



Residual Copper Effects on Toughness of GMAW Low-Carbon Steel

Higher copper content increases hardenability and reduces the toughness of the base metal, but the GMAW HAZ toughness is negligibly affected

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Abstract

Increasing amounts of residual copper in the steel scrap stream can lead to enrichment in recycled steel products. While copper is often controlled to avoid issues during thermomechanical processing, its downstream effects on microstructure and properties after welding are less known. This work explores the effect of increased copper content, up to 0.83 wt-%, on microstructure and Charpy impact behavior of hot-rolled low-carbon steel plates welded with gas metal arc welding (GMAW). Microstructural characterization of the hot-rolled base metal and GMAW heat-affected zone (HAZ) was conducted. It was found that austenite stabilization and an increase in hardenability from copper led to a higher fraction of more brittle non-ferritic microstructural constituents in the base metal, resulting in a higher 20 ft-lb ductile-brittle transition temperature (DBTT). However, the GMAW HAZ of all the alloys primarily consisted of tempered martensite, resulting in an overall lower DBTT than that of the base metals, with little effect from copper content. This work shows that copper's effects on microstructural evolution must be considered during thermomechanical processing and welding to ensure high toughness in low-carbon steel plate.

Keywords

- Arc Welding
- Steel
- Residual Copper
- Scrap

Introduction

Scrap-based steelmaking in the electric arc furnace (EAF) has steadily gained appeal from both an economic and emissions standpoint, with recovery rates of up to 90% (Ref. 1). As demand for common steel products is expected to grow, steel recycling in the EAF will become increasingly feasible compared to primary steelmaking methods (Ref. 2). However, applications of recycled steel are constrained by contamination with other metals, most notably copper (Ref. 3). End-of-life scrap is rich in copper because it contains motors and wiring from common appliances, automobiles, and other sources of recyclable material. Due to its chemical nobility, copper is not easily removed once it enters the steel melt. The most common mitigation strategy is pig iron dilution, which is used to achieve copper concentrations low enough to avoid problems during thermomechanical processing. Namely, the issue of “hot-shortness,” a form of liquid-metal embrittlement, can occur with sufficiently high copper content during hot deformation processes (Ref. 4). While copper concentration thresholds to avoid such behavior are somewhat established, less is known about the effects of copper during downstream processes, such as welding.

Copper is known to act as an austenite stabilizer, lowering the Ae1 temperature and promoting higher amounts of austenite at intercritical temperatures. Hsu and Lin found that increasing copper increases the amount of retained austenite (RA) in the final microstructure of ductile irons (Ref. 5). Copper is also known to influence the transformation behavior of austenite by increasing hardenability

Table 1 — Compositions and Carbon Equivalent for All Base Materials (wt-%)

Alloy	C	S	P	Mn	Mo	Si	Cr	Cu	CE
0 Cu	0.097	0.004	0.003	1.81	0.048	0.17	0.46	0.02	0.49
28 Cu	0.102	0.004	0.004	1.99	0.037	0.24	0.5	0.28	0.55
83 Cu	0.099	0.004	0.005	1.97	0.035	0.25	0.48	0.83	0.58
83 Cu_RMn	0.099	0.004	0.005	1.76	0.042	0.25	0.49	0.83	0.55

(Ref. 6). Hardenability is a measure of a steel's propensity to form martensite upon cooling. Higher hardenability entails a larger range of cooling rates at which austenite is capable of forming a fully martensitic microstructure, rather than other microconstituents. Many metrics of hardenability have been established. In heat-treatment settings, the "critical diameter" refers to the maximum diameter of a round bar that can be quenched to yield at least 50%, and sometimes 100%, martensite throughout its center (Ref. 7). In welding, "carbon equivalent" (CE) is most widely used, which is a linear combination of alloying elements based on their hardening potency relative to carbon (Ref. 7). A higher CE indicates a greater likelihood of martensite formation in materials undergoing rapid cooling (e.g., a heat-affected zone). In either case, these metrics describe the critical cooling rate at which austenite begins to form a phase other than martensite. With other hardening alloy elements held fixed, CE can be a useful metric that captures the hardenability contribution of copper. Furthermore, it can serve as a guide for determining how much to reduce other alloy elements to compensate for the increased copper content.

