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Mechanical behavior of tempered martensite: Characterization and modeling

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ABSTRACT

Mechanical characterization of five low-carbon martensitic steels tempered at a wide range of temperatures and times was performed. Analysis of the relationships between mechanical properties, hardness and tempering conditions was completed. Microstructural observations revealed the precipitation of carbides and consequent decrease of carbon in solid solution with tempering. To describe the mechanical behavior of tempered martensite, a simplified Continuous Composite model, reported previously, was adapted. A good agreement between the model and the experimental stress-strain curves was achieved with only one fitting parameter. Further directions for improvement of the model were also proposed.

1. Introduction

The use of hardened steels date from thousands of years ago [1]. However, characterization and understanding of their microstructures started only after the availability of analytical and testing techniques in the end of the nineteenth century. The name martensite was given to honor the important contribution of Adolf Martens [2] in the metallographic observations of the hard phase found in the quenched steels. Now, this name is used to designate any phase in ferrous and nonferrous systems formed by diffusionless and solid-state shear transformation. In steels, martensite is a very complex phase that was studied in the twentieth century by many researchers [3–8]. Its structure varies from lath-like to plate-like, its substructure from dislocations to twins and its crystallography from BCC to BCT when the carbon content increases.

Tempering process is the reheating of a martensitic steel to a certain temperature, accompanied with the holding for a certain time, in order to improve the toughness and strength-ductility balance of the martensitic product. Tempered martensite has received increasing attention since it is one of the phases present in most of advanced high strength steels (AHSS).

Speich and Leslie [9,10] gave an account of the different microstructural changes that occur during tempering. Firstly, at tempering temperatures between 100 and 200 °C, the carbon segregation to the defects, started during quenching, continues to occur. Therefore, there is a depletion of carbon in the solid solution inhibiting the ε-carbide precipitation in low carbon steels. Secondly, at temperatures between 200 and 300 °C, the decomposition of retained austenite usually begins. The precipitation of rod-shaped cementite starts at 250 °C and it tends to be replaced by spheroidal particles at 400 °C. The recovery of the very deformed martensitic structure occurs more significantly at temperatures above 400 °C. Finally, recrystallization and grain growth are the last stages of tempering. Caron and Krauss [11] also studied the changes of the lath martensite when tempering at high temperatures. They noticed that the elongated packet-lath morphology of martensite is maintained up to long times at high tempering temperatures. In addition, they observed the laths coarsening using electron microscopy. In another work, Swarr and Krauss [12] observed the presence of disc-shaped and spherical carbides with 10 nm in diameter within the lattices, and larger carbides – 20 nm in thickness and 100 nm in length - in the lath boundaries of the tempered steel. They also observed the transformations in the substructure of the as-quenched and tempered martensitic steels before and after being deformed by tensile testing. While the as-quenched sample has a cellular structure that develops during deformation, the tempered sample has a uniform distribution of dislocations that practically does not change during deformation.

Tempering also impacts hardness. Grange et al. [13] observed that the hardness of as-quenched martensite depends only on the carbon content. However, the hardness of tempered martensite substantially increases with alloying elements additions. Hollomon and Jaffe [14] modeled the hardness evolution with tempering temperatures and times. They found the following relationship between the tempering conditions and the hardness:

$$R_c = H_c - 0.00457T\,(13 + \log t) \tag{1}$$

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where Rc is the Rockwell C hardness, Hc is a characteristic hardness, T is the tempering temperature in Kelvin and t is the tempering time in seconds.

Despite all the work performed on tempered martensite [15–21], the microstructural evolution during tempering at a fine scale and the effects of different microstructural changes on the strain-hardening of tempered martensite are still unclear.

In this context, the purpose of this work is to propose a composite model for the prediction of the stress-strain curves of tempered martensitic steels. Composite models started to be applied in the 20th century. They are based on the Masing's model of polycrystal plasticity [22], in which the individual grains yield at different external stress values.

Pedersen et al. [23] proposed a composite model to explain the Bauschinger effect of workhardened copper. In this model the dislocations tangles act as hard but deformable particles in a soft phase, practically free of dislocations.

Mughrabi [24] proposed a composite model for the plasticity of crystals with cell structures. In his model, the cells walls are considered as hard phases separated by soft regions with low dislocation density.

Polák et al. [25,26] built a more sophisticated composite model. Instead of considering just two phases as in the models of Pederson and Mughrabi, i.e., one phase with high and another with low dislocation density, this model deals with a continuous distribution of volumes with different internal critical flow stresses (σic). This distribution was characterized by a probability density function that can be written as:

$$\int_0^\infty f(\sigma_{ic})d\sigma_{ic} = 1 \tag{2}$$

In the present work, five low-carbon martensitic steels were tempered at a wide range of temperatures and times and the evolutions of mechanical behavior and microstructure were characterized. From these results, two composite models were proposed and their accuracy was compared.

