Microstructure development and mechanical properties of medium carbon carbide-free bainite steels

Mohamed Soliman, Heinz Palkowski*

Institute of Metallurgy, Clausthal University of Technology, 38678 Clausthal-Zellerfeld, Germany

Abstract

This paper concerns with a further development of low temperature carbide-free bainite steels by decreasing its carbon content aiming at both, accelerating the bainite transformation reaction and obtaining a wider range of mechanical properties. For this purpose, three medium carbon steels containing 0.32, 0.42 and 0.56 % C were cast. Dilatometric measurements were performed to design the suitable heat-treatment parameters of these alloys. The structure was characterized using light optical microscopy, scanning electron microscopy and x-ray diffractometry. Decreasing the carbon content has facilitated accelerating the bainite transformation and obtaining the desired fine-structure in shorter time frames. Additionally, decreasing the carbon content below the eutectoid composition enabled introducing the polygonal ferrite phase in the structure which results in a wider window for the design of mechanical properties. The best alloy for producing a wide range of heat treatment parameters and hence a wide range of mechanical properties is that containing 0.42 % C.

Keywords: Carbide-free bainite steels; Phase transformation; Intercritical annealing

1. Introduction

Carbide-free bainitic steels alloyed with silicon and manganese are reported to achieve an excellent combination of high strength and toughness (Deliry, 1965; Caballero et al., 2009). This improvement in toughness...
is achieved by alloying the steel with about 1.5 wt% silicon to suppress the precipitation of cementite during the bainite formation.

Many research efforts are exerted to develop this structure by forming the bainite at low temperatures, currently known as low temperature bainite (Garcia-Mateo et al., 2003; Soliman et al., 2007). This type of bainite is composed of alternating, nano-size plates of ferrite and austenite/martensite. The ultra-high strength of this steel is the result of its fine bainite structure. The observed refinement is a consequence of the ability of high carbon content and low transformation temperature to enhance the strength of the austenite. It is expected that the bainite plates will become thinner as the yield strength of the austenite, from which they are formed, increases (Bhadeshia, 2001).

However, the high carbon content results in the existence of large regions of untransformed austenite in the microstructure which is known to be detrimental to the mechanical properties. The incomplete transformation phenomenon, which limits the amount of bainite that can be formed at any temperature, is the reason for the existence of these large regions of untransformed austenite (Bhadeshia, 2001). Decreasing the carbon concentration would reduce the quantity of the untransformed austenite and accelerate the bainite transformation kinetics. Decreasing the carbon content would also improve the formability and weldability of the alloy.

In this paper three alloys are introduced aiming at investigating the effect of carbon content on the microstructure development as well as mechanical properties in carbide-free fine bainite steels.

| Nomenclature | Description |
|--------------|-------------|
| $A_{cl}$     | temperature, where “ferrite” and “ferrite + austenite” phase fields under equilibrium state separate |
| $A_{c3}$    | temperature, where “ferrite + austenite” and “austenite” phase fields under equilibrium state separate |
| CS          | compression strength |
| FS          | fracture strain |
| $M_S$       | martensite start temperature |
| $T_A$       | annealing temperature |
| $T_B$       | isothermal bainitic transformation temperature (austempering temperature) |
| $T_C$       | temperature in the intercritical region at which the carbides are entirely dissolved |
| $V_T$       | retained austenite volume percentage |
| $Y_{SC}$    | yield strength under compression loading |
| $t_P$       | thickness of the bainite plate |
| $t_B$       | austempering time |
| $\alpha$    | ferrite phase |
| $\gamma$    | austenite phase |
| $\phi$      | true strain |

2. Experimental procedure

2.1. Material

The steels investigated were produced in a laboratory sand casting process by induction melting and solidifying in Y-blocks with a size of 45 × 190 × 230 mm. The blocks were machined to 40 × 187 × 227 mm to remove the cast surface structure and then heated up to 1200 °C for 1 h in a protective gas atmosphere. The material was then hot-rolled in 7 steps on a 12”-2-high rolling stand to a final thickness of 12 mm with a true strain of about $\phi = 0.17$ for each step. The chemical compositions of the steel studied are listed in Table 1. Mn and Cr had been added for hardenability, Mo to prevent temper embrittlement due to P and Si to prevent the precipitation of cementite during the bainite formation. Cementite is a cleavage and void-initiating phase to be avoided in steels of high strength. Al and Co had been added to accelerate the bainite transformation process.
2.2. Transformation kinetics

The transformation kinetics were studied by dilatometric experiments on a Bähr-DIL805A/D dilatometer with cylindrical samples of 5 mm diameter and a length of 10 mm. To control and monitor the programmed thermal cycle, thermocouples of type S "Pt/Pt-10% Rh" with a nominal diameter of 0.1 mm were individually spot welded to the specimen’s surface in central position. The dilatometric curves were recorded along the thermal cycle with the help of a computer data accusation system. Further details to the experimental device are given in Soliman and Palkowski (2008). Dilatometric experiments were applied to:

