The Weld Heat-Affected Zone Liquation Cracking Susceptibility of Naval Steels

Hot-ductility testing was employed to determine the relative cracking susceptibilities of three naval steels and a high-strength, blast-resistant steel recently developed for naval applications

BY J. L. CARON, S. S. BABU, AND J. C. LIPPOLD

ABSTRACT

A high-strength, blast-resistant steel known as BlastAlloy160 (BA-160), recently developed for naval applications, and naval alloys HY-100, HSLA 100, and HSLA-65, were subjected to hot-ductility testing to assess their relative susceptibility to heat-affected zone (HAZ) liquation cracking. The widest liquation cracking temperature range (LCTR) of 148°C was exhibited by BA-160, which suggested moderate cracking susceptibility. Its susceptibility was attributed to several factors including a completely austenitic microstructure at elevated temperatures, enrichment of Ni and Cu along grain boundaries, good wetting and penetration of liquid films along the austenite grain boundaries, and a large HAZ grain size. A slightly wider LCTR of 62°C was exhibited by HY-100 as compared to the HSLA steels. Microstructure characterization suggested the onset of liquation to occur within the austenite single phase and wider grain boundary enrichment was exhibited compared to the HSLA steels. The HSLA steels exhibited similar overall behavior and narrow LCTRs (HSLA-100, 40°C; HSLA-65, 43°C). Their sharp decrease in on-heating (OH) ductility over a narrow temperature range of approximately 10°C was attributed to the formation of δ-ferrite within the same temperature range as incipient grain boundary melting. The OH liquation mechanism in the HSLA steels was directly related to the δ-ferrite and reverse peritectic transformations. Ineffective wetting and liquid discontinuity in the HSLA steels was established through metallographic and fractographic observations.

Introduction

Steel is currently the principal structural materials in existing naval ships and submarines. This will continue to be the trend in the near future, given their relatively low cost, suitable combination of mechanical properties, and existing infrastructure for their processing and fabrication (Ref. 1). As such, the development, certification, and utilization of high-strength steel is an active research area for the U.S. Naval Research Laboratory. Naval steels must exhibit good weldability and maintain adequate properties over a temperature range of approximately −35°C to 50°C (−31°F to 122°F). Two major series of naval alloys in use today are the high-yield (HY) steels and the high-strength low-alloy (HSLA) steels.

The HY steels have been used extensively for ship and submarine construction since the 1950s (Ref. 2). These alloys were developed for thick-plate structural applications and their microstructures typically consist of quench and tempered martensite with a fine distribution of carbides. During welding of HY steels, the formation of brittle martensite in the heat-affected zone (HAZ) makes them susceptible to hydrogen-induced cracking (HIC) (Ref. 3). To limit HIC susceptibility, preheat and weld interpass temperatures are carefully controlled to slow the cooling rate such that martensite is not formed in the HAZ. This action also provides longer time and higher temperatures for hydrogen to diffuse out of the weld, and reduces the tensile residual stress level. However, the required preheat/interpass procedure results in lower productivity and higher fabrication costs. Also, it has been reported that HAZ liquation cracking is often encountered during welding of HY steels for shipbuilding purposes (Ref. 4).

The HSLA steels began their development in the 1980s as replacements for the HY steels (Ref. 5). The incentive for developing the HSLA steels was for improved weldability through the minimization or elimination of preheat, utilization of a wider range of energy inputs, and employment of more relaxed fabrication controls. This was achieved by developing steels containing lower carbon contents than the HY steels, and which obtain their strength through additions of Cu for precipitation strengthening and Cr, Mo, V, Nb, and Ti for carbide strengthening (Ref. 2).

The current research is directed toward developing naval steels with yield strengths of 1030 to 1240 MPa (150–180 ksi) and Charpy V-notch impact toughness values above 115 J (85 ft-lb) measured at an approximate room temperature of 22°C (72°F) (Ref. 6). BlastAlloy160 (BA-160) is a high-strength transformation-toughened alloy that was developed to meet the property requirements for blast-resistant naval material applications (Refs. 6–8). The minimum yield strength of 1103 MPa (160 ksi) is provided by fine (3 to 5 nm) Cu-rich precipitates and M2C carbides (where M represents the elements Cr, Mo, and V) in a mixed martensitic/bainitic matrix. An enhanced impact toughness level of 176 J at a temperature of 20°C is attributed to the dispersed...
The phenomenon, which occurs when Ni-stabilized austenite precipitates undergo a martensitic transformation in a frontal zone ahead of a crack tip (Ref. 9). The strain-hardening effect from the martensite that is formed in advance of the crack tip offsets the softening from microvoid formation and forces strain localization elsewhere, thus delaying the onset of shear localization during ductile fracture (Ref. 10). The heat treatment procedure for BA-160 is provided in Table 1.

Previous investigations have been conducted to characterize the microstructure evolution in the weld fusion zone (FZ) and HAZ of BA-160 (Refs. 11, 12). Characterization of simulated HAZ regions revealed Cu segregation, leading to concerns of HAZ liquation cracking susceptibility (Ref. 13). The metallurgical basis for HAZ liquation cracking involves the simultaneous presence of continuous liquid films at grain boundaries and a critical level of restraint (Ref. 14). These liquid films cannot accommodate the strains that accompany thermal and mechanical restraints during cooling, leading to cracking along the grain boundaries in the partially melted zone (PMZ).

The hot-ductility test is a simple and accurate test method to evaluate the HAZ liquation cracking susceptibility of an alloy (Refs. 15–17). These tests simulate the metallurgical degradation that occurs during a weld thermal cycle and provides a measure of the ductility of the alloy that would be experienced at high temperatures in the HAZ. Ductility is a key component of overall weldability because weld cracking is often associated with an exhaustion of available ductility. To determine the elevated temperature ductility “signature” of an alloy, both on-heating (OH) and on-cooling (OC) ductility tests are performed.

