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Strength increase in the coarse-grained heat-affected zone of a high-strength, blast-resistant steel after post-weld heat treatment

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ABSTRACT

The microstructure and precipitation behaviors in the coarse-grained heat-affected zone (CGHAZ) of a blast-resistant steel during post-weld heat treatment (PWHT) were investigated. It was found that using PWHT at 650 °C for 1 h is beneficial for the strength recovery in the softened CGHAZ in the as-welded condition. This is attributed to the re-precipitation of strengthening phases during the PWHT process that are dissolved in the CGHAZ during heating to the high temperature and do not re-precipitate completely during cooling. The strength change in the CGHAZ is therefore related with the change in microstructure after PWHT and rationalized by the model used in the present study.

1. Introduction

A high-strength, blast-resistant steel, BlastAlloy 160 (BA-160) was developed by Northwestern University to meet the requirement of the US Navy for structural applications for surface ships. This alloy can achieve a yield strength of 1100 MPa (160 ksi) and impact toughness of 176 J at 25 °C following appropriate treatment. The high strength results from a martensite/bainite microstructure with additional strengthening provided by nanometer-sized Cu-rich precipitates and M_2C carbides (where M = Cr, Mo and V). High toughness is a result of finely dispersed Ni-stabilized austenite, based on a dispersed phase transformation toughening mechanism. Details on the design and microstructure development of BA-160 can be found in publications by Saha et al. [1,2].

Because shipbuilding is heavily reliant on welding as a primary fabrication technique, any high performance naval material must also possess good weldability. However, the carefully designed base metal microstructure is inevitably altered in the heat-affected zone (HAZ) as a result of the weld thermal cycle, resulting in mechanical properties that may not be optimum in this region [3,4].

As Cu and M_2C carbides precipitates play an important role in strengthening BA-160, research has been conducted to investigate their stability and evolution in HAZ during continuous heating and cooling as a result of the welding process. Yu et al. [5,6] investigated microstructural evolution in four regions of the BA-160HAZ in the as-welded condition, and reported the coarsening of Cu precipitates in the subcritical HAZ (SCHAZ), coarsening and partial dissolution in the intercritical HAZ (ICHAZ), almost complete dissolution in the fine-grained HAZ (FGHAZ), and dissolution and possible re-precipitation of Cu in the coarse-grained HAZ (CGHAZ). In addition, martensite morphology differs in the four HAZ regions, which was shown by electron backscatter diffraction (EBSD) analysis. Because of the variation in Cu precipitation and martensite morphology existing in different HAZ regions, softening occurs in the CGHAZ, FGHAZ and SCHAZ as compared to the base metal, with the CGHAZ suffering the most serious strength loss. The softened CGHAZ in the as-welded condition is therefore the “weak link” in the entire weldment and limits the achievable mechanical properties. Therefore, in order to optimize weldment mechanical properties (strength), the strength loss in the CGHAZ must be recovered using an appropriate postweld treatment.

Besides the CGHAZ softening problem, HAZ hydrogen-induced cracking (HIC) is another important weldability issue for BA-160, which must be alleviated by optimizing the welding process. When designing the alloy composition for BA-160, its carbon content was intentionally constrained at 0.05 ± 0.01 wt%, aiming to obtain good resistance to hydrogen-induced cracking. However, due to its relatively high alloy content (especially Ni, Cr and Cu), the hardenability of BA-160 is relatively high, and martensite with coarse prior austenite grain size (reaching up to 100 μm) readily forms in the CGHAZ in a wide range of cooling rates after welding [7].

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A recent study [8] has shown that BA-160 exhibits a relatively high susceptibility to HIC in the CGHAZ. One method to mitigate this is to perform a postweld heat treatment (PWHT) which is beneficial in both relieving residual stresses and reducing diffusible hydrogen level. Historically, this has been a common practice to alleviate susceptibility to HIC in high strength steels [9].

This investigation was designed to determine if a standard, code-approved PWHT is capable of improving the overall weldment strength, as well as reducing susceptibility to HIC of BA-160. The microstructure of the CGHAZ after PWHT was studied in the present work, and was compared to that of the as-welded CGHAZ. A strengthening model was used to rationalize the strength change in the CGHAZ due to the use of PWHT as compared to the as-welded condition.

