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A model for the microstructure behaviour and strength evolution in lath martensite

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ABSTRACT

A new model describing the microstructure and strength of lath martensite is introduced. The packet and block size were found to linearly depend on the prior-austenite grain size when introducing relevant crystallographic and geometric relationships of their hierarchical arrangements. A mechanism for the lath boundary arrangement within a block is postulated to ensure complete carbon redistribution to the lath boundaries. Accordingly, the dislocation density is obtained by considering the lattice distortion energy within a lath being equal to the strain energy of the dislocation density at the lath boundaries. Tempering effects are introduced by estimating the extent of carbon diffusing away from the lath boundaries; this mechanism relaxes the Cottrell atmospheres of lath dislocations and coarsens the boundaries. The yield stress as well as the microstructure evolution during tempering are successfully predicted by combining these results. The model is further extended to describe the yield stress in dual-phase steel microstructures by employing the iso-work principle. The model predictions are validated against experimental data in seven martensitic and five dual-phase steels, where the prior-austenite grain size, carbon content, tempering conditions and martensite volume fraction are employed as input. These results cover wide composition, initial microstructure and tempering conditions.

1. Introduction

Lath martensite is one of the most important phases in low and medium carbon steels for high-strength applications. During quenching, the transformation from austenite to martensite produces high shear strains from the supersaturated carbon content in the matrix and thin laths form in order to minimise the strain energy [1,2]. This process is characterised by inducing a high dislocation density, being this one of the main contributors to the high strength of martensite [3,4]. The transformation process also results in a complex microstructure, consisting of hierarchically arranged substructures within the prior-austenite grains, namely packets and blocks of individual laths.

Extensive experimental characterisation has been performed to aid unravelling the complex nature of lath martensite. For instance, several authors have employed electron-backscattering diffraction measurements to identify the microstructural and crystallographic features of martensite [2,5–8]. They have concluded that the hierarchical arrangement of packets/blocks and lath boundaries allows to accommodate the crystallographic shears of the transformation; this ensures that the net strain in the prior austenite grain is pure dilatation. Additionally, experimental evidence has shown that the prior-austenite grain size has a strong effect on the scale of packets and blocks and almost negligible effects on the lath substructure development [9–12]. This implies that the microstructure in the lath boundaries is practically independent of the prior-austenite grain size.

Atom-probe tomography has shown that carbon atoms tend to segregate towards the lath boundaries [13]. It has been suggested in low-carbon steels that Cottrell atmospheres form at the lath boundaries for as-quenched conditions [14,15]; further carbon redistribution occurs during tempering forming precipitates at the boundaries [13]. Chang et al. [16] observed in two steels containing 0.45 and 0.85 at.% carbon that the matrix was almost depleted in carbon, whilst carbon-rich regions extending for radial distances of at least 3 nm could be seen from the dislocation lines [17]. Wilde et al. [14] extended this work to quantify the carbon atom distribution and concentration within the atmospheres; they found the carbon saturation in the atmosphere to be at about 7–8 at.%. Sherman et al. [18] have attributed in a medium carbon steel (1.31 at.% carbon) the formation of Cottrell atmospheres in the vicinity of the dislocation cores in the martensite matrix, and that approximately 90% of carbon atoms are located within the range of the atmospheres. The remaining carbon was located in

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thin films of retained austenite. Interestingly, the lath boundaries show a carbon peak concentration of $\sim$7 at.% in the specimen with the highest cooling rate in as-quenched conditions; this value corresponds to the concentration saturation of a Cottrell atmosphere. TEM measurements showed no clear images or diffraction patterns of either austenite or carbides at the lath boundaries for this case. Similarly, Miller et al. [13] measured in Fe–Ni–C lath martensite for as-quenched conditions a peak concentration at the lath boundaries of 7.8 at.%. One aim of this work is to employ these results to propose a model being able to describe the structure of lath martensite, and to relate it to its mechanical properties in as-quenched and tempered conditions.

There are a number of strengthening mechanisms in the martensitic structure, including high-angle grain boundary strengthening [9], increase in the dislocation density [2], and carbide precipitation during tempering [4]. Although the occurrence of these mechanisms is well accepted, there are limited microstructure-sensitive models linking the evolution of the martensitic structure and its respective strength for as-quenched and tempered conditions; this can be due to the fact that models for describing the dislocation density as a function of carbon content and processing parameters are practically non-existent. This is also reflected by the fact that existing modelling approaches for martensite strengthening and dual-phase steels require the “calibration” of the dislocation density to reproduce the experiments [19–21]. This in turn leads to inaccurate estimations on the individual effects of the respective strengthening mechanisms, and becomes more prominent when tempering effects are also introduced. For instance, some authors have considered in the yield strength equation a stress term of carbon acting as interstitial solid solution, as if the microstructure were ferritic [15,22]. However, as carbon strongly segregates to the lath boundaries and most likely forms Cottrell atmospheres, this suggests that the carbon content is mostly related to the dislocation density increment rather than solid solution strengthening [23].