The nil-ductility temperature (NDT) is defined as the OH temperature where ductility is reduced to zero (Refs. 16, 17). Traditionally, this is viewed as the temperature of liquation onset, where grain boundary surfaces are coated by a thin continuous liquid film. At even higher temperatures, the nil-strength temperature (NST) is reached, representing the point where the strength of the alloy drops to essentially zero. This is observed when grain boundaries are coated by a substantial thickness of liquid, such that the boundaries cannot accommodate any stress. At the NST, the continuous liquid

<table>
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<th>Temperature (°C)</th>
<th>Duration</th>
<th>T→Tambient</th>
</tr>
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<td>1 h</td>
<td>water quench</td>
</tr>
<tr>
<td>2. Liquid nitrogen hold</td>
<td>–196</td>
<td>30 min</td>
<td>air warm</td>
</tr>
<tr>
<td>3. Tempering</td>
<td>550</td>
<td>30 min</td>
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</tr>
<tr>
<td>4. Tempering</td>
<td>450</td>
<td>5 h</td>
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</table>
layer present at the NDT has thickened to the point where minimal capillary exists.

To determine the OC curve, typically samples are heated to a temperature somewhere between the NDT and NST, and are brought to the test temperature at a prescribed cooling rate. The amount of melting that can be tolerated by the sample during OC tests is often limited experimentally to a peak temperature between the NDT and NST; employing a peak temperature that is closer to the NST than the NDT provides a more sensitive and accurate indication of cracking susceptibility (Refs. 16, 18). Upon cooling from the peak temperature, measurable ductility of approximately 1 to 5% reduction of area (RA) is regained at the so-called ductility recovery temperature (DRT). At the DRT, it is assumed that liquid that has formed during the OH cycle has solidified extensively enough after cooling that measurable ductility is reached.

In this study, hot-ductility tests were conducted to determine the liquation cracking temperature ranges of BA-160 and established naval alloys HY-100, HSLA-100, and HSLA-65. From these data, the relative HAZ liquation cracking susceptibilities of the alloys could be determined.

**Experimental Procedure**

**Materials**

The measured chemical compositions of the experimental alloys are provided in Table 2. The BA-160 experimental material was provided by QuesTek Innovations LLC (Evanston, Ill.) in the form of 34.9-mm-diameter barstock. The other alloys were provided by the Naval Surface Warfare Center Carderock Division (West Bethesda, Md.). The HY-100 material was in the form of 0.25-in. (6.35-mm) rolled plate, and the HSLA-100 and HSLA-65
materials were in the form of 0.50-in. (12.7-mm) rolled plates.

**Hot-Ductility Tests**

Hot-ductility tests were conducted with the GLEEBLE® 3800 thermal-mechanical simulator using solid bar samples of dimensions 5 mm in diameter and 75 mm in length. The tests were performed with a jaw-to-jaw free span of 20 mm and within a high-vacuum (≈1.3 × 10⁻⁴ Pa) testing chamber. Temperature control was provided by Types K and R thermocouple wires percussion welded at the midsection of the sample. The OH test samples were heated at a rate of 100°C/s to the test temperature, holding for 1 s, and cooling at a rate of 25°C/s to the test temperature. The samples were held at the test temperature for 1 s prior to stroke initiation at a rate of 50 mm/s. For OH and OC tests, single samples were tested at each temperature. After the tests, ductility of the sample was evaluated by measuring the diameter of each fracture surface in two locations with digital vernier calipers to determine an average diameter for the %RA calculation. To determine the “equilibrium” NST (NST_{eq}) of each alloy under slower heating rate conditions, samples were heated to 1400°C at a rate of 100°C/s and then heated at a rate of 1°C/s until failure. To minimize the amount of force on the sample, the Gleeble “low-force” jaw set was employed for these tests, which exert a force of approximately 2 N on the sample.

![Image](image_url)

**Fig. 3 — Micrographs of hot-ductility samples adjacent to fracture location: A — BA-160 NDT; B — BA-160 NST; C — HY-100 NDT; D — HSLA-100 NST. Secondary microcracks were observed well away from fracture surface in BA-160 NDT samples. Nonetching segregated boundaries were observed in BA-160 and HY-100 samples suggesting wetting and penetration of liquid films.**

**Microstructure Characterization**

After testing, the samples selected for metallographic analysis were sectioned along their longitudinal direction using a precision diamond saw. Samples were then mounted, ground, polished, and etched. Light optical micrographs were acquired using an Olympus GX-51 inverted metallurgical microscope equipped with an Olympus DP71 digital camera. Fractographic analyses were conducted with a Philips ESEM FEG-30 scanning electron microscope. Samples were ultrasonically cleaned in ethyl alcohol, air dried, and then furnace baked before examination.

**Results and Discussion**

**Hot-Ductility Results**

Shown in Figs. 1 and 2 are the hot-ductility test curves for each of the alloys. For
BA-160, the OH ductility values remained above 75% RA up to 1351°C, where a maximum OH ductility of 89% RA was observed. Above 1351°C, the ductility rapidly decreased to 0% RA at the NDT of 1419°C. For the other naval alloys, similar OH ductility behavior was exhibited, with essentially maximum ductility (≈99% RA) being achieved over a wide temperature range before a sharp drop to the NDT over a very narrow temperature range. For HSLA-100, a drop in ductility from 92 to 0% RA occurred in the range from 1438° to 1446°C. Similarly, for HSLA-65, a decrease from 94 to 0% RA was experienced in the range from 1441° to 1450°C. The drop experienced by HY-100 was more gradual, with ductility decreasing from 94 to 0% RA in the range from 1400° to 1431°C. The order of the NSTs was seen to correlate with the order of the NDTs. The average NST of 1451°C for BA-160 was the lowest of the alloys tested; the highest NST of 1489°C was exhibited by HSLA-65.

The DRT of BA-160 was found to be 1303°C. The DRT of HY-100 occurred at 1411°C, which was lower compared to its NDT than for the HSLA-100 and HSLA-65, which exhibited immediate ductility recovery at 1442° and 1446°C, respectively. Similarly, the OC ductility of each alloy was lower than the OH ductility at all test temperatures.

