Strength Recovery in a High-Strength Steel During Multiple Weld Thermal Simulations

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BlastAlloy 160 (BA160) is a low-carbon martensitic steel strengthened by copper and M_2C precipitates. Heat-affected zone (HAZ) microstructure evaluation of BA160 exhibited softening in samples subjected to the coarse-grained HAZ thermal simulations of this steel. This softening is partially attributed to dissolution of copper precipitates and metal carbides. After subjecting these coarse-grained HAZs to a second weld thermal cycle below the A_{c1} temperature (at which austenite begins to form on heating), recovery of strength was observed. Atom-probe tomography and microhardness analyses correlated this strength recovery to re-precipitation of copper precipitates and metal carbides. A continuum model is proposed to rationalize strengthening and softening in the HAZ regions of BlastAlloy 160.

DOI: 10.1007/s11661-011-0707-y

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I. INTRODUCTION

TO meet the rigorous requirements for the United States Navy hull and deck application, a blast resistant steel, BlastAlloy 160 (BA160), was developed at Northwestern University.^[1,2] BA160 is a low carbon martensitic steel additionally strengthened primarily by nanometer-sized Cu-rich precipitates and M_2C precipitates (where $M = Cr, Mo, \text{ and } V$). The yield strength of BA160 is 160 ksi (1104 MPa). In addition to high strength, very good room-temperature Charpy impact toughness [176 J (130 ft-lb)] was achieved through precipitation of Ni-stabilized austenite within a martensitic matrix. The aim of BA160 development is to replace currently certified high-strength low-alloy (HSLA) steels used in surface ship structure. Typical shipyard welding procedures include gas metal arc welding (GMAW), submerged arc welding (SAW), shielded metal arc welding (SMAW), and flux cored arc welding (FCAW).^[3] These welding procedures will also be used for deploying BA160. To employ BA160 in ship building applications, it is important to understand its weldability, which is related to, among other factors, the complex solid-to-solid phase transformations that occur in heat-affected zone (HAZ) regions as a function of initial microstructure, heating rate, peak temperature (T_p), and cooling rate.

There are many testing methodologies available to determine the weldability of structural materials and selection of appropriate tests depends on the material.^[4] Although weldability tests are not often routinely used when optimizing mechanical properties such as tensile strength and fracture toughness during alloy development, they are of great importance during weld procedure development to ensure proper deployment in service. Weldability testing techniques, including the HAZ thermal simulations, hot ductility testing, and reheat cracking testing, were employed to evaluate the weldability of BA160.^[5] For steels, there are four distinct HAZ regions depending upon the peak temperature in a given weld thermal cycle. (1) In subcritical HAZ regions (SCHAZ; $T_p < A_{c1}$), no detectable transformation of ferrite to austenite occurs. (2) In the intercritical HAZ (ICHAZ), partial transformation of ferrite to austenite occurs because the peak temperature is between A_{c1} and A_{c3} . (3) In the fine-grained HAZ (FGHAZ), the samples are heated to a temperature slightly above the A_{c3} temperature, after complete austenitization. (4) In the coarse-grained HAZ (CGHAZ), the sample in the austenite state is heated to a peak temperature significantly above the A_{c3} . Prior research on four HAZ regions of BA160 steels, using a GLEEBLE*

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Manuscript submitted November 22, 2010.

Article published online April 27, 2011

*GLEEBLE is a trademark of Dynamic Systems Inc., Poestenkill, NY.

thermomechanical simulator, showed no strength change in the SCHAZ, some hardening in the ICHAZ, and softening in both the FGHAZ and CGHAZ. Atom-probe tomography (APT) characterization and strengthening analyses correlated the softening to the dissolution of Cu precipitates and carbides.^[6] To validate the preceding thermal simulations, gas tungsten arc (GTA) spot welds were made on BA160 samples.^[5] Vickers microhardness traverse across spot weld HAZs showed a

Table I. Chemical Composition of the BA160

Element	C	Si	Mn	Cu	Ni	Cr	Mo	Ti	Fe
Wt pct	0.059	0.015	0.001	3.390	6.800	1.900	0.610	0.016	balance
At pct	0.277	0.030	0.001	3.005	6.527	2.058	0.358	0.019	balance

Table II. Aging Treatment Procedure for BA160 Steel

Step	Temperature [K (°C)]	Duration	Poststep Procedure
1. Austenitization	1173 (900)	1 h	water quench
2. Liquid nitrogen hold	87 (−196)	30 min	air warm to room temperature
3. Tempering	823 (550)	30 min	water quench
4. Tempering	723 (450)	5 h	air cool to room temperature

similar phenomenon as that observed from GLEEBLE samples. Softening is shown close to the fusion line (*i.e.*, CGHAZ) and hardening is observed in the ICHAZ. These results confirmed the presence of four distinct HAZ regions in the spot welds.

