The effect of ultrafast heating rate on the elemental distribution between phases in a low carbon steel

M. A. Valdes-Tabernero, M. A. Monclus, I. Sabirov, J. M. Molina-Aldareguia, R. H. Petrov, J. Wang and I. Timokhina

IMDEA Materials Institute, Getafe, Madrid, Spain; Department of Electromechanical, Systems and Metal Engineering, Research Group Materials Science and Technology, Ghent University, Zwijnaarde, Belgium; Department of Materials Science and Engineering, Delft University of Technology, Delft, The Netherlands; Institute for Frontier Materials, Deakin University, Geelong, VIC, Australia

ABSTRACT

This work focuses on the effect of heating rate, i.e. conventional heating (CH) and ultrafast heating (UFH) rates on the elemental distribution between phases in a low carbon steel. Microstructural characterization was carried out using an Electron Backscatter Diffraction (EBSD) and Atom Probe Tomography (APT) technique. Nanohardness of individual microconstituents was measured. It is shown that the applied heat treatments result in the formation of multiphase microstructures consisting of a ferritic matrix with embedded martensite and retained austenite. The ferritic matrix of the CH material was fully recrystallized, whereas both recrystallized (coarser) and non-recrystallized (finer) ferritic grains were present in the matrix of the UFH material. APT analysis indirectly confirmed that recrystallized grains after both heat treatments have a lower carbon content, whereas the non-recrystallized grains after UFH have a higher carbon content. It correlates with the nanohardness results, i.e. non-recrystallized grains show higher hardness, while recrystallized grains have a lower hardness. The segregations of C and Mn atoms at the martensite/ferrite interface were observed after both treatments. It is hypothesized that the segregations are formed under negligible partitioning local equilibrium condition during CH, whereas the solute drag effect results in the formation of interface segregation during UFH.

ARTICLE HISTORY

Received 21 December 2021
Accepted 28 March 2022

KEYWORDS

Steel; ultrafast heating; atom probe tomography; elemental distribution; interface segregations
1. Introduction

Nowadays, steel manufacturers and the automotive industry face significant pressure to reduce energy consumption and carbon emission. Rapid manufacturing of advanced high-strength steels (AHSSs) showing enhanced mechanical performance can address both problems. Rapid manufacturing can dramatically reduce the energy costs during steel processing, while applying these AHSSs in cars will lightweight the parts, thus decreasing the CO\textsubscript{2} emissions. Ultrafast heating (UFH) of carbon steels has recently attracted significant attention from the steel research community, as it allows to process multiphase AHSSs with improved mechanical strength and performance within seconds (Castro Cerda et al., 2016; 2017; Lolla, Cola, Narayanan, Alexandrov, & Babu, 2011; Matlock, Kang, De Moor, & Speer, 2020; Petrov, Kestens, Kaluba, & Houbaert, 2003). These steels typically have a complex multiphase microstructure consisting of ferritic matrix with embedded martensitic grains and low fraction of retained austenite (Valdes-Tabernero, Celada-Casero, Sabirov, Kumar, & Petrov, 2019; Valdes-Tabernero et al., 2020). There are numerous reports focused on the microstructure evolution during UFH processing of AHSSs (Castro Cerda et al., 2016, 2017; Valdes-Tabernero, Celada-Casero, et al., 2019; Valdes-Tabernero et al., 2020). It was shown that the microstructure evolution during UFH process is a very complex phenomenon involving phase transformations, recovery, recrystallization, grain growth, and diffusion of alloying elements. Their kinetics strongly depend on the heat treatment parameters, where heating rate (Castro Cerda et al., 2016, 2017; Lolla et al., 2011; Matlock et al., 2020; Petrov et al., 2003), soaking time (Valdes-Tabernero, Celada-Casero, et al., 2019), and peak temperature (Valdes-Tabernero et al., 2020) play the key role. It was established that the UFH process of carbon steels (at \(\geq800^\circ\text{C/s}\)) with a short soaking time (\(\leq1.5\text{s}\)) results in a finer grain size and, therefore, in increased mechanical strength (Castro Cerda et al., 2016; Valdes-Tabernero, Celada-Casero, et al., 2019). This was related to shifting of the \(A_3\) and recrystallization temperatures to higher values, promoting nucleation rate of recrystallized nuclei and suppressing grain growth during such a rapid treatment (Castro Cerda et al., 2016; Valdes-Tabernero, Celada-Casero, et al., 2019).