For each alloy, the NSTeq was found to be between the NDT and NST. In each case, the NSTeq was closer to the NDT than the NST. The largest difference between the NSTeq and NDT was 13°C for HY-100. For the HSLA steels, the NSTeq was only 1°C above the NDT. A summary of the hot-ductility parameters for each alloy is provided in Table 3.

As shown in Fig. 2A–D, the ultimate force values increased with decreasing temperature both OH and OC for each of the alloys tested. Generally, the OC values

![Fig. 4 — Micrographs of NSTeq samples: A, B — BA-160; C, D — HSLA-100. In the BA-160 sample, the microstructural interface between the partially melted region and fully melted region is evident. In the HSLA-100 sample, the interface is not clear and small discontinuous cracks were observed as indicated in the micrograph.](image)

Table 3 — Summary of Hot-Ductility Test Parameters for Each Alloy

<table>
<thead>
<tr>
<th>Parameter (°C)</th>
<th>BA-160</th>
<th>HY-100</th>
<th>HSLA-100</th>
<th>HSLA-65</th>
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<td>1446</td>
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<tr>
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<tr>
<td>DRT</td>
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<td>1411</td>
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<td>1446</td>
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<td>1447</td>
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</table>
were below those measured for the OH tests. At OH test temperatures above 900°C, HY-100 consistently exhibited the lowest strength. The values were similar for the other alloys until the strength of BA-160 began to decrease below the HSLA steels above 1200°C, which could be attributed to either grain growth and/or the onset of liquation. The OC strength values of BA-160 were slightly below the other alloys at temperatures above 1100°C.

Quantitative Interpretation of Results

Several criteria can be used to interpret hot-ductility results. Among the various criteria, the temperature range between the NST and DRT, which is referred to here as the liquation cracking temperature range (LCTR), has been recommended by Lin et al. (Ref. 19) as the most appropriate for assessing HAZ liquation cracking susceptibility. The LCTR represents the maximum temperature range (NST-DRT) over which liquid films are present in the HAZ. An alloy with a narrower LCTR is deemed to have better resistance to HAZ liquation cracking because it is presumed that continuous liquid films are present in the microstructure over a narrower temperature range. The LCTR of 148°C for BA-160 indicates greater susceptibility to HAZ liquation cracking when compared with the other naval alloys tested. When compared with the results from various other alloys tested in a similar manner (Refs. 19–27), the LCTR for BA-160 is located intermediate between resistant alloys, such as 310 stainless steel (25°C), and highly susceptible alloys, such as A-286 alloy (300°C). Alloy 718 in wrought form is known to be susceptible to HAZ liquation cracking in actual practice and exhibited a LCTR of 103°C (Ref. 27). Thus, the susceptibility of BA-160 can be considered greater than 718 alloy and overall “moderate” based on a quantitative comparison to other alloys. Moderate susceptibility suggests good overall resistance to HAZ liquation cracking if weld restraint and alloy composition can be controlled carefully. Since the LCTRs of the HY and HSLA steels tested are much lower than any of the alloys exhibiting susceptibility, they are considered to have low susceptibility to HAZ liquation cracking. Of the three, HY-100 would be expected to exhibit a slightly higher susceptibility due to its slightly wider LCTR of 62°C.

Fig. 5 — Fractographs of NDT samples: A — BA-160; B — HY-100; C — HSLA-100; D — HSLA-65. Smooth intergranular fracture was exhibited by BA-160 and HY-100, which indicated continuous grain boundary liquid films. The HSLA steels exhibited primarily intergranular fracture with localized features suggesting that liquation did not occur over all the grain boundaries. Note that the fractographs are at different magnifications.

Hot-Ductility Microstructures

The microstructures of BA-160 NDT and NST samples adjacent to the fracture locations are shown in Fig. 3A–D. In the NDT sample, secondary cracks were observed at locations away from the fracture surface. The cracking occurred along prior austenite grain boundaries (PAGBs) in multiple locations and appeared to extend across adjacent grains. At higher magnification, the secondary cracks exhibited some discontinuity, suggesting discontinuous liquid films were present at temperatures slightly below the NDT. In the NST sample, nonetching grain boundary regions were observed near the fracture surface, suggesting that elemental segregation to the PAGBs had occurred during the thermal cycle. This finding also suggests that significant enrichment of the grain boundaries had occurred in the NDT to NST temperature range and that a greater quantity of liquid had formed along the boundaries prior to failure. The HY-100 NDT samples exhibited evidence of grain boundary segregation and subsequent liquid phase formation. However, secondary cracking at locations away from the fracture surface was not evident. Sim-
ilar features of secondary cracking or grain boundary enrichment were not observed in the HSLA steels. As shown in Fig. 3D, HSLA-100 exhibited a ferritic microstructure near the fracture surface. Figure 4A, B shows the microstructure of the BA-160 NSTeq sample. The center of this sample experienced complete melting, which resulted in a distinct microstructural boundary being created between the fully melted region (simulated FZ) and the partially melted region (simulated PMZ) of a weld. The microstructural interface between the simulated FZ and PMZ regions is clearly distinguished by the different etching characteristics of each region. Evident in the simulated FZ are the solidification subgrain boundaries (SSGBs) separating adjacent subgrains, which are present as cells or dendrites. These boundaries are evident due to strong compositional differences with the bulk microstructure. The SSGBs were observed to connect via an epitaxial link with the PAGBs of the PMZ, consistent with observations from autogenous weld microstructures (Ref. 11). Overall, the microstructure suggested that good wetting and penetration of liquid films occurred along the PAGBs.

To estimate the compositional enrichment occurring along the microstructural boundaries, semiquantitative energy-dispersive spectroscopy (EDS) line scans were performed across the SSGBs in the simulated FZ and PAGBs of the simulated PMZ (Ref. 28). There was found to be significant compositional enrichment of Ni and Cu at these microstructural boundaries, reaching average maximum values of 8.7 wt-% Ni and 5.6 wt-% Cu at the SSGBs and 8.8 wt-% Ni and 5.4 wt-% Cu at PAGBs.