1- determine the martensite start temperature (Mₜ). Therefore samples were heated up to 950 °C into the austenite region with a heating rate of 10 Ks⁻¹, isothermal held for 30 min and then cooled down to room temperature with a cooling rate of 10 Ks⁻¹,

2- depict the intercritical region to allow the selection of the suitable intercritical temperature for obtaining (γ + α) microstructures as described in Soliman et al. (2009),

3- monitor the kinetics of bainite transformation during austempering at temperatures between 220 °C and 320 °C after austenitizing by holding for 30 min in the γ or (γ + α) region.

2.3. Microstructure investigations

To investigate the microstructure constituents the specimens were prepared by mechanical grinding followed by polishing up to a 1 μ-grade diamond paste. The microstructure has been examined by light optical microscope after etching using LePera etchant (Girault et al., 1998). The samples for the X-ray diffraction analysis were etched a bit with nital to remove the distorted thin layer resulting from the metallographic preparation procedure, whereas those for the scanning electron microscope (SEM) investigations had been strongly nital etched. The volume fraction of retained austenite, \( V_γ \), was measured using XRD analysis as described in Cullity (1978). The experiments were conducted using a Siemens D5000 diffractometer with Fe-filtered Co-Kα radiation at 40 kV and 40 mA. The peak intensities of the whole patterns were collected by step scanning.

2.4. Mechanical properties

The mechanical properties of the materials were characterized in terms of their compression properties. Quasi-static-compression tests were performed at a cross head speed of 1 mm/min on samples having of 5 mm diameter and 10 mm length. All the samples were machined after homogenization and before the heat treatment because of the softest material condition at this stage. Hardness testing was carried out using Vickers hardness according to ASTM E384 (2011) applying a load of 20 kg (HV20).
3. Results and discussions

3.1. Heat treatment parameters

The materials were annealed by isothermal holding in the γ region as well as in the intercritical, γ+α, region. A detailed description of the methodology adopted to depict the intercritical region is given in Soliman and Palkowski (2007) and Soliman et al. (2009).

Table 2 gives the critical temperatures as well as the designed heat treatment parameters, where the annealing temperature $T_A$ is selected above the cementite dissolution temperature $T_C$ to avoid the carbide formation, whereas the austempering temperature $T_B$ is selected above $M_S$. Increasing the carbon content narrows the intercritical regions and suppresses the $M_S$. The low carbon content of alloy A1 limits obtaining the low temperature bainite because of the relatively high $M_S$. On the other hand, the very narrow temperature range between $T_C$ and the temperature separating “ferrite + austenite” and “austenite” phase fields at equilibrium state $A_{e3}$ of alloy A3 limits obtaining the polygonal ferrite phase in this alloy. The best alloy for producing a wide range of heat treatment parameters and hence a wide range of mechanical properties is alloy A2.

| Alloy | $A_{e1}$ | $T_C$ | $A_{e3}$ | $T_A$ [°C] | $T_B$ [°C] | $t_B$ [ks] |
|-------|----------|-------|----------|----------------|----------------|----------------|
| A1    | 775      | 816   | 920      | 950            | 607            | 3.7            |
|       |          |       |          | 950            | 860            | 4.8            |
|       |          |       |          | 950            | 835            | 5.0            |
|       |          |       |          | 950            | 320            | 10.0           |
| A2    | 787      | 827   | 895      | 920            | 239            | 25.9           |
|       |          |       |          | 920            | 223            | 33.4           |
|       |          |       |          | 920            | 218            | 39.0           |
| A3    | 776      | 820   | 836      | 920            | 207            | 17            |
|       |          |       |          | 820            | 250            | 47            |
|       |          |       |          | 820            | 220            | 109           |

3.2. Microstructure features

Representative structures obtained after applying the heat treatments and investigated using light optical microscope are given in Fig. 1. LePera etchant was used to differentiate the polygonal ferrite within the acicular ferrite aggregates (see Figs. 1b, c and e). As clearly seen in Fig. 1 it is not possible to distinguish individual ferrite plates within the thin aggregates using light optical microscope. Therefore, Structure investigations using SEM are conducted. Fig. 2 shows that the ferrite aggregates build up from many ferrite plates in the same crystallographic orientation. The plate thicknesses of the bainite plates $t_p$ has been estimated from scanning electron micrographs, whereas the retained austenite volume percentage $V_A$ has been measured using XRD method. Table 3 summarizes the obtained $t_p$ and $V_A$ corresponding to the alloy and the applied heat treatments parameters. The thickness of the bainite plates was refined by decreasing the bainite transformation temperature and increasing the carbon content. On the other hand, considering the same bainite transformation temperature, introducing ferrite to the microstructure resulted in marginal decreasing in $V_A$. A significant decrease in $V_A$ is observed by decreasing the carbon content of the alloy.
Fig. 1. Microstructure after austempering of alloy A1 at (a) 950-320 °C, (b) 860-320 °C and (c) 835-300 °C, of alloy A2 at (d) 920-250°C and (e) 830-250 °C and A3 at (f) 920-250 °C. The temperature combinations were selected according to Table 2.