It is well known that segregations of alloying elements can affect significantly the microstructure evolution during heat or thermo-mechanical treatment of steels, thus providing an extra tool for microstructural design (Raabe et al., 2013). Xie, Shang, Subramanian, Ma, and Misra (2017) studied the elemental distribution in a Fe–0.08C–0.5Si–2.4Mn–0.5Ni–0.42(Mo + Nb + Ti) (wt.%) steel subjected to a two-step thermal treatment, resulting in the formation of a multiphase
microstructure with 10% of film-type retained austenite. The atom probe tomography (APT) analysis at the austenite–ferrite interface revealed an enhanced concentration of C and Mn atoms in austenite, and Mn segregations were detected at the austenite/ferrite interface. Authors also employed kinetic calculations using the simulation software DICTRA to explain the phenomenon (Xie et al., 2017). According to the results, the austenite formation was divided into three main stages. In the first stage, the thickness of the austenite grains grew rapidly (within 0.1 s) under negligible partitioning local equilibrium (NPLE) conditions, as this stage was controlled only by diffusion of C. The latter rapidly diffused into austenite, and its concentration reached an equilibrium content of 0.02 wt.% in ferrite, and austenite was enriched by up to 0.7 wt.% of C. On the other hand, Mn could only segregate at the interface. In stage II, the austenite growth was slower due to the Mn partitioning from the matrix to the austenite, resulting in a concentration gradient in the interior of the austenite phase, whereas C concentration was nearly uniform in both phases. In stage III, Mn diffused along the austenite homogeneously distributing in the grain interior, accompanied by the shrinkage of the austenite grain size. Similar results were reported in Wei, Enomoto, Hadian, Zurob, and Purdy (2013) by Wei and co-authors during intercritical annealing of a quaternary system (Fe–0.1C–3Mn–1.5Si, wt.%) with initial martensite microstructure. In the latter study, the authors showed that the second stage could also be controlled by the diffusion of Si atoms. In Li et al. (2017), a pipeline Fe–0.07C–0.25Si–1.94Mn–0.081Nb–0.28Cr–0.26Mo–0.014Ti (wt.%) steel was analyzed by APT after welding. Partitioning of C and Mn into austenite/martensite from the surrounding ferritic matrix was observed, whereas other alloying elements, such as Cr or Mo, segregated only at the interface. While the distribution of Mn in the interior of the austenite/martensite and ferrite was homogeneous, C showed a large variation in concentration along the grain interior. These experimental works clearly demonstrate that the partitioning of alloying elements plays a critical role in the microstructure evolution during steel heat treatments. However, this phenomenon has never been studied in the steels subjected to heat treatments with ultrafast heating rates. The main objective of the present work is to explore the effect of alloying elements and their segregations (if present) during UFH processing of a low carbon steel on microstructural refinement, volume fraction and nanohardness of the phases.
2. Material and experimental procedures

2.1. Material and processing

A low carbon steel Fe–0.19C–1.61Mn–1.06Al–0.50Si (wt.%) or Fe–0.9C–1.6Mn–2Al–1Si (at. %) was chosen for this study. The material was supplied in the form of 1mm cold rolled (50% reduction ratio) strips with a ferritic–pearlitic microstructure. Flat specimens having a length of 90mm and a width of 10mm were machined along the rolling direction. A Gleeble 3800 thermo-mechanical simulator was used to simulate the heat treatments. In order to control the temperature, a K-type thermocouple was spot-welded to the midsection of each specimen. Ultrafast heating (UFH) treatments with a heating rate of 800 °C/s and conventional heating (CH) treatments with a heating rate of 10 °C/s were performed (Figure 1). In both cases, the samples were heated to the intercritical temperature of 860 °C and rapidly cooled to room temperature at ~160 °C/s (Figure 1). The soaking time at the peak temperature did not exceed 0.2s, so the effect of annealing time on the microstructure was suppressed.