While the microstructure of the HSLA-100 NSTeq sample did not exhibit wide grain boundary enrichment, evidence suggesting slight grain boundary enrichment and partial liquation are apparent — Fig. 4C, D. Small discontinuous cracks were observed away from the fracture surface in a location presumed to be the simulated PMZ (based on a similar distance from the fracture surface compared to the BA-160 NSTeq sample). Discontinuous darker etching regions along the grain boundaries were also noted as shown in Fig. 4D.

Hot-Ductility Fracture Surfaces

Representative NDT fracture surfaces for each alloy are presented in Fig. 5A–D. The fracture surface of the BA-160 sample exhibited clean intergranular features, supporting the assertion that failure was associated with thin liquid films present along the austenite grain boundaries. The large facets indicated extensive austenite grain growth had occurred, with grain separation evident in multiple locations as noted in the micrograph. Figure 5B exhibits a similar smooth intergranular type of fracture morphology for HY-100 at the NDT. In contrast, the HSLA-100 NDT sample exhibited intergranular fracture with elongated and step-like features. The HSLA NDT fracture surfaces exhibited a smooth intergranular morphology with localized areas decorated with dendritic features, which suggests localized melting or liquation. Higher magnification of the HSLA NDT samples revealed other areas to possess microductile rupture characteristics. This indicates that initial liquation did not occur over all of the grain boundaries in the HSLA steels. The
comparatively finer grain size of the HSLA steels compared to BA-160 and HY-100 is also noted. The NST fracture surfaces (Fig. 6A–D) exhibited similar washed-out features from significant liquation that obscured the sharper intergranular surfaces seen in the NDT samples.

The fracture surface of the BA-160 OC-1320°C sample exhibited predominantly intergranular fracture with some grain facets exhibiting ductile tearing — Fig. 7A. While the measured ductility of the sample was zero, the ductile tearing features signaled the onset of ductility recovery that occurred at a slightly lower test temperature of 1303°C. After cooling to lower temperatures, ductile failure by microvoid coalescence became the dominant fracture mode. Similar behavior was observed for HY-100 OC samples, where intergranular fracture persisted to 1411°C. Shown in Fig. 7B is the fracture surface from the HSLA-100 OC-1422°C sample. The sample exhibited predominantly ductile tearing features with isolated “pockets” of smooth intergranular or dendritic features. The pockets suggest that some discontinuous/localized liquid remained in the microstructure upon cooling from the 1465°C peak temperature. The discontinuous nature of liquid allowed for measurable ductility to be achieved by the sample.

**Microstructure Evolution at Elevated Temperatures**

Previous research related to the ductile-brittle transition in steels at near-solidus temperatures has shown this transition to be characterized by a rapid decrease in ductility over a narrow temperature range typically less than 10°C (Refs. 29, 30). The reason for the rapid decrease in ductility was attributed to incipient grain boundary melting. This mechanism involves the segregation of solute and impurity elements to grain boundaries during both initial processing and grain boundary migration upon heating to elevated temperatures near the solidus temperature ($T_{\text{sol}}$) (Ref. 29). The transition temperature was shown to be mainly dependent on C content and the level of impurity elements such as S and P (Ref. 30).

By considering the results of more than 40 compositions of low-alloy steels, Zabil’skii and Nikonova (Ref. 30) developed an empirical equation describing the ductile-to-brittle transition temperature that occurs during high temperature fracture as a function of chemical composition (wt-%):

$$T_{\text{db}} ^\circ \text{C} = 1479 – 169C – 547S – 199P – 8Mn – 6Si$$

The calculated $T_{\text{db}}$ values for each alloy are provided in Table 4. The difference between the $T_{\text{db}}$ values and the NDTs for the alloys are 48°, 11°, 13°, and –1°C for BA-160, HY-100, HSLA-100, and HSLA-65, respectively. For BA-160 and HY-100, the element contributing by far the largest decrease in $T_{\text{db}}$ is C. For the HSLA steels, the elements C and Mn have approximately the same contribution to lowering $T_{\text{db}}$. Both the multiplying factor and the magnitude of the element need to be considered together when determining the overall effect of each element. While S and P have the largest multiplying factors, their contents in the steels are also much lower than other elements, which limit

<table>
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<tr>
<th>Parameter ($^\circ$C)</th>
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<th>HSLA-100</th>
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their overall effect. Since C, S, P, Mn, and Si contents are all very low in BA-160, it reasons that incipient melting may be attributed to other alloying additions, such as Ni and Cu, which are not accounted for in the \( T_{db} \) equation of Zabilskii and Nikonova.

One of the findings in Ref. 30 was that the effect of C on \( T_{db} \) is similar to its effect on \( T_{\text{sol}} \), which is 1.8 to 2.0°C per 0.01 wt-% C according to the Fe–C binary phase diagram. By considering a similar approach, an estimate of the incipient melting temperatures was developed that includes the effect of other elements, particularly Ni and Cu. The liquidus temperature (\( T_{\text{liq}} \)) for each alloy calculated using thermodynamic data was used as the starting temperature. Examining the Fe-x binary phase diagrams provides an estimate of the effect of each element in lowering \( T_{db}^* \). Specifically, the average slope of the solidus line on the Fe-rich side of the diagram is used as a multiplying factor in a way similar to the equation developed in Ref. 30. The equation developed as a function of chemical composition (wt-%) is

\[
T_{db}^* \, (^\circ \text{C}) = T_{\text{liq}} - \left( \frac{500}{175} \right) C - 7\text{Mn} - 1236\text{S} - 175\text{P} - 7\text{Cu} - 7\text{Ni} - 1\text{Cr} - 2\text{Mo}
\]

The calculated \( T_{db}^* \) values are shown in Table 4. For C, the solidus slope can be determined from either the \( \delta \)-ferrite or austenite phases, with the respective slopes being 500 and 175°C per wt-% C. Due to the large difference in slopes, it is important to consider the high-temperature phase transformation sequence of each alloy. As will be discussed later, it is more appropriate to use the austenite solidus slope for calculating \( T_{db}^* \) of BA-160 and HY-100. When comparing \( T_{db} \) and \( T_{db}^* \) in Table 4, the consideration of Ni and Cu results in a \( T_{db} \) of 1410°C for BA-160 that is much closer to its NDT of 1419°C. The poorest agreement between \( T_{db}^* \) and NDT was exhibited by HY-100; the HSLA steels exhibited the best agreement.

