa) and ductility (~25%) due to the high amount of retained austenite remained in the microstructure at room temperature. The microstructure of the samples transformed at 600°C predominantly consists of ferrite and pearlite. This is associated with the reduction in strength and ductility compared to the samples transformed at 400°C.

Partial transformation of retained austenite, either during isothermal holding or during final water quenching, does not necessarily led to a degradation of mechanical properties. Localized work hardening of polygonal ferrite in the neighboring to hard phases areas (near martensite, bainitic ferrite or retained austenite) has contributed to the good combination of strength and ductility in this steel. Both the TRIP effect and the composite strengthening effect contribute to a pronounced enhancement of mechanical properties.

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