The goal of this work is to examine the effect of increased copper content on the heat-affected zone characteristics of gas metal arc-welded low-carbon plate steel. V-groove weld joints were produced from multiple plate compositions and sectioned for metallographic analysis, hardness measurements, and Charpy impact testing. The microstructures of both the as-rolled and welded steels were characterized using light optical microscopy (LOM) and scanning electron microscopy (SEM). Thermodynamic simulations were performed to predict austenite volume fractions and chemical makeup. Hardness measurements and Charpy impact tests were conducted on both as-rolled and welded coupons to measure shifts in hardenability and the ductile-brittle transition temperature (DBTT), respectively. Copper contents up to 0.83 wt-% were explored, and the reduction of manganese content, guided by carbon equivalent, was evaluated to compensate for the increased hardenability granted by copper. The results of this work are expected to help quantify copper's effect on both base metal and welded microstructures, as well as shifts in fracture behavior, which may arise in the welding of scrap-based steel products with residual copper.

Experimental Procedures

Base Metal Design and Production

A set of four alloys representing a low-carbon plate steel for construction, transportation, or pipeline applications was produced for this study. The alloys were melted and cast by Cleveland-Cliffs Inc. Ingots were reheated to 1230°C and hot-rolled to a final thickness of 0.625 in. (16 mm) at a minimum finish temperature of 850°C. After hot-rolling, the plates were placed in a preheated furnace at 650°C and furnace-cooled to below 150°C. Three of the compositions contained increasing amounts of copper, some well past amounts seen in production, while the fourth composition maintained a high copper content but with reduced manganese to achieve a reduced CE (Ref. 8). This alloy, 83Cu_RMn, was designed to have the same CE as the 28Cu alloy. Carbon equivalents were calculated for each of the compositions using Eq. 1 (Ref. 9). All measured compositions and CE values are listed in Table 1.

$$CE = C + \frac{Mn}{6} + \frac{Cu + Ni}{15} + \frac{Cr + Mo + V}{5} \quad (1)$$

Thermodynamic Simulations

Single-axis equilibrium simulations of austenite volume fraction and composition at the eutectoid temperature were performed in Thermo-Calc 2024a software using the TCFE13 database to estimate how increasing copper content might affect microstructure development in the base metal and heat-affected zone (HAZ). The eutectoid temperature was taken to be the lowest temperature in the austenite + ferrite two-phase field, below which austenite, ferrite, and cementite are in equilibrium.

GMAW Parameters

Sections of plate measuring approximately 3 x 6 in. were machined to produce a 25° bevel to form a V-groove geometry along the rolling direction. Oxide scale was removed from both the top and bottom surfaces of the plate near the V-groove using an 80-grit grinding disc. Plates were clamped down with a root opening of 5 mm prior to gas metal arc

Table 2 — Calculated Austenite Volume Fraction, Composition (in wt-%), and Carbon Equivalent for Each of the Alloys at Calculated Eutectoid Temperature

Alloy	Eutectoid (°C)	Vol. Frac. γ	C	Cu	Mn	Cr	Mo	Si	CE
0 Cu	697.5	0.192	0.486	0.042	4.331	0.700	0.026	0.152	1.355
28 Cu	691.4	0.213	0.442	0.590	4.700	0.730	0.018	0.225	1.414
83 Cu	682.4	0.231	0.399	1.780	4.757	0.664	0.017	0.244	1.447
83 Cu_RMn	688.7	0.228	0.414	1.714	4.238	0.689	0.024	0.236	1.377

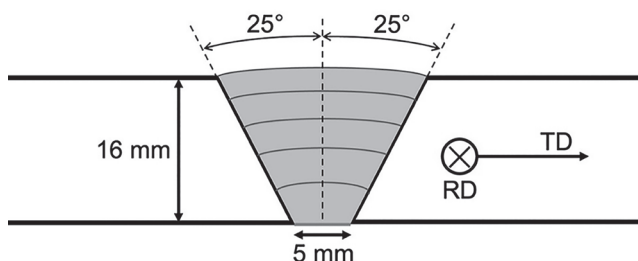


Fig. 1 — Schematic of GMAW V-groove weld geometry.

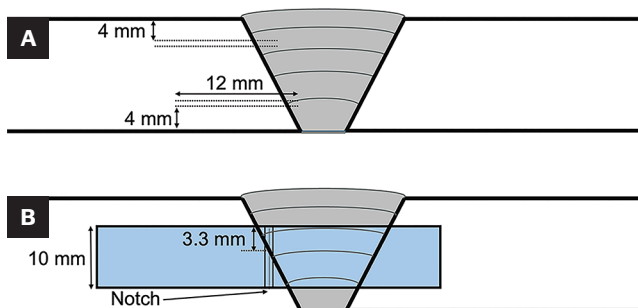


Fig. 2 — A — Vickers hardness traverse patterns; B — location and orientation of ASTM E23 machined Charpy specimens.

welding (GMAW) with a collaborative robot-controlled Fronius TPS400i heat source. All welds were performed along the rolling direction (RD) of each plate, as shown in Fig. 1.

