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Effect of ausforming on the anisotropy of low temperature bainitic transformation☆

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ABSTRACT

Ausforming is the process by which austenite is plastically deformed prior to bainitic or martensitic transformation. The advantages of such a process on the kinetics of the bainitic transformations have already been reported, but little is found about the effect on the morphology and crystallography of the final microstructure, and in most cases the studies deal with high C steels. Since ausforming has been suggested to be an alternative to transfer the concept of nanostructured bainite to medium carbon contents, the necessity to study the effect of prior plastic deformation on the subsequent bainitic transformation in those steels arises. By means of techniques such as high-resolution dilatometry, scanning electron microscopy, electron backscatter diffraction and X-Ray diffraction, it has been proven that, in a medium carbon steel (0.4 wt%), the macroscopic isotropy that characterizes the bainitic transformation can be altered, under certain conditions, by plastically deforming austenite prior to transformation. The alteration is such that, at low temperatures, the interpretation of the dilatometric response or the plate thickness measurements can be hindered. In addition, the final microstructure is also affected, presenting highly ordered bainitic ferrite plates, most of them correspondent to specific crystallographic variants.

1. Introduction

Ausforming is a thermo-mechanical treatment consisting in plastically deforming a fully austenitized steel, followed by transformation to either bainite or martensite [1]. When it comes to bainitic transformations, depending on the temperature at which austenite is deformed (T_{def}), three different ranges can be identified, all of them below the non-recrystallization temperature (TNR): High temperature (HT) ausforming, with T_{def} right below TNR; medium temperature (MT) ausforming, with T_{def} in the bay between the ferrite/pearlite and the bainite regions in the TTT diagram, i.e. the so-called hiatus, in which no isothermal transformation is expected; and low temperature (LT) ausforming, with T_{def} at the same temperature as, or close to, the isothermal holding [2–4].

Among other expected benefits, previously deformed austenite presents a decrease in the martensitic transformation temperature [5–12], Ms., which in turn allows performing isothermal treatments at lower temperatures. Secondly, as austenite is strengthen by work hardening, the concomitant increase in dislocation density is later inherited by bainitic ferrite during transformation [13,14]. In addition, given that bainite growth is displacive and there is a significant strain associated to it, the transformation product is constrained to be in the form of a thin plate, which allows the minimization of the strain energy when the plate is elastically accommodated. Therefore, it is not strange that the yield strength of the parent austenite plays a key role in defining the final size of the bainitic ferrite plates [15,16]. In summary, the expected increase in dislocation density and also the refinement of the microstructure are paramount in the control of the final strength of bainitic microstructures [17].

However, some drawbacks must be taken into account. First, plastic deformation increases the grain boundary area per volume unit and the dislocation density, and thus the amount of nucleation sites of bainite; therefore the incubation time may be shortened [18–25]. But in the same fashion, the reconstructive transformation to ferrite is also accelerated. For that reason, only steels with a high enough hardenability would be suitable for ausforming treatments [1]. Second, the so-called

☆ The raw/processed data required to reproduce these findings cannot be shared at this time as the data also forms part of an ongoing study.

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austenite mechanical stabilization, which is the term used when referring to the impossibility of bainitic ferrite to form from plastically deformed austenite [26–29]. This occurs because the displacive transformation of austenite into bainitic ferrite involves translation of glissile interfaces, whose movement can be retarded, and even impeded, by the defects introduced by plastic deformation [29]. There is an upper strain limit beyond which no bainitic ferrite transforms from the parent austenite, the so-called critical strain, ε_c. Therefore, when selecting the ausforming parameters such deformation levels must be considered and avoided under all circumstances.

LT ausforming has been proved to have plenty of advantages over higher temperature ones, i.e., HT and MT, allowing the bainitic transformation to be accelerated in a higher degree [23,25] and improving the final mechanical properties [14,30,31]. It is also being currently studied as a way of transferring the concept of nanostructured bainite to lower carbon contents, i.e. 0.4–0.5 wt% [2].

Nonetheless, most of the studies have not focused on other aspects than those previously mentioned, evading possible effects of the transformation on the microstructure anisotropy. Gong et al. [19,32] elucidated that deforming austenite at low temperature could lead to the variants selection phenomenon, which could explain the blocky austenite reduction and the highly ordered microstructures observed in some other works [21,24,30,33]. There is not an extensive body of literature on the topic though, and most of the studies focus on high carbon steels [19,32,34]. In the present work, in order to bear in mind possible differences that could come up when performing ausforming treatments in medium carbon steels, the effect of the low temperature deformation on the final bainitic microstructure has been studied in a medium C, Si rich steel. The results obtained from the LT ausforming treatment have been compared with those obtained by a MT ausforming route and a treatment in the absence of prior deformation, the latter referred to as pure isothermal in the following sections.