2.2. Microstructural characterization

Quantitative microstructural characterization of the heat-treated samples was performed through electron backscatter diffraction (EBSD) analysis. Specimens were ground and polished to a mirror-like surface applying standard metallographic techniques with final polishing using OP-U (colloidal silica). The EBSD studies were performed using an FEI Quanta™ Helios NanoLab 600i equipped with a NordlysNano detector controlled

Figure 1. Schematic drawing of applied ultrafast heating (UFH) and conventional heating (CH) treatments.
by the AZtec Oxford Instruments Nanoanalysis (version 2.4) software. The data were acquired at an accelerating voltage of 18 kV, a working distance of 8 mm, a tilt angle of 70°, and a step size of 65 nm in a hexagonal scan grid. The orientation data were post-processed using HKL Post-processing Oxford Instruments Nanotechnology (version 5.1©) software and TSL Data analysis version 7.3 software. Grains were defined as a minimum of 4 pixels with a misorientation ≥ 5°. Grain boundaries having a misorientation ≥ 15° were defined as high-angle grain boundaries (HAGBs), whereas low-angle grain boundaries (LAGBs) had a misorientation < 15°. The volume fractions of martensite, retained austenite, and recrystallized/recovered ferritic grains were determined by a two-step partitioning procedure described in (Petrov & Kestens, 2015). In this procedure, grains with high and low grain average image qualities are separated in a first step, allowing to distinguish between transformed (martensite) and untransformed (ferrite) fractions. In the second step, recrystallized and non-recrystallized ferritic grains are separated using the grain orientation spread criterion: Grains with an orientation spread below 1° are defined as the recrystallized grains, while grains with an orientation spread above 1° are defined as the non-recrystallized ones (Cao et al., 2013). Grain size was evaluated via the calculated average grain diameter. The microstructure was observed on the plane perpendicular to the sample transverse direction (the RD–ND plane).

The specimens for atom probe tomography (APT) analysis of ferrite/martensite interfaces and ferrite and martensite matrixes were prepared by the focused ion beam (FIB) in situ lift-out procedure using an FEI Quanta 3D FEG–SEM (Miller & Forbes, 2014) (Figure 2). The APT study was performed on a CAMECA LEAP-4000HR in the voltage-pulsed mode.

Figure 2. (a) SEM image of the APT tip and (b) microstructural region from which the APT samples were prepared. F is ferrite, M is martensite.
with 20% pulse fraction, and 200 kHz pulse rate at −213 °C. IVAS 3.6.12 software was used to reconstruct the APT data.

### 2.3 Nanomechanical characterization

A Hysitron TI950 Triboindenter with a Berkovich tip was employed for nanoindentation testing on the individual microconstituents. First, square areas having a size of ~10 × 10 µm² were analyzed by EBSD, and individual microconstituents were determined. At least ten areas were tested for each material’s condition. Nanoindentation tests were carried out in displacement control mode at a constant strain rate ($\dot{\varepsilon} = \dot{\varepsilon}/h$) of 0.07 s⁻¹, where $h$ is the penetration depth and $\dot{h}$ the penetration rate of the indenter. At least 20 indents were performed on each phase at an imposed maximum depth of 150 nm. The nanohardness was determined from the analysis of the load–displacement curves using the Oliver and Pharr method (Oliver & Pharr, 1992).