An identical welding program was performed for each set of plates, which consisted of a cold metal transfer (CMT) root pass, followed by four pulse multi-control (PMC) fill and cap passes, all performed with 1.1 mm (0.045 in) diameter ER70S-6 filler using 90/10 argon-CO₂ shielding gas at a flow rate of 15 lpm (30 CFH). An interpass temperature of 100°C was targeted for all passes. For the CMT root pass, the wire was fed at 315 inches per minute (ipm) with a current of 255 A at 16.5 V. For subsequent PMC passes, wire was fed at 330 ipm with a current of 220 A at 23.3 V. A travel speed of 4 mm/s was used for the first three weld passes, then slowed to 3.5 mm/s for the last two passes. This, along with gradually increased weave widths, compensated for the increased

groove width at later weld passes. Upon cooling, the welds were sectioned for metallography, hardness testing, and machining of Charpy specimens.

Vickers Hardness Measurements

Vickers hardness profiles were measured across the HAZ in weld cross-sections for each heat, using a load of 500 g. Four traverses were performed on each sample, with one pair located 4 mm from the top surface (cap) of the welded plate, and the other pair located 4 mm from the bottom surface (root). Indents were spaced 250 μ m apart and spanned a length of 12 mm, originating from the fusion zone. A schematic is shown in Fig. 2A. Hardness values from each pair of rows were averaged to give an overall hardness profile for both the root and cap sections of the weld microstructure, along with 95% confidence intervals calculated with a pooled standard deviation. Compared with the base metal, a 5 x 5 grid of indents was made in the unwelded material with identical load/spacing.

Charpy Impact Testing

ASTM E23 Charpy specimens were machined from both unwelded base metal and welded plates of each composition. Charpy samples were oriented along the transverse direction of the plate, with the notch extending along the normal direction (ND) and the crack plane of the specimen along the rolling direction. To capture the full HAZ, Charpy samples were oriented such that the notch consisted of approximately $\frac{1}{3}$ fusion zone material and $\frac{2}{3}$ HAZ material, as shown in Fig. 2B. This orientation results in the notch extending from the CGHAZ to the ICHAZ. Impact tests were carried out on an MTS Exceed E22 pendulum impact tester. For elevated temperature tests, samples were soaked in temperature-controlled water for 25 minutes prior to impact testing. A liquid nitrogen chamber onboard the impact tester was used for low-temperature tests. A threshold of 20-ft·lb absorbed energy was used to compare transition temperature ranges between the conditions. Due to limited material, replicate testing could only be performed at temperatures near the 20 ft·lb transition temperature of the welded samples.