Based on the preceding results, thermal simulations were considered to be an effective tool to evaluate the HAZ regions of BA160. The fusion zone of GTA spot welds contained cellular/dendritic solidification substructure with an average hardness of 364HV. Since both microstructure and hardness are not favorable in the fusion zone, consumables for BA160 have to be designed in the future. The focus of the current article is to develop methods to mitigate the softening in the HAZ region.

It is conceivable that postweld heat treatment (PWHT) could reinstate strength in the HAZ. For example, autogenous spot welds in the BA160 sample were postweld heat treated at 823 K (550 °C) for 30 minutes or 723 K (450 °C) for 5 hours.^[5] Strength recovery was observed in the fusion zone and HAZ. The observed strength increase was assumed to be the result of re-precipitation during the PWHT. The PWHT is not feasible for naval applications due to the size or complexity of the structures that are welded. Alternatively, strength recovery in the HAZ (or weld metal) could be achieved by designing an appropriate multipass welding procedure that leads to a similar extent of re-precipitation during PWHT. Most of the HSLA structural steel plates currently used in shipyards are in the range of 6 to 30 mm (0.25 to 1.25 in.) with yield strength from 448 to 689 MPa (65 to 100 ksi). The thickness of heavier gage plate could be as high as 83 mm (3.25 in.). Though the plate thickness of BA160 will be reduced due to the high yield strength, the envisioned thickness for naval applications (>6 mm) is still too high for single pass welding. As a result, multipass welding is required for most applications.

For some HAZ regions (CGHAZ and FGHAZ), the first thermal cycle with high peak temperature ($>A_{c3}$) causes the dissolution of precipitates. Subsequent thermal cycles with low peak temperatures (773 K to 923 K (500 °C to 650 °C)) may trigger re-precipitation. It is

Fig. 1—Thermal profile of simulated double thermal cycle HAZ.

realized that the thermal history of multipass HAZ is complex and depends on many variables, such as plate thickness, welding procedure, and field condition. The current article only focuses on validation of the hypothesis of re-precipitation and strength recovery during the second thermal cycle in multipass welding. Thermal simulations and microstructural characterizations are used to prove the hypothesis. In addition, an appropriate strengthening model is used to rationalize the results.

II. EXPERIMENTAL

The BA160 experimental material was received in the form of 35-mm-diameter bar stock from QuesTek Innovations LLC (Evanston, IL). The chemical composition and heat treatment procedure for BA160 are listed in Tables I and II. To investigate the multipass HAZ microstructural evolution, thermal-cycle simulations representative of the various HAZ regions were performed. For actual welding conditions, the thermal history is complicated. The thermal profile for a specific location may contain several thermal cycles, depending on the number of passes. Current research pertains to only re-precipitation during the second thermal cycle with a peak temperature lower than A_{c1} . As a result, four single-pass and simplified multipass HAZ regions are selected (Figure 1): (1) CGHAZ with high heat input, (2) CGHAZ + 923 K (650 °C) reheat with high heat input, (3) CGHAZ with low heat input, and (4) CGHAZ + 923 K (650 °C) reheat with low heat input. The $\Delta t_{8/5}$ (time to cool from 1073 K to 773 K (800 °C to 500 °C)) was approximately 45 seconds for high heat input and 15 seconds for low heat input. The thermal cycle simulations were performed with a GLEEBLE 3800 thermomechanical simulator using solid cylindrical samples of 5 mm in diameter and 101.6 mm in length. The sample temperature was controlled with a type-K

thermocouple, which was percussion welded at the midsection of the sample. The samples were heated to the peak temperature at a rate of $100\text{ }^{\circ}\text{C s}^{-1}$. The simulations were conducted with the test chamber evacuated to a pressure of $1.3 \times 10^{-4}\text{ Pa}$ to limit surface oxidation and thermocouple detachment. Phase transformation strains were measured using a dilatometrical dilatometer.

The microhardness of the simulated HAZ regions was determined using a LECO** M-400-H1 microhardness-

**LECO is a trademark of LECO Corporation, St. Joseph, MI.

testing machine employing a 9.81 N (1 kgf) load, in accordance with ASTM Designation E-384-08.^[7] Time of load is 10 seconds. Specimens used for the microhardness testing and microstructure characterization were taken from a single test sample to ensure consistent results.