2. Material and experimental procedures

The base metal microstructure and chemical composition of BA-160 is shown in Fig. 1 and Table 1, respectively. Samples for thermal simulation were machined into a cylindrical geometry 6.5 mm in diameter and 100 mm in length. A Gleeble 3800™ was used to simulate the weld thermal cycle of the CGHAZ. Testing was done in a partial vacuum of approximately 10^{-3} Torr. The sample was heated to a peak temperature of 1300 °C (representative of the CGHAZ temperature) with heating rate of 100 °C/s, and free cooled to room temperature. The $\Delta t_{8/5}$, which is the cooling time from 800 °C to 500 °C, was approximately 15 s.

Fig. 1. BA-160 base metal microstructure.

It is recommended [10,11] that in order to alleviate HIC, a proper PWHT should be performed in the temperature range of 600–650 °C. The soak time is usually 1 h/in. (60 min/25 mm) of steel thickness. Therefore, a PWHT of 650 °C for 1 h was chosen in the present study. After the welding simulation, the sample was heated to 650 °C at 100 °C/s, and held there for 1 h, to simulate the PWHT thermal cycle after welding.

After the welding and subsequent PWHT thermal cycle simulation was completed, a metallographic specimen was cut from the cross-section of the test sample where the thermal couple (used to measure temperature) was attached, and was subjected to a standard metallographic sample preparation procedure. EBSD map was obtained using a Philips ESEM FEG-30 scanning electron microscope equipped with an EBSD camera. An operating voltage of 20 kV and a spot size of 5 nm with a scanning step size of 0.1 μm were used. Microhardness was measured on the as-polished sample surface utilizing a Leco M-400-H1 hardness testing machine employing a load of 1 Kg, in accordance with the ASTM E 384-10. Ten measurements were taken to determine the average Vickers hardness value. An FEI Helios NanoLab™ 600 DualBeam (FIB/SEM) system was used to prepare the transmission electron microscope (TEM) foil from the polished test sample for TEM analysis. The extracted TEM foils were loaded in a Philips CM200 microscope for diffraction analysis and imaging. The FEI Tecnai F20 S/TEM microscopy was used to analyze the elemental composition of precipitates.

3. Results and discussion

The BA-160 CGHAZ in the as-welded condition was studied by Yu et al. [5]. The present work was designed to characterize the microstructure and study the precipitation behavior in the CGHAZ after PWHT. Therefore a comparison between the two conditions can be made and the strength change can be

Table 1
Chemical composition of BA-160 steel (wt%).

C	Mn	Si	P	S	Cu	Ni	Cr	Mo	V	Nb	Ti	Fe
0.059	0.001	0.015	< 0.005	< 0.001	3.39	6.8	1.9	0.61	< 0.001	< 0.001	0.016	Bal.

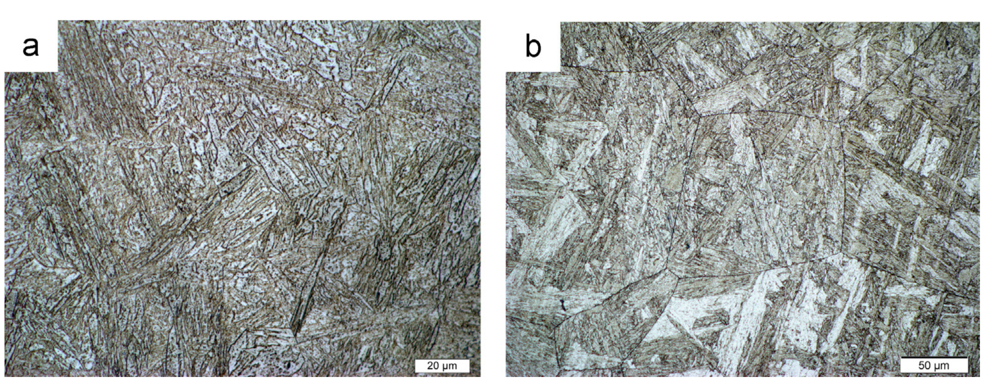

Fig. 2. Optical micrographs for the BA-160 CGHAZ: (a) as-welded and (b) after PWHT.