AUSTENITE TRANSFORMATION AND AGE HARDENING OF HSLA-80 AND ULCB STEELS

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ABSTRACT

The aim of this work was to determine the CCT diagrams of HSLA-80 and ULCB steels and its performance during age hardening. It was verified that both alloys showed bainitic microstructure with low C content, commonly designed in the literature as "granular" bainite. The hardenability of the ULCB steel was greater than the HSLA-80 due to the presence of Nb/B and Mo in the first alloy. The HSLA-80 steel developed maximum hardness during a 600°C age hardening, while for the ULCB steel this occurred at 500 or 600°C age hardening and took less time. Both steels showed a significant hardness decrease during the 700°C age hardening, which was probably due to overaging and tempering effects.

Keywords: HSLA-80 and ULCB steels; CCT diagram; precipitation; copper

1 INTRODUCTION

Two new designs of microalloyed steels were proposed some years ago as alternatives to the HY-80 and HY-100 alloys, with similar mechanical strength levels but better weldability, thanks to the minimization of their C content: ASTM A710/HSLA-80 and ULCB – Ultra Low Carbon Bainite. Their better weldability allows a decrease of up to 50% in the total fabrication costs of components and structures [1]. Due to their low C content, such alloys must have strengthening mechanisms relatively independent from this element. In the case of the HSLA-80 steel, the copper precipitation that occurs during a supplementar aging treatment represents a significant contribution to mechanical strength, while in the ULCB steel this role is played by the bainitic microstructure and solid solution hardening [2-7]. Although the presence of microalloy elements in the ULCB steel also promotes some hardening through precipitation in ferrite, this material normally is not submitted to aging treatments during its industrial processing.

Another advantage that arises from the use of such steels is the supression of the quench and tempering treatment that normally is needed to the HY-80/HY-100 steels. It is true that the HSLA-80 steels require aging treatment after hot rolling, but this has some favourable aspects during the forming process. The plate can be formed in the as-rolled condition, when its yield strength still is relatively low; the aging treatment can be applied subsequently, after welding and finishing operations, acting simultaneously as a stress relief treatment [2].

The objective of this work was to determine the CCT diagrams (transformation under continuous cooling) relative to the austenite of HSLA-80 and ULCB steels, as well to verify its behavior during aging under temperatures between 500 and 700°C.

2 EXPERIMENTAL PROCEDURE

The alloys used in this work were produced in a vacuum induction furnace. One 100 x 130 x 850 mm, 85 kg ingot was produced for each studied alloy, that is, HSLA-80 and ULCB steels. Their chemical analysis can be seen in table 1. These ingots were hot rolled into rectangular bars with 50 x 42 mm cross section in order to break and homogeneize the as cast structure. Samples for the hot rolling tests were machined from these bars. These specimens were reheated to 1200°C and hot rolled down to 7.5 mm thickness; the finishing temperature was 750°C. Finally, these hot rolled samples were machined in order to get the samples used for the determination of CCT (continuous cooling transformation) diagrams and in the aging tests.

Table 1: Chemical analysis of the heats.

Steel	C	Mn	Si	P	S	Al_{sol}
HSLA-80	0.044	0.65	0.32	0.005	0.011	0.013
ULCB	0.033	1.93	0.29	0.007	0.011	0.006
Ni	Cr	Cu	Mo	Nb	Ti	B
0.87	0.77	1.12	0.23	0.077	---	---
0.39	---	---	0.35	0.062	0.029	0.0016

The CCT diagrams of both alloys were determined in order to characterize the evolution of austenite transformation during continuous cooling. A Formastor dilatometer was used; the samples were heated to 930°C during 20 minutes; after this process both alloys showed an ASTM 9.5 austenitic grain size. Following this heating step the samples were cooled down with the following rates: 44, 30, 20, 10, 0.3, 2.67, 1.33, 0.5 and 0.25°C/s. The analysis of the volumetric variation of the samples during cooling and the final microstructures yielded data for drawing the CCT diagrams.

The aging treatments were carried out using samples of both alloys with dimensions 10 x 15 x 7 mm³. They were aged at 500, 600 and 700°C during 5, 15, 30, 60 and 120 minutes. As the experimental design included very quick aging treatments (5 and 15 minutes), they must be carried out using furnaces with high heating rate, minimizing the thermal inertia of the samples. For this reason the aging treatments were done in a liquid tin bath, heated in a electric resistance furnace. The bath temperature was controlled using a 1.5 mm diameter chromel-alumel thermocouple protected with a stainless steel sheath immersed in the metallic bath. The cooling of the samples was done in still air.