### 3. Results

Figure 3 illustrates the microstructure of the steel after CH (a,c) and UFH (b,d) treatments. It is seen that both heat treatments lead to a complex multiphase microstructure consisting of a ferritic matrix with embedded martensite and retained austenite. Table 1 outlines the average grain size, and the volume fraction of each microconstituent. The fraction of martensite decreases with increasing heating rate from 10.6% to 6.9%. The volume fraction of retained austenite is low after both treatments (2.2–3.1%). The UFH treatment results in a finer grain size for both ferrite and martensite. The ferrite in the CH samples has equiaxed recrystallized morphology with homogenous distribution of ferrite grain size from 4 to 8 µm (Figure 3). In contrast, ferrite in the UFH samples presents two morphologies: recrystallized grains and non-recrystallized (i.e. recovered) grains having a wide grain size distribution (Figure 3e). The volume fraction of recrystallized ferrite is nearly 50%. The effect of heating rate on microstructure evolution of the present steel was thoroughly discussed in our recent article (Valdes-Tabernero, Celada-Casero, et al., 2019).

| Condition | Ferrite | Martensite | RA |
|-----------|---------|------------|----|
|           | Volume fraction [%] | Grain size [µm] | Volume fraction [%] | Grain size [µm] | Volume fraction [%] | Grain size [µm] |
| CH        | 86.3    | 3.4        | 10.6    | 2.1        | 3.1        | 0.6        |
| UFH       | 90.9    | 0.6        | 6.9     | 1.0        | 2.2        | 0.5        |
Figure 3. (a, b) SEM images of the microstructure after CH (a) and UFH (b) treatment; (c, d) Image quality + phase map obtained from EBSD analysis of the CH (c) and UFH (d) samples; (e) ferrite grain size distribution in the CH and UFH treated material.
The APT samples were taken across ferrite/martensite interfaces and from the ferrite and martensite matrixes. The phases were identified based on the content of C and Mn obtained from the APT atom maps and matrix composition analyses. The distribution of solutes across different interfaces was analyzed by 1-D concentration profiles for both CH and UFH conditions.

There are three phases identified after CH treatment: ferrite, bainitic ferrite and martensite/RA (it is not possible to discriminate RA from martensite by APT since RA transforms to martensite at the temperature of probing, –213°C) (Table 2; Figures 4 and 5). It should be noted that bainitic ferrite could not be identified by SEM/EBSD in the earlier works due to its very low volume fraction and nanosize. Ferrite has the equilibrium C content of 0.05±0.01 at. % and Mn content of 1.4 at. % (Table 2). Martensite/RA exposes the average C content of 3.3±0.3 at. % and is enriched by Mn (3±0.5 at. %), which was expected since the latter is an austenite stabilizer (Table 2). The APT also indicates the presence of bainitic ferrite with an average C content of 0.1 at. % located between ferrite and martensite/RA (Table 2; Figure 4). Its formation could be explained by the slow heating rate during the CH treatment allowing a long distance for C diffusion from austenite/ferrite interface into austenite. This led to the formation of the remaining austenite with different carbon contents. The remaining austenite with higher carbon content transformed to martensite during final quenching, while the remaining austenite with lower carbon content was decomposed with formation of bainitic ferrite with carbon content of 0.1±0.02 at. %. at the temperature higher than martensite transformation temperature during final cooling. Moreover, Mn does not segregate at bainitic ferrite/martensite interfaces, whereas C