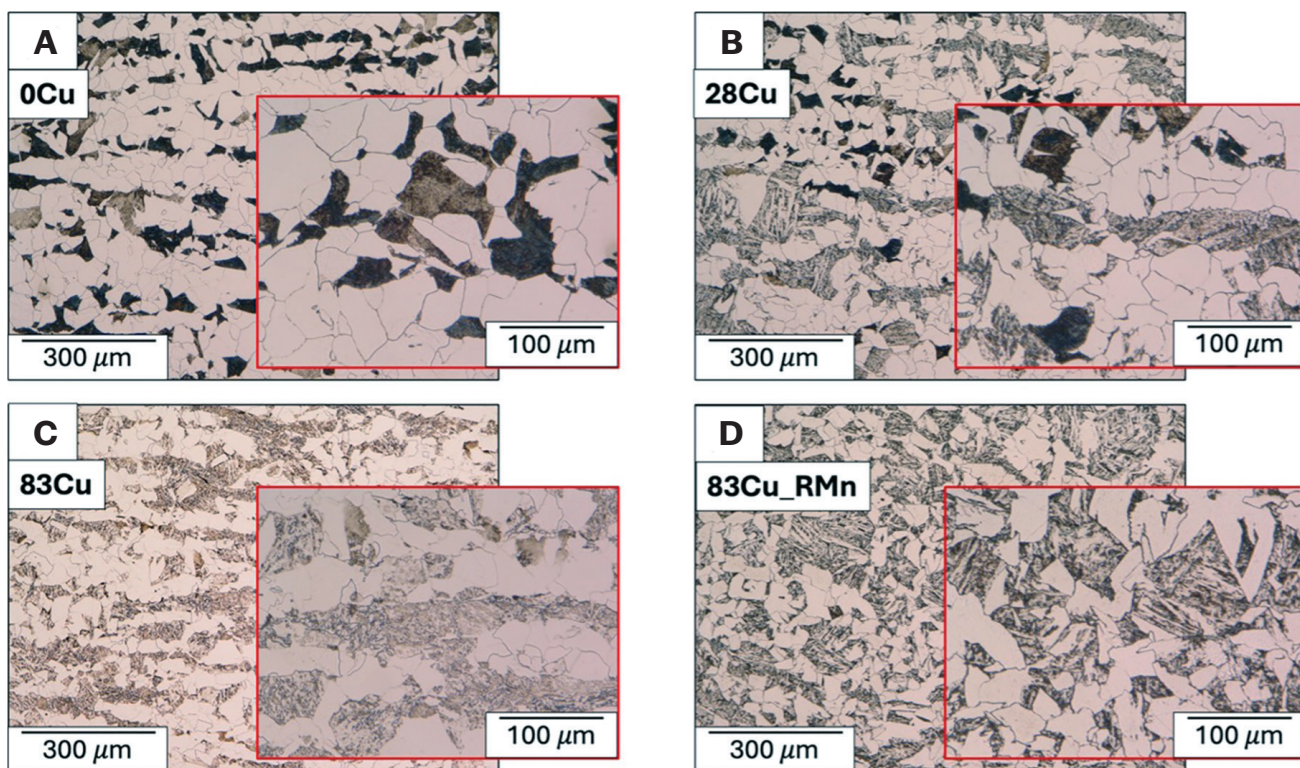


Fig. 3 — Light optical micrographs of base metal microstructures: A — 0Cu; B — 28Cu; C — 83Cu; D — 83Cu_RMn. Rolling direction is left to right.

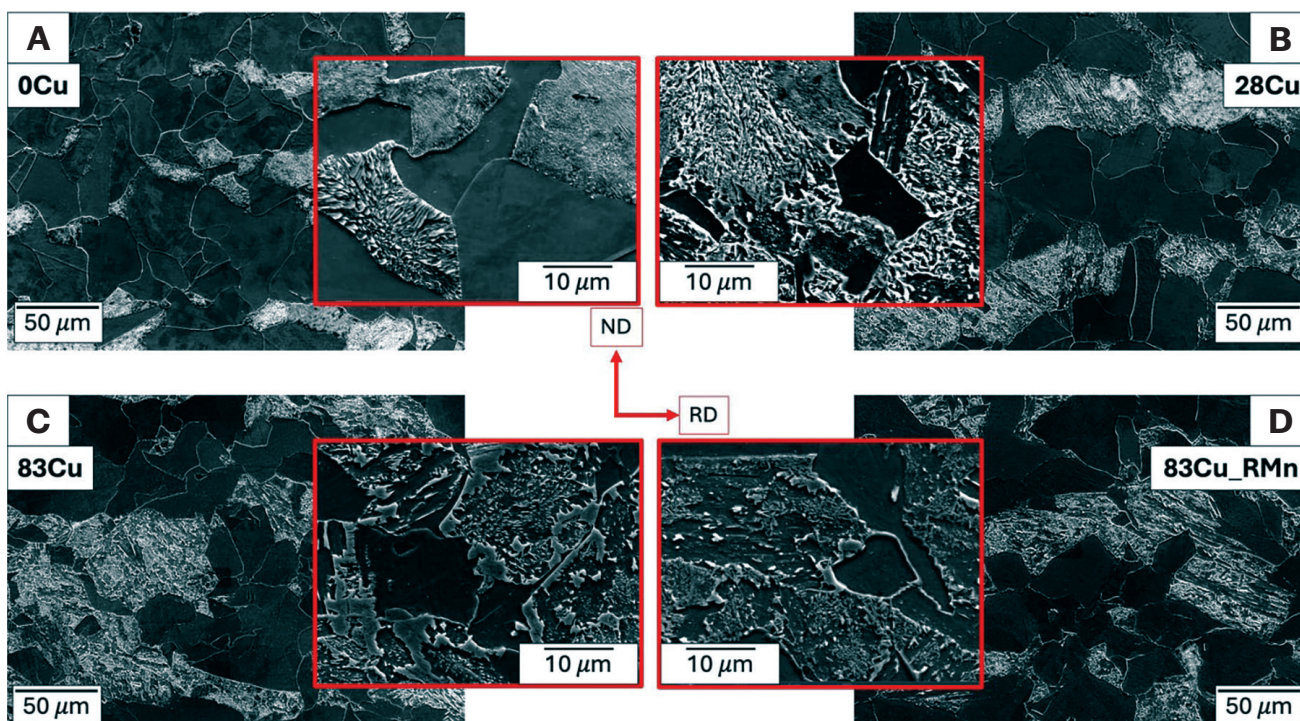


Fig. 4 — SEM SE micrographs of base metal microstructures: A — 0Cu; B — 28Cu; C — 83Cu; D — 83Cu_RMn. Rolling direction is left to right.