Previous research has shown a partial loss of ductility occurring immediately before incipient melting in low-carbon steels is related to the austenite to \( \delta \)-ferrite phase transformation and the formation of \( \delta \)-ferrite at the austenite grain boundaries (Ref. 31). The loss in ductility associated with the austenite to \( \delta \)-ferrite transformation may be related to the so-called transformation-controlled intergranular failure mechanism (Ref. 32), whereby failure occurs due to strain concentration in the thin films of ferrite surrounding the austenite grain boundaries.

The behavior of the steels near their incipient melting temperature may be dependent on the high-temperature phase transformation sequence, particularly with regard to the peritectic reaction. The peritectic transformation in steel involves the transformation of \( \delta \)-ferrite and liquid to austenite. The on-heating, or reverse, peritectic reaction involves the transformation of austenite to \( \delta \)-ferrite and liquid — Fig. 8A. As shown, the \( \delta \)-ferrite transformation temperature (\( T_{\delta-f} \)) increases with increasing C content from a temperature of 1395°C for pure Fe. In another study, the experimental austenite to \( \delta \)-ferrite transformation temperature range was determined to be 1442° to 1452°C for several low-carbon (0.08–0.10 wt-% C) steels (Ref. 31).

Thermodynamic calculations were performed with Thermo-Calc software and TCFE5 database to determine the equilibrium phase transformation sequences for the alloys:

- BA-160 – austenitic: \( \gamma \rightarrow \gamma + L \)
- HY-100 – hyperperitectic: \( \gamma \rightarrow \gamma + \delta + L \rightarrow \delta + L \rightarrow L \)
- HSLA-100 and HSLA-65 – hypoperitectic: \( \gamma \rightarrow \gamma + \delta + L \rightarrow \delta + L \rightarrow L \)

The results predict that BA-160 is fully austenitic as the \( \delta \)-ferrite phase field and peritectic reaction are eliminated due to 6.5 wt-% Ni being a strong austenite-stabilizing factor — Fig. 8B. For HY-100, liquid is predicted to begin forming from the austenite phase prior to the \( \delta \)-ferrite transformation and is deemed hyperperitectic. For HSLA-100 and HSLA-65, the transformation sequence is hypoperitectic as austenite trans-
forms to δ-ferrite before reaching the reverse peritectic transformation.

To better understand the liquation mechanisms that are discussed in the following section, it is useful to understand the differences in the predicted weld solidification for the alloys. Solidification simulations were performed with the Thermo-Calc software and TCFE5 database. The analysis did not include impurity elements, and T_{sol} and Scheil nonequilibrium solidus (T_{sol}) were considered to occur at 0.99 fraction solid (f_{s}). The results are presented in Table 4. The primary solidification phase was predicted to be austenite for BA-160 and δ-ferrite for HY-100, HSLA-100, and HSLA-65. While BA-160 is predicted to have the narrowest equilibrium solidification range of 18°C, it is also predicted to have the widest non-equilibrium Scheil solidification range of 179°C. This indicates significant partitioning of solute elements under nonequilibrium solidification conditions.

Liquation Mechanisms

Considering that the calculated T_{sol} for each alloy is well above the measured NDTs, bulk melting of the alloys during the hot-ductility tests is considered an implausible mechanism for the observed OH ductility decrease. Microstructural features of the NDT samples did not indicate complete melting and solidification near the fracture surface. Rather, smooth intergranular fracture features were mostly evident. The bulk melting mechanism is considered plausible only for very pure metals. Instead, the OH mechanism is based on incipient melting. In the following, specific liquation mechanisms will be discussed. A summary of the contributing factors to HAZ liquation cracking susceptibility are provided in Table 5 along with schematic illustrations shown in Fig. 9.

BA-160

The overall hot-ductility behavior of BA-160 stands in contrast to the other steels tested. The OH ductility results of BA-160 suggested that continuous liquid films did not exist in the microstructure until the NDT of 1419°C. The temperature range (1351° to 1419°C) of ductility decrease during the OH tests was also much wider compared to the other steels. Since the calculated T_{sol} of 1477°C is much above the NDT of 1419°C, it is clear that a homogeneous microstructure was not achieved during the OH thermal cycle. Fractographic evidence of large well-defined facets supports that fracture at the NDT occurred by continuous grain boundary melting. It reasons that the continuous grain boundary liquid films formed at temperatures below T_{sol} were generated due to a relatively uniform solute-rich layer present along the PAGBs. The measured compositional enrichment of Ni and Cu at the PAGBs suggest their role in the observed grain boundary liquation (Ref. 28). These measurements were consistent with previous atom-probe tomography results of simulated HAZ samples that revealed the cosegregation of Ni and Cu up to maximum values of 17 and 20 at-%, respectively (Ref. 13). The results also implied that as the peak temperature of the HAZ increases, the Cu concentration at microstructural boundaries also increases. Increased concentrations of both Ni and Cu would be expected to lower the local melting point.