The hardening effect due to the aging treatment was determined through the measuring of the Vickers hardness of the samples; a test load of 5 kg was used. The surface of the samples were ground and polished before the tests. Ten values of hardness were determined for each sample along the diagonal of the wider face of the sample and then a mean value was calculated. This procedure aims to minimize the influence of some eventual segregation and/or orientation present in the material.

3 RESULTS AND DISCUSSION

The CCT diagram determined for the HSLA-80 steel showed that the HSLA-80 is basically a bainitic steel, as it was detected a very low fraction (5%) of ferrite when the samples were cooled at 0.5°C/s or lower cooling rates. The same fact was verified for the ULCB steel, which hardenability is even greater, as a ferritic field did not even appeared in its CCT diagram. It must be noted that, in this case, the synergistic effect between Nb and B and the presence of Mo have an essential role in this result [8].

Figure 1 shows the hardness evolution observed in the samples of HSLA-80 and ULCB in function of the applied cooling rate. Both alloys showed a good logarithmic correlation between hardness and cooling rate, with r^2 about 0,9. It can be clearly seen that the hardness of both alloys was virtually identical for low cooling rates, from 0,25 to 0,30°C/s. However, for greater values of cooling rates, the hardness of the HSLA-80 steel increased slightly more than ULCB steel. Apparently this result can be attributed to the slight higher C content and the presence of Cu in the HSLA-80 alloy.

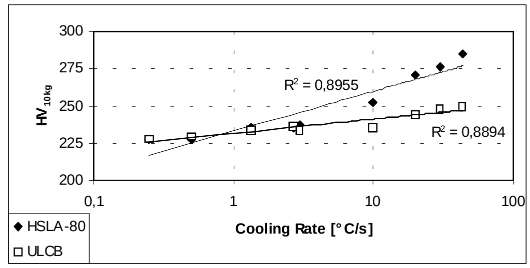

Figure 1: Hardness evolution in function of cooling rate observed for the HSLA-80 and ULCB steel samples used for the determination of the CCT diagrams.

The use of the results got in these diagrams for the definition of process parameters for hot rolling must be done with care. It must be remembered that, as the samples used for the determination of these CCT diagrams were reheated to 930°C, their soluble Nb content certainly was very low, about 0,016%, only 29% from the total available in the nominal composition of the alloy. Reheating temperatures used in the industrial hot rolling are about 1200°C or over; this assures a complete Nb solubilization. The Formastor dilatometer used in this work also did not permit the hot forming of the samples before cooling, so its effects were not reflected in the CCT diagrams determined in this work. Generally hot forming leads to an acceleration of austenite transformation [11].

The hardness evolution with aging treatment time for the alloys studied in this work, for each temperature used (500, 600 and 700°C) can be seen in figure 2. It shows that, in the case of HSLA-80, the aging treatment at 500°C promoted a very slight hardening. This sample, with an initial as-rolled hardness of 244 HV, showed a fast increase in this parameter up to 30 minutes of treatment, reaching a value of 251 HV. The hardness values kept practically constant for longer treatment times. For its turn, the aging treatment at 600°C lead to a fast hardness increase, reaching a maximum value after 30 minutes of treatment: 271 HV. From this point hardness decreased slowly, reaching 263 HV after two hours of treatment. Finally, aging at 700°C initially lead to a small hardness increase, which reached a maximum value after 5 minutes of treatment, that is, 254 HV. From this point on hardness progressively decreased. At the end of the aging treatment the sample showed a hardness even lower than the initial as-rolled sample: 237 HV.

The results show that the 500°C aging is not enough to yield a significative precipitation hardening, most probably due to the fact that this temperature is too low to promote a fast diffusion of the copper atoms for the formation of the hardening clusters. For the other side, 700°C is too high a temperature: the precipitation hardening was replaced for treatment times above 5 minutes by a simultaneous tempering of the bainitic matrix and overaging. The temperature of 600°C was the best balance for the aging treatment, promoting a fast increase in the hardness values up to its maximum value, 271 HV, after 30 minutes of treatment. Figure 2 also shows that the hardness curve along aging time for the HSLA-80 steel showed only one peak, independently of the treatment temperature used. This fact indicates that both precipitating phases, the Cu-rich ε and Nb carbonitride, are precipitating at the same temperature range. This fact was previously reported in the literature [12].