| Table 2. Content of alloying elements (at. %) in the microstructural constituents detected by the APT technique. |
|-----------------------------------------------|
| Phase                          | C (at. %) | Mn (at. %) | Al (at. %) | Si (at. %) | Cr (at. %) | Ti (at. %) |
|-----------------------------------------------|
| CH Ferrite                      | 0.05±0.01 | 1.4±0.1    | 1±0.3      | 0.3±0.07   | ~0.04      | ~0.04      |
| Martensite/RA                   | 3.3±0.3   | 3±0.5      | 2.5±0.1    | 0.95±0.05  | ~0.03      | ~0.04      |
| Bainitic ferrite                | 0.1±0.02  | 2.1±0.1    | 1.7±0.5    | 0.9±0.03   | ~0.04      | ~0.04      |
| Ferrite/martensite interface    | 4.4±0.2   | 5.5±0.1    | 2±0.1      | 1.1±0.1    | –          | –          |
| UFH Recrystallized ferrite      | 0.02±0.01 | 1.4±0.1    | 1±0.2      | 1.05±0.01  | ~0.02      | ~0.04      |
| Recovered ferrite               | 0.08±0.01 | 1.4±0.1    | 1±0.2      | 0.95±0.03  | ~0.02      | ~0.03      |
| Martensite/RA                   | 3.8±1     | 2.5±0.2    | 2.5±0.1    | 1.23±0.1   | ~0.02      | ~0.03      |
| Ferrite–martensite interface    | 5±0.2     | 5±0.2      | 1±0.5      | 1.1±0.1    | ~0.02      | ~0.04      |
shows a prominent spike of 5 at. % at the interface (Figure 4). The C maps of martensite confirmed its two types of C distribution: (i) homogeneous (Figure 5a) and (ii) inhomogeneous (Figure 4a) where C decorated microstructural features that are believed to be dislocations. Segregation of C at the dislocations observed in some of martensite grains could occur in austenite grains near ferrite grains, which were formed at the beginning of austenite to ferrite transformation. It enables carbon to diffuse into remaining austenite grain interior and segregate at the dislocations.

The concentration profile across martensite/ferrite interfaces after CH treatment exposed distinct 5.5 at. % Mn peak at the interface and another C peak of 4.4 at. % inside the martensite grain at the distance of 15 nm from the interface (Figure 5).

APT analysis of the sample after UFH treatment confirms existence of two types of ferrite grains with higher (~0.08 at. %) and lower (~0.02 at. %) C contents (Table 2). The ferrite grains with lower C content may correspond to the recrystallized ferrite grains detected by EBSD analysis, whereas ferrite grains with higher C content can be linked with the non-recrystallized ferrite grains (Figure 3). Martensite after UFH has slightly higher C content of 3.8 at. % compared to the one after CH, most likely due to the absence of C segregations at dislocations.
distribution APT analysis of martensite islands demonstrated that the distribution of C does not deviate from the corresponding random distribution that indicates the absence of carbon clusters in martensite (Figure 6c).

The composition profiles across ferrite/martensite interfaces of the UFH sample showed the distinct peaks of Mn and C at the interface (Figure 6b). The spikes have a sharper shape compared to the peaks after CH treatments. Moreover, the Mn segregation at the ferrite/martensite interface after CH treatment (5.5 ± 0.1 at. %) was slightly higher than after UFH (5 ± 0.2 at. %), whereas the amount of C atoms segregated at the interface after CH was lower (4.4 ± 0.2 at. %) than after UFH (5 ± 0.2 at. %) (Table 2; Figures 4–6). The segregation of other alloying elements, such as Si and Al, was negligible.

The nanohardness values for the recrystallized and non-recrystallized ferrite grains after CH and UFH heat treatments are summarized in Table 3. The nanohardness of recrystallized ferrite was similar after both treatments, whereas the nanohardness of non-recrystallized ferrite was higher than the nanohardness of recrystallized ferrite. This can be associated with the higher C content measured in recovered ferrite, as well as its enhanced dislocation density (Valdes-Tabenero, Celada-Casero, et al., 2019).