From thermodynamic calculations, the maximum solubility of Cu in austenite is predicted to be 13.9 wt-% for a nominal alloying addition of 6.5 wt-% Ni — Fig. 10A. The temperature of maximum Cu solubility is predicted to occur at 1377°C. Beyond the point of maximum Cu solubility in austenite for the Fe-Cu-6.5Ni system, a sharp decrease in solubility is observed up to T_{liq} — Fig. 10B. While realizing the complex multicomponent nature of the alloy and the nonequilibrium weld thermal cycle conditions, the temperature range for the sharp decrease in the maximum Cu solubility is in general agreement with the OH ductility transition temperature range for BA-160. Once a critical level of segregation had occurred that exceeded the solubility limits of the system, liquation would have proceeded along microstructural boundaries. After heating to a peak temperature above the NDT, increasing amounts of liquid would then be formed along these boundaries. The OC results indicate that the amount of solute-rich liquid that was formed after heating above the NDT did not fully solidify until cooled to temperatures approaching the DRT of 1303°C.

The observed compositional enrichment suggests a segregation-induced liquation mechanism for BA-160. In this mechanism, solute elements segregate to grain boundaries and depress the local melting point such that liquation occurs. In the absence of liquating particles, grain boundary liquation occurs from a localized variation in composition relative to the surrounding matrix. Lippold et al. (Ref. 33) suggested that grain boundary segregation in actual welds could occur from 1) equilibrium diffusion of solute elements to the grain boundaries, 2) grain boundary “sweeping” of solute elements into migrating grain boundaries during grain growth, and 3) “pipeline” diffusion of impurity and solute elements along the
Table 5 — Summary of Contributing Factors to HAZ Liquation Cracking Susceptibility for each Alloy

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Compositional</th>
<th>Microstructural</th>
</tr>
</thead>
<tbody>
<tr>
<td>BA-160</td>
<td>• high Ni and Cu content stabilizes austenitic phase at elevated temperatures</td>
<td>• incipient grain boundary melting within single austenite phase due to compositional enrichment of Ni and Cu</td>
</tr>
<tr>
<td></td>
<td>• low impurity contents in experimental lab heat</td>
<td>• good wetting/penetration of liquid films along austenite grain boundaries</td>
</tr>
<tr>
<td></td>
<td></td>
<td>• large HAZ grain size promotes continuous liquid films and increases strain accommodation requirements</td>
</tr>
<tr>
<td></td>
<td></td>
<td>• solute elements are rapidly transported across fusion boundary (“pipeline diffusion”) due to completely austenitic structure over the entire solidification temperature range</td>
</tr>
<tr>
<td>HY-100</td>
<td>• high C and Ni content produces a reverse hyper-peritectic transformation sequence at elevated temperatures</td>
<td>• incipient grain boundary melting within austenite single phase</td>
</tr>
<tr>
<td></td>
<td>• low impurity contents</td>
<td>• good wetting/penetration of liquid films along austenite grain boundaries</td>
</tr>
<tr>
<td></td>
<td></td>
<td>• large HAZ size promotes continuous liquid films and increases strain accommodation requirement</td>
</tr>
<tr>
<td></td>
<td></td>
<td>• non-epitaxy restricts pipeline diffusion mechanism</td>
</tr>
<tr>
<td></td>
<td></td>
<td>• possible formation of $\delta$-ferrite at higher temperatures would restrict wetting/penetration of liquid films</td>
</tr>
<tr>
<td>HSLA-100</td>
<td>• low C and alloy content produces a reverse hypo-peritectic transformation sequence at elevated temperatures</td>
<td>• formation of $\gamma$-ferrite in the same temperature range as incipient melting contributes to a sharp decrease in ductility over a narrow temperature range</td>
</tr>
<tr>
<td>HSLA-65</td>
<td>• micro-alloying additions form stable precipitates in HAZ</td>
<td>• incipient grain boundary melting at austenite grain boundaries during reverse peritectic transformation</td>
</tr>
<tr>
<td></td>
<td></td>
<td>• $\gamma$-ferrite formation restricts wetting/penetration of liquid films</td>
</tr>
<tr>
<td></td>
<td></td>
<td>• small HAZ grain size promotes thin and discontinuous liquid films</td>
</tr>
</tbody>
</table>

solidification grain boundaries (SGBs) of the FZ into the PAGBs of the HAZ/PMZ due to the epitaxial link established at the weld interface. Metallographic examination of the NSTeq sample (Fig. 4B) strongly suggests pipeline diffusion as an operative mechanism of grain boundary segregation. The epitaxial link observed in BA-160 is due to its completely austenitic structure in the HAZ/PMZ coinciding with primary austenitic solidification in the FZ. Grain boundary enrichment by this mechanism can be dominant because grain boundary diffusion is much faster than bulk diffusion since these interfaces have a more open structure and offer less resistance to atomic diffusion. Furthermore, as the solidification simulations showed, significant partitioning of solute elements is expected to occur in BA-160 under nonequilibrium weld solidification conditions. This is partially due to negligible backdiffusion in the solid due to face-centered cubic (fcc) solidification and the higher alloy content of BA-160. When comparing the thermodynamic phase transformation data with the Scheil solidification results, it was found that a completely austenitic HAZ is in contact with the solidifying (austenite + liquid) weld pool over the entire weld solidification temperature range of 179°C. While the pipeline diffusion mechanism cannot be conclusively proven, it is supported by optical metallographic evidence and similar grain boundary compositions measured in the simulated FZ and PMZ (Ref. 28). This is due to the fact that rapid grain boundary diffusion ensures that the boundary concentration equilibrates.

Segregation-induced liquation by itself is not enough to produce a crack-susceptible microstructure. It is essential that the liquid species penetrate and wet the boundary to create a continuous liquid film that limits solid-solid contact along the grain boundaries. Metallographic evidence suggesting good wetting and penetration of liquid films along the PAGBs was apparent in the simulated PMZ. These observations can be attributed partially to a completely austenitic (fcc) microstructure at elevated temperatures. The enhanced wettability of austenite/austenite grain boundaries compared to austenite/ferrite grain boundaries was established by Matthews and Savage (Ref. 34). Their research indicated good wetting and penetration of liquid Cu-Ni weld metal along austenite grain boundaries in the HAZ of HY-80 naval steel. The liquid did not tend to penetrate alloys possessing a body-centered cubic (bcc) structure at elevated temperatures, even under the application of a considerable amount of strain.