The hardness evolution of the ULCB steel during the aging treatments was faster than observed for the HSLA-80 steel, probably due to the different of precipitates involved. In the ULCB steel there is only precipitation of Nb(CN). The aging of this alloy at 500°C promoted a continuous hardness increase, from the starting value of 260 HV up to 295 HV after a two-hour treatment.The increase of the temperature of the aging treatment to 600°C accelerated significantly the hardness response of the material, which increased from 260 to 286 HV after only five minutes of treatment. From this point on hardness increased lightly, reaching a 291 HV value after one hour of aging. After this point hardness stabilized; after two hours of treatment the sample showed almost the same value: 290 HV. For its turn, the behavior of the ULCB samples during aging under maximum temperature (700°C) was similar to that observed for the HSLA-80 steel: a slight hardness increase after 5 minutes of treatment, from 260 to 265 HV, that was followed by a decrease for longer times: after two hours of aging the hardness fell to only 243 HV, a value lower than the observed in the as rolled sample. In this latter case apparently occurred the same mechanism observed in the HSLA-80 samples aged at 700°C, that is: the precipitation hardening was counterbalanced by tempering of the bainitic matrix and by precipitate overaging.

When one compares the behavior of the steels studied in this work it is interesting to note that the ULCB steel was slightly higher than HSLA-80 in the as-rolled state: 260 HV versus 244 HV, respectively. This hardness difference increases after an one hour-aging treatment at 600°C: 291 HV versus 268 HV respectively. This behavior is opposite to what was observed during the determination of the CCT diagrams, where the hardness of the HSLA-80 samples was greater than the measured for the ULCB samples. This result can be explained by the fact that the samples submitted to the aging treatments were reheated to 1200°C before hot rolling, which promoted a major solubilization of the microalloying elements of the ULCB steel, specially niobium. Under these conditions the synergistic effect between this element and boron can be fully used to increase austenite hardenability [13,14].

4 CONCLUSIONS

The CCT diagrams determined for HSLA-80 and ULCB steels showed that both alloys formed a bainitic microstructure along a wide range of cooling rates, because their alloy content is relatively high. The HSLA-80 samples used for the determination of the CCT diagram were slightly harder than its counterparts of ULCB steel. This result can be attributed to the slightly higher C content of the first alloy and the low austenitizing temperature used, which contributed to a incomplete solubilization of microalloying elements in both steels.

Both alloys showed a significant hardness increase when submitted to aging treatments at 600°C. This can be attributed to precipitation hardening in the bainitic matrix. In the case of the HSLA-80 steels the precipitates are supposed to be Cu plus Nb carbonitride, while in the ULCB steel only the latter constituent is present.

The speed of aging is clearly dependent of the temperature used in this treatment. At 500°C this speed was minimal. At intermediate temperature, 600°C, there was an optimized balance between precipitation speed and maximum hardening. At 700°C the carbonitride precipitation apparently is too fast, promoting overaging. Its implies in loss of the precipitation hardening that, associated with the simultaneous tempering of the bainitic matrix, led to significant softening in both steels.

The precipitation hardening promoted by the aging treatment was faster for the ULCB steel, where only niobium carbonitride precipitated during this step.

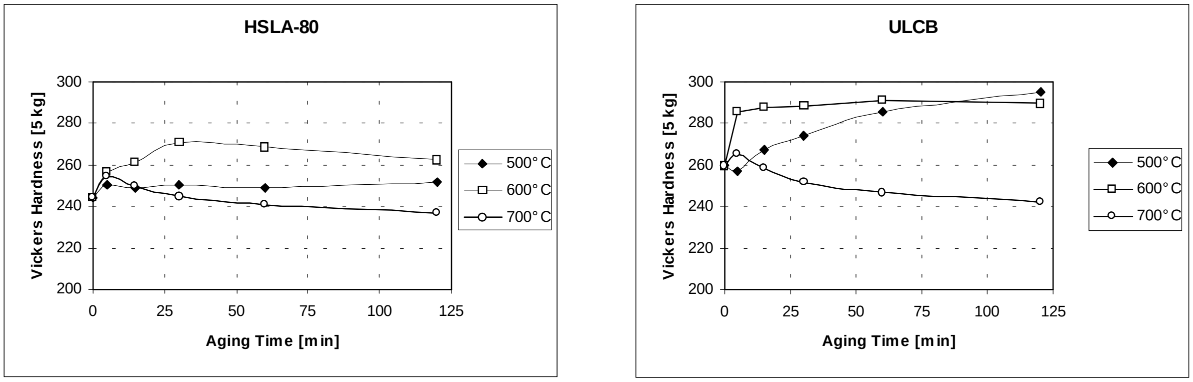

Figure 2: Hardness evolution along aging time for the alloys studied in this work. The aging treatments were carried out under temperatures of 500, 600 and 700° C.

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