The objective of this work is to introduce a model for describing the microstructure and strength evolution of lath martensite and to validate it with low and medium carbon martensitic and dual-phase steels. This includes postulating: (1) crystallography-based relationships interlinking the prior-austenite grain, packet and block boundaries; (2) physics-based expressions linking carbon content with the dislocation density and lath boundary width; (3) predicting the yield strength evolution of martensitic and dual-phase steels as a function of the internal microstructure and carbon content; and (4) describing carbon diffusion effects during tempering and the respective recovery effects. Possible scenarios for alloy and heat treatment design to increase the yield strength are also explored. It is demonstrated that the complex martensitic hierarchical structure can be described by the present model by employing as input parameters the prior-austenite grain size, carbon content and tempering conditions. A summary of the main equations for the strength and microstructure in martensite is shown in Appendix A.

2. Materials and processing conditions

Table 1 shows the chemical composition and denomination of the materials studied in this work; all experimental results have been obtained from the literature. They cover seven martensitic and five dual-phase steel families. In most cases, low alloying additions of up to six substitutional elements are present. Contributions from other elements are neglected.

Table 2 shows the initial microstructure and tempering conditions for all steels considered here. t and T stand for the tempering time and temperature, respectively; D_g stands for the prior-austenite grain size; r_p and f_p are respectively the mean radius and volume fraction of the carbides forming during tempering; and V_f is the martensite volume fraction of the dual-phase steels. It is worth noting that for Mart1 and Mart5, no precipitation characterisation is shown since only the microstructure is described and no yield strength is predicted. Additionally, no prior-austenite grain size and carbide characterisation were reported for Mart6, and these values were assumed to be consistent with typical tempering conditions [4]. In the case of DP1, DP2 and DP3, no carbide characterisation was reported, and fixed volume fraction and particle size were assumed in all cases to be consistent with the experimental characterisation in pure martensitic steels, as it has been shown that tempering in dual-phase steels display similar precipitation kinetics than pure martensitic steels [24].

3. Microstructure evolution modelling

This section presents the theoretical aspects of the model to describe the packet/block and lath size, as well as the dislocation density as a function of the prior-austenite grain size, carbon content and tempering conditions. This work is defined upon four main assumptions to simplify the analysis:

(*i*) It is assumed that martensite has been fully formed at room temperature after quenching. No cooling rate and M_s effects are considered to affect the initial microstructure.

(*ii*) No precipitation kinetics during tempering is described and the volume fraction and mean carbide radius are obtained directly from the experiments.

(*iii*) Only carbon is assumed to control martensite’s behaviour and to fully segregate to the lath boundaries. No other compositional effects in the martensitic structure evolution are considered.

Table 1
Chemical composition (in wt%) of the steels tested in this work.

Steel	C	Si	Mn	Cr	Al	Ni	Mo	Co	W	Ref.
Mart1	0.076	–	–	8.83	–	–	–	2.99	3.11	[65]
Mart2	0.12–0.48	0.23–1.25	0.77–1.7	0–1.31	0.04–0.13	0–0.81	–	–	–	[15]
Mart3	0.2	0.01	0–2	–	–	–	–	–	–	[9]
Mart4	0.25	0.26	0.5	1	0.03	0.04	0.75	–	–	[10,11,66]
Mart5	0.39	0.27	1.36	1.95	–	1	–	–	–	[12]
Mart6	0.4	0.27	0.69	0.79	0.03	1.78	0.26	-	-	[67]
Mart7	0.56	1.4–2.3	0.69–0.72	0.89	–	–	–	–	–	[25]
DP1	0.034–0.23	0.01	0.22–0.35	–	–	–	–	–	–	[60]
DP2	0.05–0.18	1.47	2	–	–	–	–	–	–	[68]
DP3	0.1	0.92	2	0.03	0.03	–	–	–	–	[55]
DP4	0.13	0.3	1.18	0.047	–	–	–	–	–	[69]
DP5	0.21	0.1	1.7	–	–	–	–	–	–	[61]