For APT samples, coupons ($0.3 \times 6 \times 6\text{ mm}^3$) were cut from the center of the Gleeble specimens using a LECO VC-50 precision diamond saw. APT tip blanks ($0.3 \times 0.3 \times 6\text{ mm}^3$) were cut from the coupons and electropolished using a two-step method.^[8] Initial electropolishing was performed with a solution of 10 vol pct perchloric acid in acetic acid at 10 to 25 Vdc at room temperature. This was followed by a manually controlled pulsed final-polishing step using a solution of 2 vol pct perchloric acid in butoxyethanol at 10 to 25 Vdc at room temperature, producing a tip with a radius of $<50\text{ nm}$. The APT data were collected using the Cameca local-electrode atom-probe (LEAP)[†]

[†]LEAP and Cameca are trademarks of CAMECA Instruments, Inc., Madison, WI.

tomograph at the Northwestern University Center for Atom-Probe Tomography (NUCAPT, Evanston, IL). The data were acquired at a specimen temperature in the range of 75 to 85 K under ultrahigh vacuum conditions of approximately $1.0 \times 10^{-8}\text{ Pa}$. Short-duration laser pulses (50 pJ pulse^{-1}) are used to induce evaporation of ions at a pulse repetition rate of $5 \times 10^5\text{ Hz}$ and a target evaporation rate (ions pulse^{-1}) of 0.5 to 1 pct. The maximum dc voltage employed was 12 kV. Then, the acquired atomic position data are calibrated and three-dimensional reconstruction is created using Cameca's IVAS[‡] program.

[‡]IVAS is a trademark of CAMECA Instruments, Inc., Madison, WI.

To characterize the martensite matrix in BA160, a PHILIPS[§] ESEM FEG-30 scanning electron micro-

[§]PHILIPS is a trademark of FEI Company, Hillsboro, OR.

scope equipped with an electron backscatter diffraction

Fig. 2—Relative sample radius change and temperature trace during (a) coarse-grained HAZ + 650 with high heat-input and (b) coarse-grained HAZ + 650 with low heat-input thermal cycles from dilatometry experiment.

(EBSD) camera was used, whose accelerating voltage is 20 kV and spot diameter is 5 nm, with a scanning step size of $0.1\text{ }\mu\text{m}$. EBSD maps were analyzed using OIM^{§§}

^{§§}OIM is a trademark of EDAX, Mahwah, NJ.

Analysis Software.

III. RESULTS

A. GLEEBLE Simulation and Microhardness Measurement

The experimentally measured thermal profiles from both high heat-input and low heat-input thermal cycles are shown in Figure 1. The samples were allowed to cool to room temperature from 1573 K (1300 °C) during the initial HAZ thermal cycle, prior to reheating to a peak temperature of 923 K (650 °C). The measured dilatation data during cooling from 1573 K (1300 °C) to room temperature, heating to 923 K (650 °C) from room temperature, and cooling from 923 K (650 °C) to room temperature are presented in Figure 2. On cooling, data for both low (Figure 2(a)) and high heat-input (Figure 2(b)) conditions show the martensite start

Table III. Microhardness of As-Received Materials and Different HAZ Regions (Average Hardness of 10 Measurements)

Condition	As-Received	CGHAZ High Heat Input	CGHAZ + 650 High Heat Input	CGHAZ Low Heat Input	CGHAZ + 650 Low Heat Input
Vickers microhardness	402	340	405	351	404

Fig. 3—3-D LEAP tomographic reconstruction displaying Cu atoms for (a) coarse-grained HAZ with high heat input, (b) coarse-grained HAZ + 650 with high heat input, (c) coarse-grained HAZ with low heat input (d) coarse-grained HAZ + 650 with low heat input.

transformation temperature to be approximately 629 K (356 °C). As expected, no phase transformation was detected during the reheat thermal cycle because the peak temperature (923 K (650 °C)) is lower than A_{c1} . This is indicated by the absence of inflection in the dilatation curve. Vickers microhardness data for these samples are presented in Table III. For single-pass HAZ simulation samples (coarse-grained HAZ with high and low heat input), the microhardness is about 340 HV, which is much smaller than that of the as-received base metal, 402 HV. For double-pass HAZ samples, the microhardness is approximately 405 HV. This result demonstrated that our hypothesis for strength recovery in CGHAZ by double-pass thermal cycles is indeed

valid. The magnitude of this recovery is similar to that of samples subjected to PWHT.^[5] Although this strength recovery is believed to be the by-product of re-precipitation reactions, it is important to confirm this hypothesis.

