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Understanding the factors controlling the hardness in martensitic steels

E.I. Galindo-Nava *, P.E.J. Rivera-Díaz-del-Castillo

SKF University Technology Centre, Department of Materials Science and Metallurgy, University of Cambridge, 27 Charles Babbage Rd, Cambridge CB3 0FS, UK

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ABSTRACT

A unified description for the hardness in martensitic steels for a wide range of carbon contents is presented. It is based on describing the strength contributions of lath and plate martensite, precipitates and retained austenite. Descriptions of the dislocation density in both martensitic structures are obtained in terms of carbon content and tempering conditions. It is shown that a peak in hardness usually observed for carbon contents ranging 0.6–1 wt% is a result of a compromise between the strength of martensite, and the increase in retained austenite. A parametric analysis is performed suggesting possible scenarios for hardness improvement.

Martensitic steels are amongst the strongest materials for structural applications. This is attributed to their complex microstructure, which is highly affected by chemical composition. Two distinct martensitic structures are identified in low-alloy steels depending on carbon content [1]: Laths form in the compositional range 0–0.6 wt%, whereas plates become the dominant structure above 1 wt%; mixed lath and plate structures are present in the 0.6–1 wt% C range, where the fraction of plate martensite increases with increasing carbon content. Lath martensite consists of fine units (~100–300 nm thick) hierarchically arranged in substructures within the prior–austenite grains, namely packets and blocks of individual laths. These complex arrangements accommodate the crystallographic distortions during the transformation from austenite and ensuring that the net strain in the prior austenite grain is pure dilatation [2]. This structure is characterised by containing a high dislocation density and carbon redistribution at the lath boundaries. Conversely, plate martensite does not display an evident hierarchic structure [1,3], and plates appear as individual units of various sizes. These units are composed by a set of finely spaced transformation twins crossing throughout the plate (midribs) and dislocation arrays at the boundaries [4]. Additionally, Sherby et al. [5] have pointed out that the crystal structure transitions from being mainly cubic (BCC) to tetragonal (BCT) at the critical carbon content of $x_C^0 = 0.6$ wt%; plates start forming at this point. Additionally, the fraction of retained austenite increases significantly for carbon contents above this value, although thin films have also been found in steels with lower carbon content [6]. Fig. 1 shows a schematic representation of the martensitic structures for various carbon contents. It is well accepted that retained austenite aids in improving toughness and ductility, which content is highly affected by the processing conditions and carbon content. However, it also reduces the hardness or can lead to microstructural instabilities during tempering [7].

The variation in these microstructural features also reflects the wide spread in hardness of martensitic steels for various carbon contents [8]. For instance, the hardness increases monotonically in as–quenched conditions when increasing carbon content up to $\sim x_C^0$. This increment is directly related to the increase in dislocation density, grain boundary area and precipitation nucleation. However, pronounced variations in the hardness have been observed for higher carbon contents [1]. This is mostly due to the increase in the fraction of retained austenite and its variation with the quenching conditions [5]. For instance, Litwinchuk et al. [9] observed a peak in the hardness at a carbon content of ≈1 wt% in as–quenched conditions, further decreasing with increasing carbon content. However, with proper quenching and tempering conditions it is possible to increase the hardness up to 1100 HV by reducing the fraction of retained austenite and promoting precipitation hardening [5]. These results illustrate the significant challenges in prescribing the hardness of martensitic steels in terms of their initial microstructure and chemical composition.

The objective of this work is to postulate a model for describing the hardness evolution in martensitic Fe–C steels. This includes identifying how retained austenite affects the overall strength of the martensitic matrix and describing the strength of plate martensite stemming from dislocations at the plate interiors and midribs. Tempering conditions and chemical composition are included in

* Corresponding author.
E-mail address: eg375@cam.ac.uk (E.I. Galindo-Nava).

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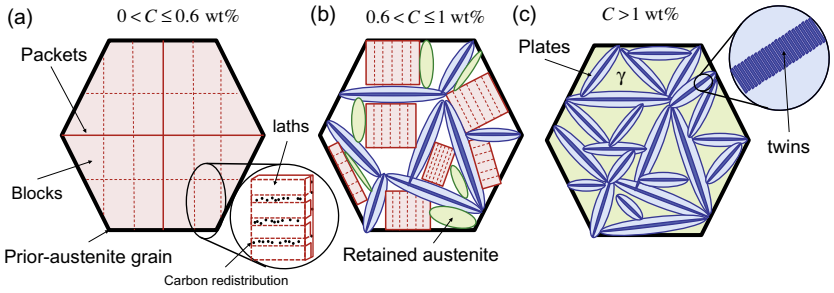

Fig. 1. (a) Lath martensite with hierarchically arranged blocks and packets produced by carbon redistribution; (b) mixed structure of laths, packets and retained austenite; and (c) plate martensite containing plates with midribs (twins) and a considerable fraction of retained austenite.

the evolution of the dislocation density. A model describing the microstructure and strength evolution of lath martensite has been introduced in previous work [10]. This article further extends the theory to define a unified approach for steels with carbon content in the range 0–2 wt%, where both lath and plate martensite feature. It is shown how these microstructures affect the hardness and a parametric analysis is performed to suggest possible scenarios for hardness improvement.