Fig. 2. Micrographs after austempering of alloy A1 at (a) 950-320 °C, (b) 860-320 °C, (c) 835-320 °C and (d) 835-290 °C, of alloy A2 at (e) 920-250 °C and (f) 850-250 °C and of alloy A3 at (g) 920-290 °C and (h) 920-250 °C.
3.3. Mechanical properties

Table 3 summarizes the basic mechanical characteristics of the alloys achieved by compression tests. The highest fracture strain (FS) and lowest compression strength (CS) is observed for alloy A1. Alloy A3 shows the highest CS. The alloy A2 recorded a yield strength (YS_c) comparable to that one of alloy A2.

Table 3. Structural and mechanical characteristics of the alloys A1, A2 and A3.

| Ti  | Tb  | tf  | Vr  | HV20 | CS  | YS_c | FS  |
|-----|-----|-----|-----|------|-----|------|-----|
|     |     |     |     |      |     |      |     |
| A1  | 950 | 320 | 183 | 13.9 | 2203| 1365 | 0.84 |
|     | 860 | 320 | 181 | 12.4 | 2226| 1386 | 0.874|
|     | 835 | 320 | 162 | 12.5 | 2179| 1279 | 0.90 |
|     | 300 | 110 | 10.8|      | 2198| 1347 | 0.84 |
| A2  | 920 | 290 | 126 | 20.1 | 2300| 1444 | 0.62 |
|     | 250 | 85  | 17.5|      | 2556| 1462 | 0.46 |
|     | 850 | 290 | 112 | 17.9 | 2390| 1300 | 0.70 |
|     | 250 | 77  | 16.3|      | 2624| 1710 | 0.48 |
|     | 230 | 70  | 16.8|      | 2667| 1560 | 0.48 |
| A3  | 920 | 290 | 82  | 22.8 | 2470| 1481 | 0.76 |
|     | 250 | 66  | 18.7|      | 2645| 1747 | 0.68 |
|     | 220 | 52  | 18.8|      | 2854| 1810 | 0.58 |

4. Conclusions

Decreasing the carbon content of the carbide-free fine bainite steels resulted in accelerating the bainite transformation and obtaining a fine bainite structure in shorter time frames. Furthermore, the produced bainite percentage is increased on the expense of the retained austenite. Decreasing the carbon content below the eutectoid composition enabled the introduction of the polygonal ferrite phase in the structure. This resulted in a significant increase in the fracture strain and marginal decrease in the strength of the studied alloys.

Acknowledgements

The authors acknowledge the financial support from Deutsche Forschungsgemeinschaft (DFG) – PA 837/16-1.

References

Bhadeshia, H.K.D.H., 2001. Bainite in steel, IOM Commercial Ltd, London.
Caballero, F.G., Chao, J. J., Cornide, Garcia-Mateo, C., Santofimia, M.K., Capdevila C., 2009. Toughness deterioration in advanced high strength steels, Mater. Sci. Eng., A525, 87-95.
Cullity, B.D., Stock, S.R., 2001. Elements of X-ray diffraction, Pearson/Prentice Hall, Inc.
Deliry, J., 1965. Mem. Sci. Rev. Metall., 7–8, 527-550.
Garcia-Mateo, C., Caballero, F.G., 2003. Acceleration of low-temperature bainite, ISIJ Int., 43, 1821-1825.
Girault, E., Jacques, P., Harlet, P. H., Mols, K., Van Humbeeck, J., Aermond, E., Delannoy, F., 1998. Metallographic Methods for Revealing the Multiphase Microstructure of TRIP-Assisted Steels Mater. Charact., 40, 111-118.
Soliman, M., Palkowski H., 2007. Ultrafine bainite structure in hypoeutectoid steels, ISIJ Int., 47, 1703-1710.
Soliman, M., Palkowski H., 2008. On factors affecting the phase transformation and mechanical properties of cold-rolled transformation-induced-plasticity-aided steel, Metall. Mater. Trans., A 39, 2513–2527.
Soliman, M., Weidenfeller, B., Palkowski, H., 2009. Metallurgical phenomena during processing of cold rolled trip steel, Steel Research Int., 80, 1, 57-65.
ASTM E384 - 11e1, 2011. Standard test method for Knoop and Vickers hardness of materials, ASTM Int., PA, USA.