B. Microstructure Characterization

1. Cu precipitate characterization

The evolution of Cu-rich precipitates in steels is studied extensively by APT.^[9–11] Atomic maps of copper from APT analyses for all samples are summarized in Figure 3. The Cu-enriched regions are not observed in samples subjected only to CGHAZ simulations

Table IV. Radius and Number Density of Cu Precipitates for As-Received Material and Different Simulated HAZ Regions

	As-Received	CGHAZ High Heat Input	CGHAZ + 650 High Heat Input	CGHAZ Low Heat Input	CGHAZ + 650 Low Heat Input
Radius (nm)	2.4 ± 1.2	1.7 ± 0.5	2.4 ± 1.5	1.9 ± 1.4	2.6 ± 1.6
Number density (10 ²³ m ⁻³)	4.2 ± 2.2	2.0 ± 1.1	9.4 ± 3.1	2.0 ± 1.4	6.4 ± 2.5

Table V. Composition of Martensite Matrix* (Atomic Percent) Measured by Atom Probe

	C	Cr	Fe	Ni	Mo	Cu
CGHAZ high heat input	0.11 ± 0.001	1.94 ± 0.003	87.84 ± 0.007	6.73 ± 0.006	0.36 ± 0.001	2.86 ± 0.004
CG HAZ + 650 high heat input	0.05 ± 0.001	1.98 ± 0.007	90.28 ± 0.015	6.43 ± 0.012	0.38 ± 0.003	0.85 ± 0.004
CGHAZ low heat input	0.37 ± 0.003	1.95 ± 0.006	86.33 ± 0.016	7.82 ± 0.012	0.48 ± 0.003	2.95 ± 0.008
CGHAZ + 650 low heat input	0.16 ± 0.002	2.15 ± 0.008	89.45 ± 0.017	6.73 ± 0.014	0.37 ± 0.003	0.90 ± 0.005

*Due to the envelop method used in this analyses, the martensite matrix may contain a small amount of metal carbide and retained austenite.

(Figures 3(a) and (c)). The Cu atoms are homogenously distributed for both the high and low heat-input conditions. In contrast, the Cu-rich precipitates are clearly seen in the samples subjected to a second thermal cycling (Figures 3(b) and (d)). To obtain quantitative information about the Cu-rich precipitates, a cluster-search algorithm (envelope method) was employed.^[8] The following parameters were used: (1) maximum Cu atom separation = 0.6 nm, (2) minimum number of Cu atoms in a cluster = 30, and (3) grid resolution = 0.12 nm. The sensitivity of the parameters is discussed in Reference 1. The number density, N_v , is calculated using

$$N_v = \frac{N_p \zeta}{n \Omega} \quad [1]$$

where N_p and n are the number of precipitates and the total number of atoms detected in the volume, respectively; Ω is the average atomic volume, $1.2 \times 10^{-29} \text{ m}^{-3}$ for bcc Cu; and ζ is the detection efficiency of the microchannel plate detector, which is taken to be 0.5. The summary of the results for the Cu precipitates is given in Table IV. Cu precipitates found using the envelope method are extracted and the rest of the atoms are considered as martensite matrix (may contain small amounts of metal carbide and retained austenite). Table V shows the composition of the martensite matrix.

Even though Cu-enriched regions cannot be seen in the CGHAZ for both high heat input and low heat input, Cu precipitates with small radii are still detected by the cluster search algorithm. We note that the size and number density of Cu precipitates are strongly correlated with the parameters chosen for the algorithm.^[12] The small precipitates with average radius of 1.7 nm in the CGHAZ could represent artifacts produced using the cluster search algorithm. The emphasis of this study, however, is to compare the characteristics of Cu precipitates between a single thermal-cycle HAZ and double thermal-cycle HAZ. The absolute properties of Cu precipitates are considered to be of secondary importance.

Fig. 4—Major alloy element composition profile across the copper precipitates boundary in (a) coarse-grained HAZ + 650 with high heat input and (b) coarse-grained HAZ + 650 with low heat input.

The average radius and number density of Cu precipitates in heat-treated BlastAlloy 160 are 2.4 nm and $4.2 \times 10^{23} \text{ m}^{-3}$, respectively. The radius and number density of Cu precipitates for single-pass CGHAZ in Table IV indicate almost full dissolution of Cu precipitates. Compared with single-pass CGHAZ, double-pass HAZ samples contain larger Cu precipitates and higher number density. These Cu precipitates should have formed within the martensite matrix during the second thermal cycle. Besides the precipitation reaction, the Cu

Fig. 5—1 pct carbon (plum) and 8 pct copper (golden) isoconcentration surface from (a) coarse-grained HAZ with high heat-input sample and (b) coarse-grained HAZ + 650 with high heat-input sample. (c) Composition profile across the metal carbide boundary shown by the arrow in (a).

concentration in the martensitic matrix decreases from 2.86 to 0.85 at. pct after the second thermal cycle for high heat-input condition (Table V). The preceding result confirms that the re-precipitation reaction occurred during the second thermal cycle.