2. Experimental

The steel used in this study is Sidenor's commercial steel SCM40, a medium C (0.4 wt%) high Si (3 wt%) steel. Cylindrical specimens of different size were used: $10 \times 4\,mm^3$ for pure dilatometry tests and $10 \times 5\,mm^3$ for deformation tests.

Tests were performed in a Bahr 805D high-resolution dilatometer equipped with an induction heating coil. Helium was used as quenching gas and the temperature was controlled by a type K thermocouple welded to the central part of the sample surface. The equipment enabled thermal and thermomechanical treatments while tracking phase transformations by monitoring the change in length of the sample. The dilatometry tests were performed using a specific module, equipped with fused silica push-rods to measure longitudinal length changes. With respect to the compression tests, they were carried out by using a deformation module, with silicon nitride punchers separated from the sample by molybdenum films in order to reduce friction. A second identical dilatometer with an additional laser-interferometer was used under specific conditions, for the simultaneous measurement of both the specimen radius and the length changes during testing.

Quantitative X-ray diffraction (XRD) analysis was used to determine retained austenite and bainitic ferrite volume fractions. Samples were step-scanned in a Bruker AXS D8 X-ray diffractometer with a rotating Co anode X-ray tube as a radiation source, Goebel mirror optics and a LynxEye Linear Position Sensitive Detector for ultra-fast XRD measurements. A current of 30 mA and a voltage of 40 kV were employed as tube settings. Operational conditions were selected to obtain X-ray diffraction data of sufficiently high quality. XRD data were collected over a 2 h range of 35–135° with a step size of 0.01°. Volume fractions have been obtained from the integrated intensities of the (002) (112), and (022) peaks, which correspond to the ferrite planes, and from the (200) (220) and (311) austenite peaks [35]. Using several peaks has been proved to avoid non-accurate results related to crystallographic texture [36].

EBSD measurements were carried out using a Zeiss Auriga Compact FIB-SEM, operating at 20 kV. On the undeformed sample, a $230 \times 173\,\mu m^2$ area was scanned on the transverse section, using a step size of 0.35 μm. On all deformed samples, $57 \times 43\,\mu m^2$ areas were scanned on the longitudinal sections, using a smaller step size of 0.15 μm, to avoid possible indexing problems arising from the dislocation density increase due to plastic deformation. The subsequent analysis was performed by means of a HKL Channel 5 system (Oxford instrument) and Matlab®.

The sample preparation for XRD was performed using standard metallographic procedures, introducing a final set of cycles of etching and polishing in order to remove the surface layer that had been plastically deformed during the grinding step. That surface layer might contain traces of martensite induced by the sample preparation, which would underestimate its real fraction.

The sample preparation for EBSD measurements was similarly to this for XRD, but the set of cycles of etching and polishing was replaced by a polishing step with 50 nm colloidal silica suspension.

The microstructure was revealed by the same process that has just been described, followed by a final etching with a 2% nital solution, more appropriate for fine microstructures than the commonly used metallographic preparation procedure [37,38].

The microstructure was observed under a JEOL JSM-6500 field emission gun scanning electron microscope (FEG-SEM) operating at 10 kV. Specimens were inspected throughout the transverse and the longitudinal sections. Because of the barrelling effect observed in the ausformed samples, the microstructure was in all cases examined at the central part of both sections, where the local plastic strain approaches the macroscopic strain [21,39].

In order to estimate the bainitic ferrite plate thickness (t), the procedure explained in reference [40] has been followed. It is based on the measurement of linear intercepts on SEM micrographs plus a subsequent stereographic correction applied to the mean linear intercept (LT), as follows:

$$t = 2\overline{L_T^{\alpha}}/\pi \text{ with the 95\% confidence error } E = \pm 2{\cdot}\sigma_L^{\alpha}/\pi\sqrt{N},$$

where $\sigma_L{}^{\alpha}$ stands for the standard deviation of the intercepts and N for the number of measurements. This stereographic projection should be applied to take into account the non-perpendicularity of ferrite plates with respect to the observed section. In this case, plate thickness was measured in both transverse and longitudinal sections.

The thermal etching technique has been used to reveal the Prior Austenite Grains (PAG). This technique allows the austenite grain boundaries to be revealed in dummy samples previously subjected to the same austenization treatments but vacuum-cooled to room temperature instead, in order to avoid displacive transformations. Special sample preparation conditions are given elsewhere [41]. Once the parent grains were revealed, their areas were measured on light optical micrographs by the image processing program ImageJ [42], from which the equivalent diameter of the grains could be calculated.

Hardness was measured as HV(10) and the presented results are an average of at least 3 values. Measurements have been made on both transverse and longitudinal sections to take into account anisotropic effects.

All necessary thermodynamic calculations were performed by means of. MTDATA, which relies on the NPL-plus database for steels [43]. The piece of software MUCG-83, developed by Bhadeshia at Cambridge University [44,45] was used to obtain the theoretical TTT diagram and bainite start temperature B_S.