Effects of thermomechanical processing on the microstructural development of continuously cooled carbide-free bainitic steels

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Continuous cooled carbide-free bainitic forging steels (CCBS) combine several metallurgical concepts in the design philosophy of their microstructure. The "carbide-free" term originates from the inhibition of carbide precipitation in austenite, due to the judicious introduction of silicon or aluminum the driving force for carbide precipitation is lowered, allowing austenite enrichment by carbon. Continuous cooling is employed after forging at high temperature, leading to the final microstructure. In the present case, the microstructure evolves to display a bainitic matrix formed along the cooling path. The combination of these concepts, give rise to a competitive class of steels exhibiting adjusted hardenability from the bainite and also good ductility due to the retained austenite deformation capability. However, the distinctive characteristic of these steels originates from the continuous cooling. This is the case, because continuous cooling offers several advantages in comparison to quenching and isothermal procedures, as it reduces distortion/cracking and simplifies the heat treatment considerably. Currently, there are still knowledge gaps in CCBS, in particular regarding the kinetics of the bainitic transformation in combination with previous deformation and about the mechanisms of morphology evolution to obtain a granular shape. This work aims to describe the kinetics of the austenite decomposition during continuous cooling. For this, several thermomechanical treatments were performed with a dilatometer and monitored in-situ via high-energy X-Ray diffraction. The treatments included variations of deformation conditions after austenitization, including hot and warm deformations, aiming to assess how the transformation kinetics are affected. In comparison to the undeformed condition, a delay in the transformation kinetics could be observed for all investigated conditions for two CCBS steel grades with low and medium carbon contents. Furthermore, the mechanisms which cause this delay are assessed and described based on in-situ data and also on post-process analysis of microstructural modifications caused by thermomechanical processing.

Keywords: thermo-mechanical treatment; synchrotron diffraction; bainitic steels; continuous cooling; carbide-free bainite

1. Introduction

It is well known that deformation can affect the kinetics of displacive transformations[1,2]. Kinetic variations can directly influence the microstructure and, consecutively, the performance of thermomechanically processed bainitic and martensitic components. Nevertheless, the root causes of these kinetic modifications are still being debated, with many contradicting works in the literature[3,4,5]. This work aims to investigate the kinetic variations brought upon by thermomechanical processing in continuously cooled carbide free bainitic steels with by means of in-situ high energy Synchrotron experiments. The main goal is to discuss possible mechanisms that cause the reported variations in kinetics.

2. Experiment

A medium-carbon carbide-free (≈1 wt.-% Si and 0.4 wt.-% C) was employed in this investigation. Compression mode thermomechanical treatments were conducted by a Bähr 805 A/D dilatometer. After austenitization at 1050 °C, samples were cooled to two temperatures to apply the deformation step. In the first case, 0% or 40% hot deformation was applied at 950 °C, followed by fast cooling at 40 °C/s until 600 °C. In the second case, the samples were cooled to 600 °C at a rate of 40 °C/s, where 40% warm deformation was applied. In both cases, a transition to continuous cooling (CC) at a rate of 0.5 °C/s until room temperature was carried out after deformation.

In-Situ HEXRD experiments of the aforementioned processing route were conducted in the beamline P07-EH3 at DESY in Hamburg, Germany. The X-Ray beam had a spot size of 1 mm × 1 mm and an energy of 100 keV, allowing the collection of 19 full diffraction rings with in transmission mode. The Debbye-Scherrer rings were recorded with an acquisition rate of 5 Hz by a large 2D-Detector placed behind the samples. Estimation of the dislocation density of austenite was carried out via the approach developed by Williamson and Smallman[6]. Silveira, et al[7], have shown that this can be used for evaluating dislocation density, even in extreme conditions such as in Laser Metal Deposition. In order to analyze how the deformation affected austenite, etchings with a solution of picric acid were carried out to observe the state of the PAG right after 0 and 40 % deformation.

3. Results

3.1 Kinetics of phase transformation

The evolution of BCC phase content *versus* temperature upon cooling is shown in Figure 1.

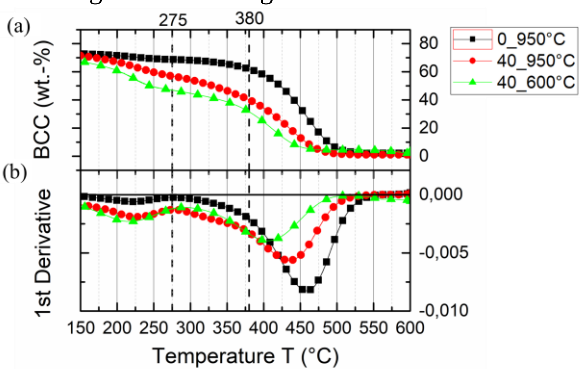

Figure 1. (a) BCC Phase evolution upon cooling. (b) Shows the 1st derivative of the curves shown in (a). The legend indicates the deformation degree in % and deformation temperature.