To evaluate the strengthening caused by these precipitates, the elemental composition within the precipitates and at the matrix/precipitate interfaces has to be

characterized. Proxigram concentration profiles^[13] of iron, nickel, and copper across one representative precipitate (CGHAZ + 650 with high heat-input and low heat-input samples) are displayed in Figure 4. The copper concentration in the precipitate core is about 66 at. pct. This concentration is smaller than that of the core concentration in precipitates for the peak-aged condition (> 90 at. pct).^[14]

Fig. 6—Measured EBSD images from (a) coarse-grained HAZ with high heat input and (b) coarse-grained HAZ with low heat input.

Table VI. Packet Size, Block Width, Slip-Plane Length, and for Different HAZ Regions

Condition	Packet Size, d_p (μm)	Block Width, d_b (μm)	Slip Plane Length, M (μm)
CGHAZ high heat input	56.0 ± 7.5	12.4 ± 3.5	25.3 ± 5.0
CGHAZ + 650 high heat input	56.0 ± 7.5	12.4 ± 3.5	25.3 ± 5.0
CGHAZ low heat input	52.3 ± 7.2	11.9 ± 3.4	24.0 ± 4.9
CGHAZ + 650 low heat input	52.3 ± 7.2	11.9 ± 3.4	24.0 ± 4.9

Both iso-concentration surfaces corresponding to 1 at. pct C and 8 at. pct Cu for CGHAZ and CGHAZ + 650 with high heat input are displayed in Figure 5. Only faint precipitate-like features could be visualized in the data from CGHAZ. In contrast, the iso-concentration surfaces from the CGHAZ + 650 high heat-input sample (Figure 5(b)) revealed copious Cu precipitates and metal carbides. One of the precipitates was selected for further analysis (indicated by the bold arrow in Figure 5(b)). Proxigram concentration profiles across the metal carbide interface are plotted in Figure 5(c). The metal carbide is enriched with Cr and Mo. Since the C concentration in the core of the carbide precipitates is approximately 30 at. pct, this metal carbide is most likely M_2C type (where $\text{M} = \text{Fe}, \text{Cr}, \text{Mo}$). The Cu is depleted within the carbide precipitate and enriched near the carbide/martensite interface. This observation supports the published report of co-location of Cu precipitates and Mo/Cr-enriched metal carbide^[14] for the peak-aged condition.

C. Martensite Substructure

Previous study on HAZ properties of BA160 found, besides Cu precipitate strengthening, that martensite substructure plays an important role in strengthening.^[6] Fine martensite substructure contributes to the hardening in the ICHAZ. The current investigation also employed the EBSD technique to study the morphology and crystallography of the martensitic matrix. The

martensite substructure of a single-pass BA160 HAZ was reported.^[6] Two EBSD maps of CGHAZ with high

^{*}There is a typographical error in Ref. 5 regarding to slip plane length of coarse-grained HAZ. In this article it is reported as $2.53 \mu\text{m}$. The correct slip plane length for coarse-grained HAZ is $25.3 \mu\text{m}$.

heat input and low heat input are shown in Figure 6. Since dilatation measurements showed no reverse transformation (martensite to austenite) during the second thermal cycle, martensite morphology is assumed similar to that of CGHAZ samples. Therefore, the martensite substructures in CGHAZ + 650 high heat-input and CGHAZ + 650 low heat-input samples were not measured. The characteristics of the martensite substructure are summarized in Table VI.

IV. DISCUSSION

A. Precipitate Evolution in Heat-Affected Zone

APT results show re-precipitation of Cu and metal carbide precipitates during the second heat treatment of the double pass HAZ. In the HAZ of BA160, the solid-state transformations are complex due to on-heating transformation of martensite to austenite, precipitate growth and coarsening, precipitate dissolution, reformation of precipitates during cooling, and on-cooling transformation of austenite to martensite. A schematic

illustration of phase transformation stages^[6] in the CGHAZ sample is shown in Figure 7. The Cu solid solubility in austenite and ferrite, evaluated with THERMO-CALCTM software using the TCFE5

TMTHERMO-CALC is a trademark of Thermo-Calc, Stockholm.

thermodynamic database, is also shown in Figure 7.^[15] During the first CGHAZ thermal cycle, when the material is heated from P1 to P4, martensite transforms to austenite and all precipitates dissolve in the austenite matrix. On cooling from P4 to P5, due to the high solubility of Cu and C in austenite and a short cooling period, re-precipitation is limited. Below M_s , precipitation is not expected because the diffusivities of Cu and C are small. During the second thermal cycle, on heating to a temperature below A_{c1} , the nucleation and growth of Cu and metal carbide precipitates is expected within the martensite matrix. Figure 7(a) shows that the solubility of Cu in ferrite at 923 K (650 °C) is about 0.7 at. pct. Since the Cu concentration in the material is 3.3 at. pct, Cu is still supersaturated at this temperature. High Cu diffusivity and supersaturation of Cu promote the nucleation and growth of Cu precipitates.