The average nanohardness value of martensite formed after CH treatment was higher than after UFH treatment, most likely due to the segregation of C atoms at the dislocations present in the martensite of the CH treated samples leading to the hardening effect of this martensite.
4. Discussion

From these experimental results, it is clearly seen that the heating rate has a significant effect on the microstructure (on micro- and atomic levels) and properties of the individual microconstituents formed in the studied alloy. During the CH treatment, the slow heating controlled the phase transformations, resulting in a more homogenous C distribution over microconstituents. The ferrite grains after the CH treatment were fully recrystallized, and the C content of ferrite in these samples was homogenous and at the equilibrium composition (0.05 ± 0.01 at. %) (Table 2). All ferrite/martensite interfaces analysed by APT in the CH samples showed segregations of C and Mn. Since at slow heating, the ferrite to austenite transformation is fully accomplished by diffusion, the

---

**Figure 6.** Representative APT analysis of the phases formed after UFH: (a) 5 at. % C and 5 at. % Mn iso-surface atom maps; (b) 1-D concentration profile across interface shown by arrow in (a); (c) nearest C neighbor distribution graph of martensite. F is ferrite, M is martensite, dash line represents the interface position.

| Heating rate               | CH       | UFH     |
|----------------------------|----------|---------|
| Recrystallized ferrite (GPa) | 2.7 ± 0.2| 2.6 ± 0.1|
| Non-recrystallized ferrite (GPa) | n/a| 3.2 ± 0.2|
| Martensite (GPa)          | 8.1 ± 1.0| 7.6 ± 2.4|

**Table 3.** Nano-hardness of microconstituents after CH and UFH heat treatments.
segregation at the interfaces could be explained by the mode of phase equilibria at the migrating interface, such as negligible partitioning local equilibrium (NPLE) and para-equilibrium (PE) (Agren, Brechet, Hutchinson, Philibert, & Purdy, 2016). The PE model represents a limiting case, where the substitutional atoms are completely immobile (unlike interstitials), and the compositional ratio of substitutional solute and solvent elements remains constant across the interface (Agren et al., 2016). In the NPLE model, even though long-range diffusion is not possible, the substitutional atoms can still move in the vicinity of the interface and build up a solute spike in front of the interface as the interface passes through (Agren et al., 2016). The NPLE condition could occur at the slow interfacial velocity, which appeared to happen during CH treatment, when the transformation was controlled by slow heating, and the diffusion and segregation of C (up to 4.4 at. %) and Mn (up to 5.5 at. %) atoms were enabled by the slow mobility of the interface during phase transformation (Table 2; Figures 4 and 5). Moreover, slow heating stimulated the diffusion of carbon in the remaining austenite and its segregation at the dislocations (Figure 4).

The SEM–EBSD analysis indicates the formation of two types of ferrite grains after UFH treatment: recrystallized with the average nano-hardness of 2.6 GPa and non-recrystallized with the average nano-hardness of 3.2 GPa (Tables 1 and 3). It is well correlated with the C content of ferrite grains analysed by APT, i.e. some grains have a C content of 0.08 ± 0.01 at. % (non-recrystallized), and the others of 0.02 ± 0.01 at. % (recrystallized) (Table 2). The difference in the ferrite grain morphologies could be associated with inhomogeneous redistribution of carbon in austenite due to the UFH treatment of ferrite/pearlite initial microstructure. Interestingly, the recrystallized grains formed after CH showed a similar range of grain size, C content (0.05 ± 0.01 at. %) and nanohardness level (2.7 GPa) (Tables 1–3). APT analyses of the interfaces formed after UFH treatment also showed the segregation of C and Mn at the ferrite/martensite interfaces. However, the C peak was more prominent (5 at. %), and the segregation of Mn was less prominent (5 at. %) compared to the interfaces formed after CH treatment (Table 2). These observations cannot be explained based on the equilibrium state at the interface (PE, NPLE) during the UFH process, since (unlike in the CH case) the diffusion of Mn is suppressed. Instead, we can use the concept of the solute drag (Hillert, 1999). The solute drag model is based on the dissipation of the Gibbs free energy by substitutional solute atoms that move by short-range diffusion across the interface and the possible interaction between substitutional atoms and the migrating interface (Clark, Janik, Lan, & Sridhar, 2017; Hillert, 1999). Potentially the solute drag effect can explain the segregation of Mn, which can alter the interface mobility.
Since the phase transformations occurred at a very high rate and Mn was inhomogeniously distributed between the parent phases (ferrite and pearlite), different amounts of Mn can segregate at the different interfaces changing their mobility, which leads to the formation of two types of ferrite grains (i.e. recrystallized and non-recrystallized) with different size, carbon content and nanohardness (Tables 1–3). The amount of elements (C, Mn) segregated at the interfaces after CH and UFH mentioned above can indirectly confirm this hypothesis. It should also be noted that C did not show any segregation at dislocations or C clustering in martensite, which can occur due to the pipe diffusion along the dislocation core. This can also be correlated with the inhibition of the carbon diffusion during UFH process.