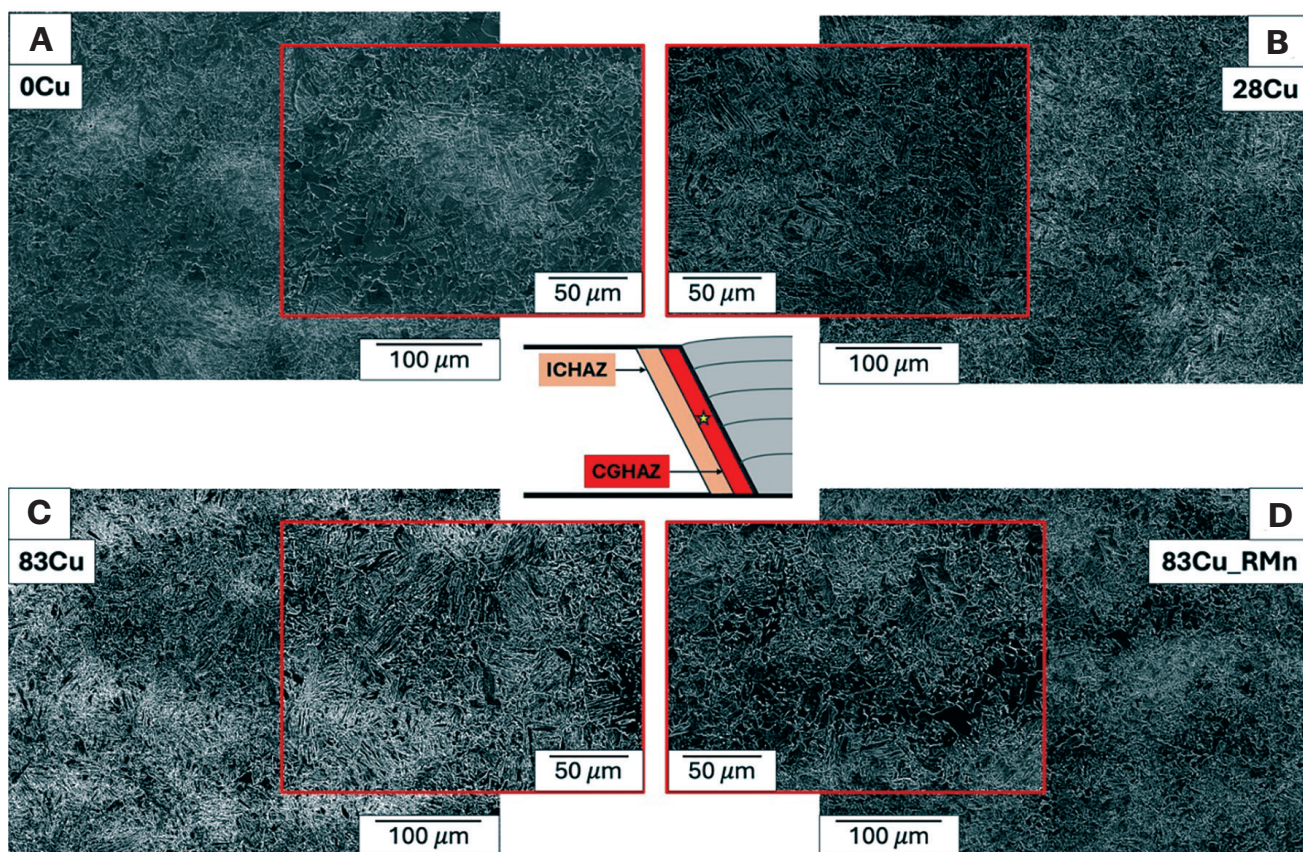


Fig. 5 — SEM SE micrographs of the CGHAZ: A — 0Cu; B — 28Cu; C — 83Cu; D — 83Cu_RMn.