Most Cu-bearing steel studies are focused on isothermal aging. Only limited articles report Cu evolution during nonisothermal conditions. It is well known that during aging of Fe-Cu or Fe-Cu-Ni alloys, the transformation sequence for Cu precipitates is bcc → 9R → 3R → fcc with increasing aging time. However, the kinetics of Cu precipitate evolution is slow. During

welding, the HAZ only experiences a short period of heating, which is not enough to bring the Cu precipitate to overage condition. The single thermal cycle HAZ simulation confirmed no significant overaging in simulated HAZ even when the peak temperature is higher than A_{c1} . In multipass HAZ, however, if the peak temperature of a subsequent pass is higher than A_{c3} , the dissolution of precipitates is expected and strength loss will occur. The welding procedure of BA160 is still under development, and detailed thermal histories in multipass HAZs are not available. Therefore, the precipitations after many (>2) thermal cycles are not discussed here. By comparing the results between high heat input and low heat input, one can conclude that precipitation is not sensitive to heat input. In samples subjected to both high and low heat-input conditions, atom probe results show similar Cu precipitate sizes and number densities. This again suggests that the welding procedure window for BA160 may be wide and conducive to field welding conditions.

B. Modeling of Strengthening Mechanism

A previous study^[6] proposed a strengthening model by considering Cu precipitation and martensite substructure strengthening to interpret the hardening and softening phenomena in a simulated HAZ of BA160 (Appendix). However, due to the dissolution of metal carbide, this strengthening model for a single-pass HAZ did not consider the metal carbide contribution. Since metal carbides reprecipitate during the second thermal cycle, the strengthening contribution from metal carbide precipitates should be taken into account. Precipitates are bypassed by the Orowan dislocation looping mechanism, which is dominant for hard precipitates in soft

Fig. 7—(a) Predicted solubility of Cu in austenite and ferrite is displayed as a function of temperature. (b) Schematic illustration of different stages in microstructure evolution in coarse-grained HAZ region.^[6]

Table VIII. Strengthening Contributions in Different HAZ Samples (σ_p —Strengthening by Cu Precipitates, σ_g —Strengthening by Martensite Lath Boundaries, σ_c —Strengthening by Metal Carbides, and $\Delta\sigma$ —Overall Strengthening)

Condition	σ_p (MPa)	σ_g (MPa)	σ_c (MPa)	$\Delta\sigma$ (MPa)
SCHAZ high heat input	282.5	127.0	0.0	409.5
ICHAZ high heat input	190.0	290.0	0.0	480.0
FGHAZ high heat input	4.8	213.0	0.0	217.8
CGHAZ high heat input	15.8	72.3	0.0	88.1
CGHAZ low heat input	57.3	74.1	0.0	131.4
CGHAZ + 650 high heat input	182.5	72.3	110.9	365.7
CGHAZ + 650 low heat input	176.0	74.1	105.3	355.4

Table VII. Radius and Number Density of M2C Precipitates for As-Received Material and Different HAZ Regions

	CGHAZ High Heat Input	CGHAZ + 650 High Heat Input	CGHAZ Low Heat Input	CGHAZ + 650 Low Heat Input
Radius (nm)	0	1.23 ± 1.1	0	1.19 ± 0.7
Volume fraction	0	0.00023	0	0.00020

Fig. 8—Predicted strength increase due to martensite substructure and copper precipitates compared with experimentally measured microhardness changes in the HAZ regions.

matrix.^[16,17] In this article, the Orowan hardening mechanism is used to predict the contribution of metal carbide to the overall strengthening. The Orowan–Ashby^[18] equation for looping of dislocations between precipitates is given by

$$\tau_c = \frac{0.81Gb}{2\pi(1-\nu)^2L} \ln\left(\frac{2r'}{b}\right) \quad [2]$$

where τ_c is the resolved shear strength increase due to the metal carbide, G is the shear modulus (77 GPa), ν is Poisson's ratio (0.3), L is the effective interprecipitate spacing,^[6] r' is the mean precipitate radius, and b is the Burger's vector of a dislocation (0.25 nm). The properties of metal carbides can be estimated from the APT analysis results listed in Table VII.