**Conclusions**

The effect of heating rate (ultrafast (UFH) vs. conventional (CH)) on the microstructure of a Fe–0.19C–1.61Mn–1.06Al–0.50Si (wt.%) steel was studied on the micro- and atomic scales and was correlated with the nanohardness of individual microconstituents. The following conclusions can be drawn based on the analysis of the obtained experimental results.

1. A complex multiphase microstructure consisting of ferritic matrix with embedded martensite and retained austenite grains is formed after both CH and UFH treatments. The ferritic matrix of the CH treated material is fully recrystallized, whereas both recrystallized and non-recrystallized ferrite grains are observed in the UFH treated material. Additionally, APT reveals nanosized areas of bainitic ferrite at the martensite phase boundaries in the CH material. The formation of the latter is related to the slow heating rate allowing a long distance for C diffusion from austenite/ferrite interface into austenite. UFH process results in a finer microstructure compared to the CH treatment.

2. There are segregations of C and Mn atoms at the martensite/ferrite interface after both CH and UFH treatments. The Mn atoms do not segregate at martensite/bainitic ferrite interfaces, whereas C shows a prominent spike. These segregations are formed under NPLE condition during CH treatment, whereas the solute drag effect controls their formation during UFH treatment.

3. It appeared that the higher nanohardness of the non-recrystallized ferritic grains after UFH compared to the recrystallized ferritic grains for both UFH and CH can be correlated with their higher C content and, in turns, higher dislocation density.
Acknowledgements

Deakin University’s Advanced Characterization Facility is acknowledged for use of the CAMECA LEAP 4000 HR.

ORCID

J. M. Molina-Aldareguia  
http://orcid.org/0000-0003-3508-6003

References

Agren, J., Brechet, Y., Hutchinson, C., Philibert, J., & Purdy, G. (2016).  
Thermodynamics and phase transformations: The selected works of Mats Hillert.  
France: EDP Sciences.

Cao, Y., Di, H., Zhang, J., Zhang, J., Ma, T., & Misra, R. D. K. (2013). An  
electron backscattered diffraction study on the dynamic recrystallization behavior of a nickel–chromium alloy (800H) during hot deformation.  
Materials Science & Engineering: A, 585, 71–85. https://doi.org/10.1016/j.msea.2013.07.037.

Castro Cerda, F. M., Vercruysse, F., Minh, T. N., Kestens, L. A. I., Monsalve, A., & Petrov, R. H. (2017). The effect of heating rate on the recrystallization behavior in cold rolled ultra-low carbon steel. Steel Research International, 88(1), 1600351. https://doi.org/10.1002/srin.201600351.

Clark, S., Janik, V., Lan, Y., & Sridhar, S. (2017). Interphase precipitation: An interfacial segregation model. ISIJ International, 57(3), 524–532.

Hillert, M. (1999). Solute drag, solute trapping and diffusional dissipation of Gibbs energy. Acta Materialia, 47(18), 4481–4505. https://doi.org/10.1016/S1359-6454(99)00336-5.