The average prior austenite grain diameters measured for a peak temperature of 1300°C were 76, 70, 54, and 24 $\mu$m for BA-160, HY-100, HSLA-100, and HSLA-65, respectively. The relatively large grain size of BA-160 would exacerbate the factors mentioned previously. The linear relationship between grain size and cracking susceptibility has been demonstrated (Ref. 35) and is based on a strain accommodation and liquid distribution argument. As the total grain boundary area decreases (larger grain size), it reasons that the fraction of the total strain accommodated by a single boundary increases. Furthermore, as grain size increases for a fixed volume-percent of liquid, the boundary liquid film becomes thicker and/or more continuous. Control of austenite grain size in BA-160 should be considered for improved HAZ liquation cracking resistance.

As detailed in the preceding, HAZ liquation cracking susceptibility of BA-160...
can be attributed to the following: 1) a completely austenitic microstructure at elevated temperatures, which would promote pipeline diffusion of solute elements in an actual weld; 2) a segregation-induced liquation mechanism involving the compositional enrichment of Ni and Cu along austenite grain boundaries; 3) observed good wetting and penetration of liquid films along like austenite/austenite grain boundaries; and 4) a large HAZ grain size, which promotes continuous grain boundary liquid films and increases the strain accommodation requirement for a given grain boundary.

HY-100

The HY-100 steel exhibited a much narrower LCTR compared to BA-160, but a slightly wider LCTR compared to the HSLA steels. This greater susceptibility can be partially attributed to the onset of liquation within the austenite single phase. The higher C and Ni contents compared to the HSLA steels allows for a higher concentration of these elements for an equivalent level of segregation. Similar to BA-160, metallographic features of wider grain boundary enrichment were observed, which is attributed to the better wettability of austenite/austenite boundaries.

The higher C content (0.18 wt-% C) of HY-100 is well outside the C range of steels that were used to experimentally determine the δ-ferrite transformation temperature noted previously in Ref. 31. However, increasingly higher C contents increase the δ-ferrite transformation temperature up to the peritectic transformation temperature in the hyperperitectic range. Thermodynamic calculations predict liquid to begin forming from the austenite single phase either due to incipient melting or through the transformation of austenite to liquid prior to the δ-ferrite transformation at 1495°C.

The transformation of austenite to δ-ferrite at higher temperatures would act to limit the wetting and penetration of liquid films in HY-100. In the presence of a steep temperature gradient between the weld pool and HAZ, it is possible that δ-ferrite does not form. If the transformation does not occur and the matrix remains completely austenitic in the HAZ/PMZ regions near the weld interface, epitaxial growth of the bcc δ-ferrite weld metal would be inhibited due to its different crystal structure (Ref. 36). As such, the HAZ grain boundaries would be noncontiguous with the SGBs in the FZ and pipeline diffusion would be greatly restricted. Thermodynamic calculations predict that an austenitic HAZ is in contact with the solidifying weld pool at $f_s > 0.49$ over a temperature range of 68°C, which are much narrower ranges than those calculated for BA-160.

HSLA-100 and HSLA-65

The calculated $T_{eq}$ of the HSLA steels were in the range of 29°C to 42°C above their NDTs. This is in good agreement with the experimental findings of Weinberg (Ref. 29) for steels containing 0.05 to 0.12 wt-% C, where brittle failure due to incipient grain boundary melting was noted to occur in a range beginning approximately 40°C below the equilibrium $T_{eq}$. The sharp drop in the OH ductility of the HSLA steels was also very similar to the behavior of plain carbon steels reported in that study.

The NDT values for both HSLA steels (Table 3) were within the experimental austenite to δ-ferrite transformation temperature range noted previously from Ref. 31 (1442°C to 1452°C). The δ-ferrite transformation temperature was calculated to be 1484°C and 1461°C for HSLA-100 and HSLA-65, respectively. The better agreement for HSLA-65 between the experimental and calculated values is not surprising given that its composition is a better match with the steels in that experimental study. The higher Ni content of HSLA-100 compared to HSLA-65 is expected to stabilize the austenite phase to a higher temperature, as predicted by thermodynamic calculations.

Since the calculated $T_{eq}$ and $T_{eq}$ values are near the experimental austenite to δ-ferrite transformation temperatures, it is possible that the formation of δ-ferrite occurs within the same temperature range as incipient melting. The formation of δ-ferrite actually promotes incipient melting at the austenite grain boundaries. When a particular steel is in the two-phase austenite + δ-ferrite region, partitioning of alloying elements occurs. During growth of δ-ferrite, austenite stabilizing elements such as C, Mn, and Ni segregate to the austenite phase. Since C is a very fast interstitial diffuser, it easily becomes enriched in the austenite phase. Due to the reverse peritectic transformation, incipient melting occurs at the austenite/austenite grain boundaries, but not at the austenite/δ-ferrite or δ-ferrite/δ-ferrite boundaries. The strong effect of C on the incipient melting temperature is borne out in both of the $T_{eq}^\prime$ and $T_{eq}^\prime$ equations. Additionally, since bcc Fe has a higher solid solubility for impurity elements, such as S (Ref. 37) and P (Ref. 38), than fcc-Fe, it can be reasoned that initial liquation due to impurity element segregation is not promoted at the δ-ferrite grain boundaries. It can be concluded that the volume of liquid that forms along austenite/austenite boundaries is probably far greater than at austenite/δ-ferrite and δ-ferrite/δ-ferrite boundaries, which has been supported elsewhere through microstructural examination of stainless steels (Ref. 33).

With reference to diffusion data calculated at 1400°C (Ref. 39), the diffusion coefficients of C in austenite ($≈ 3.5 \times 10^{-4}$ cm²/s) and δ-ferrite ($≈ 1.0 \times 10^{-4}$ cm²/s) are orders of magnitude greater than for Ni or Cu in austenite ($≈ 5 \times 10^{-16}$ cm²/s). Since it is proposed that the liquation mechanism in HSLA steels is dominated by the fast interstitial diffusion of C during the reverse peritectic reaction, it is expected that sample failure in the slower heating rate NSTeq tests would occur near the NDT. This is what was observed for the HSLA steels, where the NSTeq values were only 1°C above the NDT (Table 3). However, where the liquation mechanism is associated with slower diffusing elements such as Ni and Cu, the NSTeq was shown to be further above the NDT as was the case for BA-160 and HY-100.