By inputting characteristics of the martensite, Cu precipitates, and metal carbides into the strengthening model, the total strengthening contributions can be calculated (Table VIII). Data from the single-pass HAZ^[6] is also included to evaluate current strengthening models. The predicted strength change is plotted as a function of observed microhardness change (Figure 8) for all HAZ regions. In these calculations, the base metal microhardness is used as a reference. A near linear relationship between the predicted strength difference and microhardness difference suggests a validation of the strength model proposed here.

C. Suggestions for Modified Heat Treatment of BA160 Steel

Most of the research on Cu precipitate-strengthened steels has focused on isothermal aging treatments.^[19–21] The aging times are typically from 0.5 to 320 hours. These aging treatments are long and energy consuming. This study suggests an alternative path forward for Cu precipitation in BA160 using a rapid aging treatment. This rapid heat treatment involves traditional austenitizing (Table II) and quenching to liquid nitrogen temperature (78 K (−195 °C)) followed by rapid heating of 373 K (100 °C) per second to 923 K (650 °C). Optimization of these nonisothermal heat treatments will require detailed kinetic modeling.

V. CONCLUSIONS

The softening in the CGHAZ region of BA 160 is partly attributed to the dissolution of Cu and metal carbide precipitates.

Recovery of strength during multipass welding was demonstrated through two-cycle thermal simulations. During the first cycle, on heating the steel above A_{c3} , softening was observed due to precipitate dissolution. During the second thermal cycle, on heating the steel below A_{c1} , re-precipitation of copper precipitates and carbides was promoted. This leads to an increase in hardness from 340 to 405 HV.

A strength model, including precipitation and martensite substructure strengthening, describes the observed softening and hardening in the HAZ simulated samples.

The current study also suggests a rapid nonisothermal aging treatment (heating rate = 100 °C/s and peak temperature = 923 K (650 °C)) for strengthening of the original BA160 steel, instead of long isothermal aging treatments.

ACKNOWLEDGMENTS

We acknowledge financial support from the Office of Naval Research, Drs. J. Christodoulou and W. Mullins, grant monitors. Atom-probe tomographic measurements were performed at the Northwestern University Center for Atom-Probe Tomography (NUCAPT). The authors also thank Professor G. Olson, Northwestern University, for providing the samples. The LEAP tomograph was purchased and upgraded with funding from NSF-MRI (Grant No. DMR-0420532) and ONR-DURIP (Grant Nos. N00014-0400798, N00014-0610539, and N00014-0910781).

APPENDIX

Modeling of Strengthening Mechanisms

For martensitic strengthening, the yield stress of the lath martensite is calculated by summing the contributions from different hardening mechanisms:^[22]

$$\sigma_{YS} = \sigma_0 + \sigma_s + \sigma_p + \sigma_g + \sigma_p \quad [A1]$$

where σ_0 is the friction stress to move dislocations for pure Fe, σ_s is the yield strength increment due to solid-solution hardening, σ_p is the strengthening term due to the forest dislocation density, σ_g is for grain-boundary strengthening, and σ_p is for precipitate strengthening. For the four mechanisms, Cu is the major element varied in the matrix. According to Reference 23, the yield strength increment due to 1 wt pct Cu in solid solution is <50 MPa. Additionally, the dislocation density is a function of M_s and the variation of M_s in different HAZ regions is small.^[6] On the other hand, the strengthening contribution on yield strength from Cu

precipitates and martensite block and packet boundaries could be several hundred mega-Pascals.^[1,22] As a result, in the present study, only precipitation and grain boundary strengthening are considered in the overall strengthening.

Martensite Matrix Strengthening

The packet size and block width in the different regions of the HAZ were measured from the EBSD maps using the mean linear-intercept method.^[24] Most researchers take the contributions from block width and packet size into account separately when considering grain boundary strengthening.^[25,26] The two important parameters, packet size and block width, describe the morphology of lath martensite. Since packet and block boundaries are high-angle boundaries (>15 deg), they should be considered together for grain boundary strengthening. Since the block boundaries inside a packet act as obstacles to dislocation motion, the slip plane length can be described to a certain degree by block width and packet size, which is elaborated in the following content. Assuming that the block morphology is rectangular and the length of the block is equivalent to the packet size, the average slip plane length, M , can be calculated from the block width, d_b , and packet size, d_p .^[25]

$$M = \frac{1}{\pi/2} \left[\int_0^{a \cos(d_b/d_p)} \frac{d_b}{\cos \theta} d\theta + \int_{a \cos(d_b/d_p)}^{\pi/2} d_p d\theta \right] \quad [A2]$$