The yield strength of martensitic steels mainly stems from four contributions [1]: (1) solid solution σ_{ss}, (2) lath/plate strength σ_{Mart}, (3) precipitation hardening σ_p and (4) retained austenite. Most precipitation occurs where carbon partitions at dislocations, lath/plate boundaries or midribs [11]; this suggests that the strength of martensite structures and the precipitation hardening contributions act in a combined form [12]: $\sqrt{\sigma_{Mart}^2+\sigma_p^2}$. Additionally, the strengthening effects of laths/plates and precipitates are effective only in the areas where martensite forms, *i.e.* in the volume fraction $(1-f_\gamma)$, where f_γ is the volume fraction of retained austenite, and the strengthening contribution of the previous decreases according to $(1-f_\gamma)$ [1,8] (item 4). The effective strength of γ is $f_\gamma\sigma_\gamma$, where σ_γ is the yield stress of austenite. However, in practice it is difficult to measure σ_γ in martensitic steels. Nevertheless, σ_γ is much lower than the strength of the matrix ($f_\gamma\sigma_\gamma \ll \sigma_{Mart}$) [1] and it can be assumed that $f_\gamma\sigma_\gamma \sim 0$. Since most of the measurements have been reported in terms of the Vickers hardness H_v and the previous strengthening contributions are defined in terms of the yield stress σ_Y, the following equation for martensitic steels is employed to validate the strengthening mechanisms with experimental data [13]:

$$H_v = 0.4(\sigma_Y + 110) = 0.4\left(50+\sigma_{ss}+\left(1-f_\gamma\right)\sqrt{\sigma_{Mart}^2+\sigma_p^2}+110\right), \quad (1)$$

where the term 50 MPa is the lattice friction stress [10]. σ_{ss} is obtained with Fleischer's equation estimating the increment in the critical resolved–shear stress due to the presence of substitutional solute atoms [14]: $\sigma_{ss}=\sum_i(\beta_i^2x_i)^{1/2}$, where x_i is the atom fraction of alloying element i and β_i is a constant accounting for the local modulus and lattice distortions of element i with respect to pure iron. This formulation has been successfully applied to martensitic steels [10]. The Orowan–Ashby equation dictates the increase in the applied stress for dislocations to bypass fine carbides [10]: $\sigma_p = 0.26\frac{\mu b}{r_p}f_p^{1/2}\ln\left(\frac{r_p}{b}\right)$, where f_p and r_p are the volume fraction and mean radius of the carbide, respectively. σ_{Mart} depends on the relative fraction of lath and plate martensite and it can be described by a mixture rule:

$$\sigma_{Mart} = \sigma_{lath}f_{lath} + \sigma_{plate}(1-f_{lath}), \quad (2)$$

where σ_{lath} and σ_{plate} are the strength of lath and plate martensite, respectively, and f_{lath} is the volume fraction of lath martensite.

The strength of lath martensite is controlled by the increase in grain boundary area and dislocation density [10]. The former is expressed in terms of a Hall–Petch equation for the block size d_{block}, as it is considered as the "effective" grain size [15,10]. The dislocation density in the laths has been obtained by equating the dislocation energy at the lath boundaries and the lattice strain energy produced by carbon redistribution. σ_{lath} equals to:

$$\sigma_{lath} = \frac{300}{\sqrt{d_{block}}} + 0.25Mb\mu\sqrt{\rho_{lath}}$$
$$\rho_{lath} = \frac{3E}{(1+2\nu^2)\mu}\frac{4\varepsilon^2 d_{Cottrell}}{d_{lath}^2 b}, \quad (3)$$

where $M=3$ is the Taylor orientation factor, b = 0.286 nm is the magnitude of the Burgers vector, μ = 80 GPa is the shear modulus, E = 211 GPa is the Young's modulus, ν = 0.3 is the Poisson ratio, ε is the lattice strain produced by carbon redistribution, d_{lath} is the lath boundary thickness, and $d_{Cottrell}=7$ nm is the thickness of a Cottrell atmosphere. Details on ε estimation can be found in [10]. The block size is proportional to the prior–austenite grain size D_g according to: $d_{block}=0.067D_g$. d_{lath} has been proposed to be arranged in such form that it ensures complete carbon segregation to the lath boundaries and it equals to [10]: $d_{lath}=d_{Cottrell}(x_C^{\alpha'})^{-2/3}+\lambda_0 x_C^{\alpha'}\sqrt{D_{diff}t}$, where $x_C^{\alpha'}$ is the carbon atom fraction in the matrix, $\lambda_0=b/d_{Cottrell}$ is a constant accounting for a diffusion barrier for carbon atoms segregated into the Cottrell atmospheres, and $D_{diff}=6.2\times10^{-7}\exp\left(-\frac{80.000}{RT}\right)$ m^2/s is the diffusion constant of carbon in iron [8]. The first term in d_{lath} represents the lath thickness for as–quenched conditions and $x_C^{\alpha'}\sqrt{D_{diff}t}$ represents the mean carbon diffusion length during tempering.

The strength of plate martensite is dictated by grain boundary strengthening, high dislocation density and transformation twins [1]. However, as opposed to the hierarchically arranged structure of lath martensite, the effect of these features in the overall strength of low–alloy steels is less understood. For instance, although the apparent twin density at the midribs increases with carbon content [4,16], it is not evident how plate refinement affects the strength. Nevertheless, it has been observed that the plate size decreases when decreasing the prior–austenite grain size, whilst preserving its morphology and aspect ratio [17,18]. This suggests that the grain boundary strengthening produced by the plate size is to some extent analogous to the Hall–Petch equation in lath martensite when considering the prior–austenite size: $\frac{600}{\sqrt{D_g}}$; this value is obtained from [19].

The midribs mark the starting point of the transformation, where they grow by the formation of mechanical twins [4]. Carbon atoms redistribute mostly within the twins [20]; this mechanism has been observed directly by atom–probe

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