The impact of deformation on the bulk transformation can be assessed by comparing the evolution of the BCC fraction between the undeformed and deformed specimens in Figure 1 (a). The transformation rate, obtained from the 1st derivative of the phase amount over the temperature, is shown in Figure 1 (b). The observed phase transformation kinetics could then be divided in three stages:

1. Accelerated kinetics (500 to 380 °C): the experiments with deformation exhibit lower ferrite amounts for the same temperature, characterizing a kinetic delay. Nevertheless, this stage shows the overall fastest bainitic ferrite increase;

2. Slower kinetics (380 to 275 °C): in this stage the kinetics of all experiments slow down;

3. Transformation end (275 °C to RT): a rapid increase in BCC fraction at low temperature is seen in this stage, which can be attributed to the formation of martensite.

Mechanical stabilization of austenite due to a high density of defects is often reported to explain such kinetic delays in the bainitic transformation. To investigate this hypothesis, Figure 2 shows the evolution of the estimated dislocation density in austenite. Since a considerable kinetics delay was detected, the dislocation densities are shown as a function of the of the BCC phase amount.

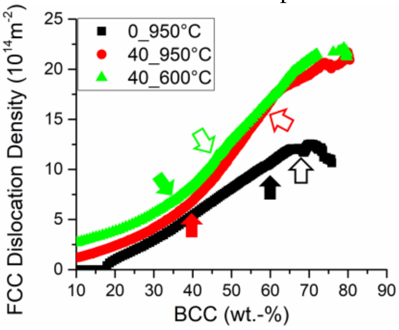

Figure 2. FCC dislocation densities evolution *versus* BCC phase content. The bold and hollow arrows indicate the phase fraction correspondent to 380 °C and 275 °C, respectively.

Figure 2 shows that a faster dislocation accumulation is seen for the deformed specimens, strongly diverging from the undeformed variant at ≈ 40 wt.-% BCC. In the case of displacive transformations, austenite has to accommodate the volume increase and shear deformation of bainite or martensite[2], resulting in a defect increase within austenite. The exponential-like shape of the deformed specimen curves, indicates that the austenite is increasingly deformed as the transformation progresses. This is expected, since the austenite phase content reduces along the transformation what results in a smaller austenite population increasingly affected by the surrounding formed BCC phase.

3.2 Prior austenite grain

The morphology of the Prior Austenite Grain, PAG undeformed and after 40% hot deformation experiments are shown in Figure 3.

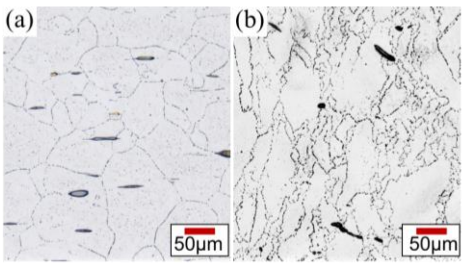

Figure 3. Austenite grain morphology of (a) undeformed and (b) hot formed sample with 40 % deformation at 950°C.

As expected, deformation-induced recrystallization caused an average PAG refinement, as seen in Figure 3. The refinement and aspect of the grain boundaries in (b) made the quantification of grain size impractical, however, it is clear that the average size is strongly reduced compared to the undeformed counterpart. In the undeformed specimen, a uniform distribution of equiaxed grains is seen, whereas the deformed variant displays two populations of grains: small quasi-equiaxed and elongated grains, both exhibiting “jagged” boundaries.

4. Discussion

4.1 Mechanical Stabilization

Mechanical stabilization states that bainitic growth is hindered by a high degree of defect accumulation in austenite. This phenomenon is more prone to occur in the scenario of warm deformation, as most of the dislocations generated during strain hardening should remain in austenite, since recrystallization/recovery is inhibited in this temperature range. As it was shown in Figure 2, it appears that the performed experiments obey this premise, since the warm formed variant displayed the higher dislocation density values and also the lowest transformation rate for the same amount of BCC phase. As this logic is also directly applicable to the hot formed variant, it is plausible that the delay of the first stage of the transformation is caused by the increased defect amount.

Regarding the deceleration beginning at ≈ 380 °C, a marked transition into a lower transformation rate is seen for all experiments. Assuming that the same hypothesis of mechanical stabilization is also causing this deceleration, and that the most probable environment for it to occur would also during warm deformation, it is then plausible that a ‘critical amount’ of dislocations was reached in the warm formed specimen, strongly restricting bainite growth. Taking this into account, the dislocation density in austenite at the onset of the second transformation stage is ≈ 7.5 * 10^{14} m^{-2}. By comparing this value with the other experiments in Figure 2, the hot formed variant reaches this mark at ≈ 43 wt.-% BCC while the undeformed variant at ≈ 47 wt.-% BCC. Thereby, a hypothesis of critical dislocation density is plausible for the hot formed variant, since the deceleration is seen at around similar amounts of BCC and the transformation rate and amount (≈ 15 wt.-% BCC) within the second transformation stage is similar to the warm formed variant. However, in consideration of the undeformed counterpart, a very different behavior is shown. Although the same critical dislocation density level is achieved at ≈ 47 wt.-% BCC, there are no signs of decreasing transformation rate. Instead, a deceleration is seen only at about 60 wt.-% BCC.