2. Materials and methods

Five steels with different combinations of C, Mn, Si and Al were produced using vacuum induction melting. Their compositions are given in Table 1. The notation employed in Table 1 is LC for the low C (0.1%), LM for the low Mn (1.8%), HS for the high Si (1.4%), HCHS for the high C and high Si (0.25%C and 1.6%Si) and HA for the high Al (1.1%) steel.

The ingots were then hot rolled to approximately 4 mm thickness and coiled at 550 °C. Hot rolled sheets were cold rolled to about 1 mm thickness. Dilatometer tests and thermodynamic simulations were performed to determine the austenitization conditions. Further, samples for all the steels, except the HA one, were austenitized at 870 °C (the HA steel was austenitized at 950 °C) for 100 s and then water quenched. Thereafter, tempering treatments were conducted in oil baths for tempering at 150 °C and in salt pots for all other temperatures (from 200 °C to 500 °C). Holding time was varied from 10 s to 3 h. At the end of holding the samples were air cooled.

Three tensile tests at constant strain rate of 0.007 s^{-1} were performed for each tempering condition of different steels. The samples gauge length was 50 mm and the width was 12.5 mm, according to the ISO procedure.

Table 1
Measured chemical composition of alloys, in wt%, balance Fe.

Steel	C	Mn	Si	Al	Cr	Nb
LC	0.10	2.46	0.30	0.01	0.00	0.04
LM	0.21	1.76	0.27	0.05	0.00	0.00
HS	0.21	2.22	1.44	0.04	0.21	0.00
HCHS	0.25	2.43	1.57	0.02	0.00	0.00
HA	0.26	2.46	0.00	1.13	0.00	0.00

The presence, size and morphology of cementite was observed with SEM after etching with Picral reagent (4 g of picric acid in 100 ml of ethanol). Finally, conventional Vickers micro-hardness measurements were performed at 13 locations of the polished surfaces using the load of 1 kg and holding time of 10 s.

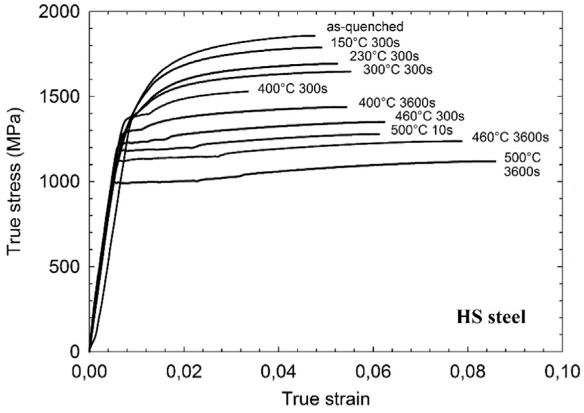

Fig. 1. Experimental true stress-true strain curves of some tempering conditions performed with the HS steel.

3. Results

Fig. 1 shows some of the experimental tensile curves obtained in the present study for the HS steel. These curves illustrate the mechanical behavior evolution with tempering and the expected softening of the steel.

Similarly, Fig. 2 presents some of the experimental tensile curves for the other steels (LC, HCHS, HA and LM). The softening behavior observed in Fig. 1 also occurs for these other steels. However, the LC steel presented some sort of temper embrittlement when tempered at 400 °C and 460 °C, breaking before reaching the plastic regime.

The measured Vickers hardness (HV), yield strength (YS), ultimate tensile strength (UTS) and uniform elongation (Uel) values are shown in Tables 2–6 for the LC, LM, HS, HCHS and HA steels, respectively. The hardness measurements were performed in most of the tempering conditions for all steels, except for the LM steel due to the unavailability of material. As the LC steel presented tempering embrittlement when tempered at 400 °C and 460 °C, it was not possible to determine the YS, UTS and Uel in these conditions. Finally, the as-quenched HCHS sample broke in a brittle manner and it was also impossible to obtain its UTS and Uel.

For the HS steel, the evolution of the yield strength and UTS with tempering temperature and time is illustrated in Fig. 3. In Fig. 3(a), the samples were tempered for 5 min therefore all the strength variation is caused by the temperature effect. While the UTS decrease in all temperature range, the yield strength stay constant up to 400 °C and then start to decrease. On the other hand, in Fig. 3(b), the samples were tempered at 460 °C and only the effect of the time is considered. Both yield strength and UTS decrease similarly since low values of tempering times. Additionally, it is possible to observe that the effect of temperature on the strength of tempered martensite is considerably greater than the effect of time.

Similar evolutions of the UTS and YS with tempering temperature and time were obtained for the other steels. The only difference is that the yield strength start to decrease at lower temperature for the LM and HA steels (approximately 300 °C instead of 400 °C for the other steels).