Li, X., Shang, C., Ma, X., Gault, B., Subramanian, S. V., Sun, J., & Misra, R. D. K. (2017). Elemental distribution in the martensite–austenite constituent in intercritically reheated coarse-grained heat-affected zone of a high-strength pipeline steel. Scripta Materialia, 139, 67–70. https://doi.org/10.1016/j.scriptamat.2017.06.017.

Lolla, T., Cola, G., Narayanan, B., Alexandrov, B., & Babu, S. S. (2011). Development of rapid heating and cooling (flash processing) process to produce advanced high strength steel microstructures. Materials Science & Technology, 27(5), 863–875. https://doi.org/10.1179/174328409X433813.

Matlock, D. K., Kang, S., De Moor, E., & Speer, J. G. (2020). Applications of rapid thermal processing to advanced high strength sheet steel developments. Materials Characterization, 166, 110397. https://doi.org/10.1016/j.matchar.2020.110397.

Miller, M. K., & Forbes, R. G. (2014). Atom-probe tomography. Boston, MA, USA: Springer.

Oliver, W. C., & Pharr, G. M. (1992). An improved technique for determining hardness and elastic modulus using load and displacement sensing indentation
experiments. *Journal of Materials Research*, 7(6), 1564–1583. https://doi.org/10.1557/JMR.1992.1564.

Petrov, R. H., & Kestens, L. A. I. (2015). Advanced high-strength steels: Electron backscatter diffraction (EBSD). *Encyclopedia of Iron, Steel, & Their Alloys*, 46–69. https://doi.org/10.1081/E-EISA-120050786.

Petrov, R., Kestens, L., Kaluba, W., & Houbartaert, Y. (2003). Recrystallization and austenite formation in a cold-rolled TRIP steel during ultrafast heating. *Steel Grips*, 4, 289–294.

Raabe, D., Sandlobes, S., Millan, J., Ponge, D., Assadi, H., Herbig, M., & Choi, P. P. (2013). Segregation engineering enables nanoscale martensite to austenite phase transformation at grain boundaries: A pathway to ductile martensite. *Acta Materialia*, 61(16), 6132–6152. https://doi.org/10.1016/j.actamat.2013.06.055.

Valdes-Tabernero, M. A., Celada-Casero, C., Sabirov, I., Kumar, A., & Petrov, R. H. (2019). The effect of heating rate and soaking time on microstructure of an advanced high strength steel. *Materials Characterization*, 155, 109822. https://doi.org/10.1016/j.matchar.2019.109822.

Valdes-Tabernero, M. A., Kumar, A., Petrov, R. H., Monclus, M. A., Molina-Aldareguia, J. M., & Sabirov, I. (2020). The sensitivity of the microstructure and properties to the peak temperature in an ultrafast heat treated low carbon steel. *Materials Science & Engineering: A*, 776, 138999. https://doi.org/10.1016/j.msea.2020.138999.

Valdes-Tabernero, M. A., Petrov, R. H., Monclus, M. A., Molina-Aldareguia, J. M., & Sabirov, I. (2019). The effect of soaking time after ultrafast heating on the microstructure and mechanical behavior of a low carbon steel. *Materials Science & Engineering: A*, 765, 138276. https://doi.org/10.1016/j.msea.2019.138276.

Wei, R., Enomoto, M., Hadian, R., Zurob, H. S., & Purdy, G. R. (2013). Growth of austenite from as-quenched martensite during intercritical annealing in an Fe-0.1C-3Mn-1.5Si alloy. *Acta Materialia*, 61(2), 697–707. https://doi.org/10.1016/j.actamat.2012.10.019.

Xie, Z. J., Shang, C. J., Subramanian, S. V., Ma, X. P., & Misra, R. D. K. (2017). Atom probe tomography and numerical study of austenite stabilization in a low carbon low alloy steel processed by two-step intercritical heat treatment. *Scripta Materialia*, 137, 36–40. https://doi.org/10.1016/j.scriptamat.2017.05.002.