The fact that the transformation rate in the undeformed specimen remained constant after reaching a ‘critical’ dislocation level suggests that the cause of the second stage delay is not mechanical stabilization of austenite. It could be argued that austenite is stabilized via the incomplete reaction phenomenon[2], since the BCC wt.-% is higher than in the deformed counterparts, however, the results in Figure 1 show that the transformation is only slowed, and about 8 wt.-% BCC is formed before the third transformation stage.

4.2 Austenite grain modifications due to recrystallization

By comparing the morphology and size of both grain populations of the hot formed with the undeformed specimens, it was concluded that it underwent only partial recrystallization. Therefore, while bainite has grown from defect-poor equiaxed grains within the undeformed specimen, the bainite of the deformed variant developed from austenite with a heterogeneous grain size and defect accumulation. In this scenario, the kinetics of bainite growth should depend on the specific characteristics of its mother grain, what can lead to heterogeneous kinetics throughout the same specimen. Therefore, it is possible that the transformation occurs faster in the populations of low-defect grains, while the opposite is true for the strain-hardened elongated grains.

The aforementioned hypothesis states that the kinetics vary from grain to grain, depending on the recrystallization extent. This explains the behavior seen for the hot-formed specimen, suggesting that the fast kinetics in the first stage of the transformation takes place at the expense of the defect-poor grains. Later on, when the transformation has stopped within the defect-poor grains, the kinetics would reflect the transformation within the strain-hardened population. However, by applying this hypothesis to the case of warm forming, some contradictions are noticed, i.e., the warm deformed specimen cannot exhibit a defect-poor population of grains, as recrystallization is not expected in this temperature range, implying that transformation has to start from a strain hardened austenite. Hence, the hypothesis of a bimodal distribution of defects/grains in austenite cannot explain the deceleration seen at 380 °C. At this point, it has been ruled out that the second stage deceleration is caused by modifications connected to recrystallization, mechanical stabilization or the incomplete reaction phenomenon. A remaining hypothesis is an alteration in the kinetics due to a morphological transition from upper bainite (UB) to lower bainite (LB) during CC.

4.3 Transition from upper to lower bainite

Takahashi and Bhadeshia developed a model to describe the transition from UB to LB[8]. They considered that bainitic ferrite is formed by displacive mechanism with a supersaturated carbon content. Depending on the diffusivity of C at a given temperature, carbon would diffuse out or precipitate within the laths, determining whether upper or lower bainite is obtained, respectively. Later on, Chang[9] investigated 8 different carbide-free bainitic steels, developing an equation to quantify the lower bainite start temperature (LB_s). Besides supporting the supersaturated bainite model of Takahashi and Bhadeshia, he also showed the mutual transformation of UB and LB in the same specimen. The same author[10] also reported an acceleration in the transformation kinetics under the LB_s temperature, what was attributed to transformational stress effects. Lee, Park and Lee[11] investigated the effect of coarse and fine PAGS on the kinetics of continuously cooled UB and LB. They observed a deceleration in kinetics during the continuous cooling of a medium carbon steel, attributing it to the UB to LB transition. Furthermore, the authors stated that a constant LB_s of ≈ 430 °C was found for the studied steel, matching the equation proposed by Chang, concluding that the PAGS did not influence LB_s.

According to the aforementioned works, the transition from upper to lower bainite during CC or isothermal holding can generate a change in transformation kinetics. Here, the transition is not governed by the grain size of the austenite, what is reasonable, since the carbon precipitation must depend on its amount and diffusivity within the newly formed bainitic ferrite. In the present work, the equation of Chang was applied to estimate the LB_s of the applied steel, which resulted in the temperature of 376 °C. This value has a good match with the observed deceleration temperature of 380 °C. Furthermore, it has been shown that the delay on the bulk transformation of all experiments occur within the same temperature range, irrespective of the condition of the grain size and strain hardening. Another phenomenon detected at the LB_s temperature range, is a faster increase in the dislocation density of austenite relative to the undeformed specimen, as indicated by the bold arrows in Figure 2. This suggests that the bainite mechanism transition to lower bainite is also accompanied by a faster increase in dislocation density, what may be connected to the lower thickness bainitic lathes. Therefore, it is likely that deceleration in the second stage of the transformation seen in this work is caused by a mechanism change in the formation of upper to lower bainite.

5. Conclusions

1. At the transformation onset, hot and warm deformation lowered the obtained BCC phase content, in comparison to the undeformed experiment. This was attributed to a high defect accumulation in austenite, hindering the bainitic transformation.
2. With the progression of cooling, a second deceleration was seen in all studied scenarios, irrespective of deformation was applied or not. This phenomenon was attributed to a change in the transformation mechanism from upper to lower bainite.
3. At the final stage of the transformation, the lesser amount bainite in the deformed experiments limited the enrichment of austenite, what enabled an increased amount of low temperature BCC.

6. References

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