Microscopy

Sections of the HAZ were metallographically polished and etched with 1% nitric acid in ethanol prior to imaging with a Tescan S8252G scanning electron microscope (SEM). A 20 keV accelerating voltage and 10 nA beam current were used, with images captured using a secondary electron (SE) detector. The same SEM settings were used for the fractography of failed Charpy specimens. Light optical images were taken on an Olympus DSX500 digital optical microscope.

Results

Thermodynamic Simulations

Eutectoid temperatures and corresponding austenite phase fractions and compositions for each of the alloys are shown in Table 2. With increasing copper content, the eutectoid temperature decreases and the amount of austenite stable at the eutectoid temperature increases, as would be expected from an austenite stabilizer. Furthermore, with increasing copper content, the hardenability of austenite at the eutectoid temperature increases, as evidenced by higher CE values. The reduction in Mn in the 83Cu_RMn alloy results in an increase in eutectoid temperature, reduction in austenite, and reduction in the CE of the austenite, but not to the degree that they are equivalent to the 28Cu alloy that has the same CE of the bulk composition. Based on the results in

Table 2, it would be expected that harder microconstituents like bainite and/or martensite would more likely form in the higher copper content alloys at a given cooling rate from high temperatures.

Base Metal Microstructure

Light optical micrographs of each base metal were collected and are shown in Fig. 3, with the rolling direction oriented left-right. The microstructure consists largely of ferrite (lighter etching features), with the amount of secondary constituent (darker etching features in horizontal bands) increasing as copper is added. Further, the secondary constituent transitions from what appears to be pearlite in the 0Cu composition to a combination of martensite/bainite/austenite (MBA) in the higher copper samples. SEM SE micrographs of the same samples were imaged (in the same orientation) to further analyze the morphology of the non-ferritic constituents, and the results are shown in Fig. 4. Figure 4A shows that the non-ferritic constituent is pearlite in the 0Cu composition, whereas the higher copper compositions show acicular features consistent with bainite or auto-tempered martensite. The 83Cu and 83Cu_RMn alloys also show non-etching regions that appear similar to martensite and retained austenite. It appears that the higher hardenability (higher CE) and higher stability of the austenite with increasing copper content manifests itself in increased non-ferritic microconstituents that have more MBA character rather than pearlite.