From the calculated average slip-plane length, the strengthening introduced by block boundaries can be calculated from the Hall–Petch equation:

$$\sigma_g = K_y M^{-1/2} \quad [A3]$$

where K_y is 0.363 MPa m^{1/2}.^[25]

Copper Precipitation Strengthening

Precipitate shearing and precipitate bypass by dislocation looping are two mechanisms generally used to explain precipitate strengthening.^[27,28] Bcc Cu precipitates in martensite matrix are considered as weak precipitates.^[31] The strength contributions arising from Cu precipitation are difficult to model. Many factors, such as misfit strengthening, chemical strengthening, modulus difference strengthening, and dislocation core-precipitate interaction strengthening contribute to the overall Cu precipitate strengthening in steels.^[29] The incorrect equation used to describe modulus difference strengthening by Fine and Isheim^[29] has been pointed out by Liu *et al.*^[30] Since modulus strengthening based upon the model of Russell and Brown^[31] plays the most important role among all the factors, the current focus is on evaluating the component of Cu precipitation strengthening provided by modulus strengthening. The Russell–Brown model assumes that the modulus strengthening effect is due to the relative difference in dislocation energy between the matrix and Cu precipitates

as a result of the modulus difference, as a dislocation passes from the matrix through the Cu precipitate and back into the matrix. The critical shear stress increase caused by the Cu precipitates can be expressed as

$$\begin{aligned} \tau &= 0.8 \frac{Gb}{L} \left[1 - \frac{E_P^2}{E_M^2} \right]^{1/2}; \quad \sin^{-1} \frac{E_P}{E_M} \leq 50^\circ \\ \tau &= \frac{Gb}{L} \left[1 - \frac{E_P^2}{E_M^2} \right]^{3/4}; \quad \sin^{-1} \frac{E_P}{E_M} \geq 50^\circ \\ \frac{E_P}{E_M} &= \frac{E_P^\infty \log \frac{r}{r_0}}{E_M^\infty \log \frac{R}{r_0}} + \frac{\log \frac{R}{r}}{\log \frac{R}{r_0}} \end{aligned} \quad [A4]$$

where G is the shear modulus in the matrix, taken to be 77 GPa; L is the interprecipitate spacing; b is the Burger's vector of dislocation, which is 0.25 nm; E_P is the dislocation energy in the precipitate; and E_M is the dislocation energy per unit length in the matrix. The term E_P^∞ is the dislocation energy per unit length in the precipitate; E_M^∞ is the dislocation energy per unit length in the matrix; r is the average radius of the precipitates; r_0 is the inner cut-off radius, 0.7 nm; and R is the outer cut-off radius, 1000 r_0 . It is assumed that the shear modulus for bcc Cu in the Fe matrix is equivalent to the shear modulus for fcc Cu. For a screw dislocation, the estimated dislocation energy ratio per unit length in an infinite media, E_P^∞/E_M^∞ , is equal to the shear modulus ratio (G_P/G_m). This study used the bulk shear modulus ratio of fcc Cu and bcc Fe, which is 0.6 in the calculation. Substituting the precipitate properties obtained from the atom-probe tomographic analysis into the modulus difference (Eq. [A4]), an estimate of the Cu precipitate strengthening can be provided. It is realized that the Russell–Brown strengthening model is very sensitive to the modulus difference. The measurement of the bcc Cu shear modulus is complicated, since bcc Cu in the α -Fe matrix is metastable and the lattice parameter of bcc Cu is very close to that of bcc Fe. First-principle calculations predicted that the tetragonal shear modulus for bcc Cu could be negative^[30,32,33] Liu *et al.* predictions depended on Fe concentration in the precipitate; the shear modulus of bcc Cu varies from 80 GPa to –20 GPa. If the shear modulus of bcc Cu is positive, the Russell–Brown modulus-mismatch hardening model could be used as a good estimate for bcc Cu precipitate strengthening. Harry and Bacon^[34] proposed an alternative strengthening mechanism. When the unstable bcc Cu precipitate is sheared by a dislocation, the bcc Cu lattice may transform to a close-packed structure. During shearing, strength is increased by the structural-transformation mechanism. This mechanism may be considered for the conditions, which leads to a negative shear modulus of bcc Cu. Since the shear modulus of bcc Cu and the strengthening mechanism are still being debated, this study applies the classic Russell–Brown model and the parameters to estimate the strengthening by Cu precipitates. We note that the model could be improved if an accurate bcc Cu shear modulus and effective interprecipitate distance could be estimated. This will require *in-situ* measurement of the Cu precipitate modulus using *in-situ* diffraction techniques.^[35]

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