This nucleation and growth of δ-ferrite would also promote a sharp OH ductility decrease from a mechanical standpoint. Since the self-diffusion coefficient of bcc $\Delta Fe$ ($≈ 6 \times 10^{-4}$ cm²/s) is at least two orders of magnitude greater than that of fcc γ-Fe ($≈ 6 \times 10^{-16}$ cm²/s) (Ref. 39), the rate of recovery is also higher in bcc δ-Fe. This mechanism is valid in low-carbon steels in the vicinity of the incipient melting temperature, whereby a rapid drop in the flow stress is attributed to the austenite to δ-ferrite transformation (Ref. 31). Also, the formation of δ-ferrite has the potential to cause grain boundary decohesion and cracking, and reduce ductility similar to that observed in stainless steels, where δ-ferrite is avoided during thermo-mechanical processing.

The OH loss in ductility of the HSLA steels can be attributed to two factors: 1) the formation of δ-ferrite at austenite grain boundaries and 2) incipient melting at the austenite/austenite grain boundaries. However, the formation of δ-ferrite within the same temperature range as incipient melting is considered beneficial to cracking susceptibility. The formation of δ-ferrite restricts wetting and penetration of liquid films since two-phase boundaries do not wet nearly as effectively as like boundaries promoting liquid discontinuity. This beneficial effect of δ-ferrite formation was observed by Kujanpaa et al. (Ref. 40) in austenitic stainless steels, where a small amount of ferrite formed along austenite grain boundaries in the HAZ was effective in reducing susceptibility to cracking relative to completely austenitic alloys. Any evidence of δ-ferrite was most likely masked by subsequent re-transformation to austenite upon cooling. The fracture surfaces of the HSLA-100 (Fig. 5C) and HSLA-65 (Fig. 5D)
NDT samples were comparatively more decorated than for BA-160 and HY-100, suggesting failure did not occur solely due to continuous grain boundary liquid films in these alloys.

Due to their lower overall alloy contents, grain boundary liquation in the HSLA steels is expected to be limited when compared to the more highly alloyed BA-160 and HY-100 steels. The finer HAZ grain size of the HSLA steels is also beneficial since any liquid that does form has much less capability of wetting and forming continuous liquid films compared to the larger HAZ grain size of BA-160 and HY-100. As such, the liquid films that do form along the grain boundaries in the HAZ are predicted to be extremely thin and/or discontinuous. This was best exemplified by the isolated pockets of smooth intergranular and dendritic fracture observed on the fracture surfaces of HSLA-100 OC samples — Fig. 7D.

Limitation of Hot-Ductility Approach

It should be cautioned that the hot-ductility approach for determining HAZ liquation cracking susceptibility provides information for only one location in the HAZ. Since the DRT is determined using a single peak temperature, it approximates only a single location in the HAZ/PMZ that is close to the weld interface. Although the NDT is viewed as being a constant for a given heating rate, the DRT is known to be a function of the OC peak temperature (Ref. 18). Selection of a peak temperature close to the NST is thus imperative for obtaining representative values of the LCTR that relate to locations in the PMZ where HAZ liquation cracking occurs. Upon further scale-up of the BA-160 alloy, spot Varestraint testing of plate/plate material is recommended to provide an additional assessment of HAZ liquation cracking susceptibility. The technique developed by Lin et al. (Ref. 19) could be employed to accurately determine the thermal crack-susceptible region surrounding the weld within which HAZ liquation cracking is possible. The determination of a precise temperature range for cracking through spot Varestraint testing would provide more information toward verifying the current assessment of “moderate” HAZ liquation cracking susceptibility.

Conclusions

In this study, hot-ductility testing was employed to determine the LCTR and evaluate the HAZ liquation cracking susceptibility of BA-160, an alloy recently developed for naval applications, relative to established naval alloys HY-100, HSLA-100, and HSLA-65.

1. A LCTR of 148°C was determined for BA-160, which was the widest of the alloys tested and indicates the greatest susceptibility to HAZ liquation cracking. In comparison to published results from a wide range of alloys tested in a similar manner, BA-160 is considered to have moderate susceptibility. The other alloys exhibited narrower LCTRs (HY-100, 62°C; HSLA-100, 40°C; HSLA-65, 43°C) that suggest low cracking susceptibilities.

2. The HAZ liquation cracking susceptibility of BA-160 can be attributed to the following: a) a completely austenitic microstructure at elevated temperatures, which would promote pipeline diffusion of solute elements in an actual weld; b) a segregation-induced liquation mechanism involving the compositional enrichment of Ni and Cu along austenite grain boundaries; c) observed good wetting and penetration of liquid films along like austenite/austenite grain boundaries; and d) a large HAZ grain size, which promotes continuous grain boundary liquid films and increases the strain accommodation requirement for a given grain boundary.

3. The slightly wider LCTR of HY-100 compared to the HSLA steels was attributed to the onset of liquation within the austenite single phase. Microstructure characterization revealed wider grain boundary enrichment, and enhanced wetting and penetration of liquid films compared to the HSLA steels.

4. The HSLA-100 and HSLA-65 steels exhibited similar overall hot-ductility behavior. Their sharp decrease in OH duc- tility over a narrow temperature range of approximately 10°C was attributed to the formation of δ-ferrite within the same temperature range as incipient grain boundary melting. The OH liquation mechanism was related to the reverse peritectic transformation. Their narrow LCTRs were attributed to restricted wetting and penetration of liquid films due to the presence of δ-ferrite and a relatively small HAZ grain size. Ineffective wetting and liquid discontinuity was established through metallographic and fractographic observations.

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