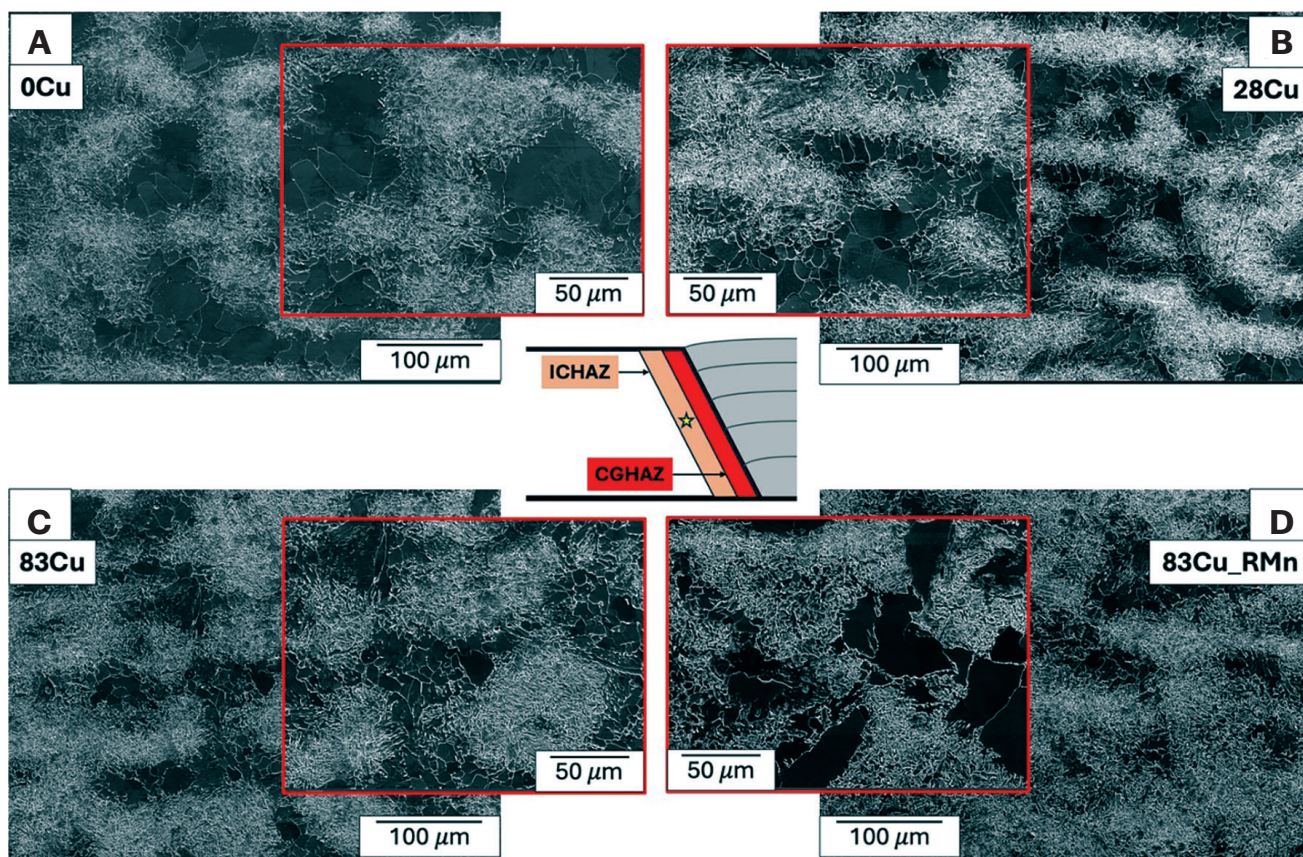


Fig. 6 — SEM SE micrographs of the ICHAZ: A — 0Cu; B — 28 Cu; C — 83Cu; D — 83Cu_RMn.

Weld Microstructures and Hardness

SEM SE images of the CGHAZ, the HAZ region nearest the fusion zone, are shown in Fig. 5 for all compositions. All of the alloys display acicular features consistent with tempered martensite, likely a result of the higher cooling rates (relative to the base metal processing) and subsequent reheating in this region of the HAZ. Similar micrographs were collected for the region of the HAZ furthest from the fusion zone that has a noticeable change in microstructure relative to the base metal, the ICHAZ, and SEM SE micrographs shown in Fig. 6. Similar to the base metal, there is a greater amount of non-ferritic constituent with increasing copper content. However, the non-ferritic constituent appears to be tempered martensite for all the alloys, which is likely due to the higher cooling rates experienced in the ICHAZ compared to base metal processing. In general, there are much smaller differences in the microstructure of the HAZ as a function of copper content compared to the base metal counterparts.

Hardness

Average hardness profiles of the HAZ, both near the cap and root, are shown in Figs. 7A and B, respectively. The hardness of each base metal is shown as a “band,” centered about its average hardness with a width equal to its 95% confidence interval. The HAZ may be visualized by the elevated hardness region between the fusion zone (right edge) and the intersec-

tion with a given base metal hardness band. In general, the highest hardness portion of the heat-affected zone increases with increasing copper content. The 83Cu_RMn alloy shows a reduction in peak hardness relative to the 83Cu alloy, but not to the extent of the 28Cu alloy, which has the same bulk CE. The width of the HAZ also appears to be wider with the higher copper alloys, likely due to increased hardenability and/or austenite stabilization.

Charpy Impact Testing

Charpy impact energy versus temperature for base metal and GMAW HAZ samples of each alloy are plotted in Fig. 8. A horizontal line noting 20 ft-lb is also shown. A rather drastic increase in the transition temperature is observed in base metal samples with increasing copper content. The 0Cu base metal sample with a ferrite + pearlite microstructure has a considerably lower transition temperature compared to 28Cu, 83Cu, and 83Cu_RMn, which all have ferrite + MBA microstructures. The 83Cu_RMn alloy shows a slightly lower transition temperature than the 83Cu alloy, which is similar to the 28Cu alloy with the same bulk CE.

Individual test values are indicated with symbols, and the curve fit goes through the average value when replicate tests are present. “BM” indicates base metal, and “WELD” indicates the GMAW HAZ.

Unlike the base metal, the GMAW HAZ samples show a much smaller effect of copper content on the transition tempera-

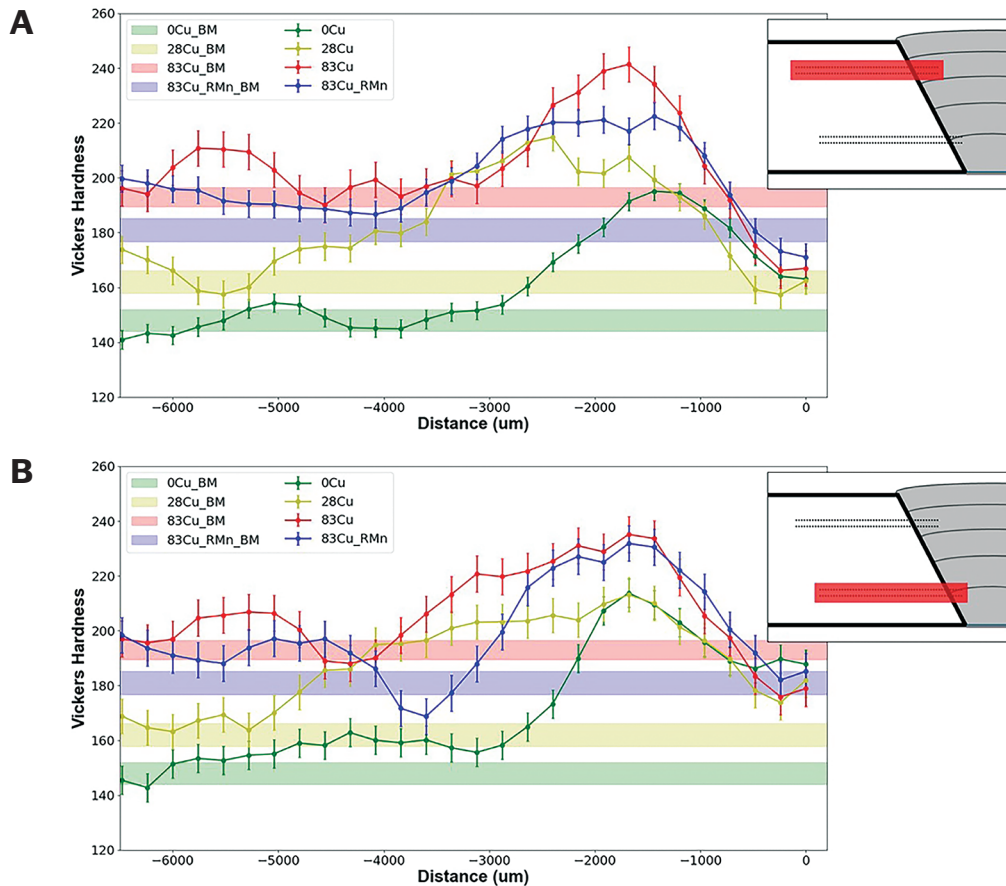


Fig. 7 — Average hardness profile: A — Upper “cap” region of the weld; B — lower “root” region of the weld. The error bars indicate the 95% confidence intervals of the mean hardness in each traverse. The solid color bands indicate the 95% confidence intervals of the mean base metal hardness values.

ture, likely due to their more similar microstructures in the HAZ. Overall, the transition temperatures of all the GMAW HAZ samples are lower than those of their base metal counterparts, likely due to their tempered martensite microstructures compared to the untempered MBA constituents in the base metal. There does not seem to be a systematic trend in transition temperature with copper content in the GMAW HAZ samples, other than the 83Cu alloy, which shows a slightly higher transition temperature compared to lower Cu alloys and the reduced manganese alloy.

Fractography

Tests performed at 20°C exhibited the largest range in energy absorption values between the various alloys, particularly in the base metal. The fracture surfaces from all the samples tested at 20°C were evaluated to determine fracture mechanisms associated with the different absorbed energy values. The fracture surfaces of base metal samples from the 0Cu and 83Cu compositions are shown in Fig. 9. The higher energy absorption in the 0Cu alloy exhibits ductile microvoid coalescence. However, the lower energy absorption (below 20 ft·lb) for 83Cu shows brittle cleavage fracture similar to the other base metal samples tested at 20°C. The more brittle untempered MBA constituents in the higher copper alloys

are likely responsible for the brittle impact behavior. Samples from the GMAW HAZ all showed much higher impact energies than the base metal counterparts, and the fracture surfaces from the 0Cu and 83Cu alloys tested at 20°C are shown in Fig. 10. Both alloys show ductile microvoid coalescence, albeit at a much finer scale, which is likely related to the tougher tempered martensite microstructure in the HAZ of both alloys. All GMAW HAZ samples tested at 20°C showed similar microvoids, as shown in Fig. 10.

Discussion

Influence of Residual Copper on Microstructure Evolution

Figures 3 and 4 show significant differences in the amount and morphology of the non-ferritic microstructural constituents in the base metal with increasing copper content. The relatively slow cooling from hot rolling indicates that the equilibrium predictions of austenite amount and composition shown in Table 2 are reasonable approximations to the microstructure evolution in the base metal. With increasing copper, the amount of austenite increases while the hardenability of the austenite increases at the eutectoid temperature. These effects are observed in the base metal microstructures, where

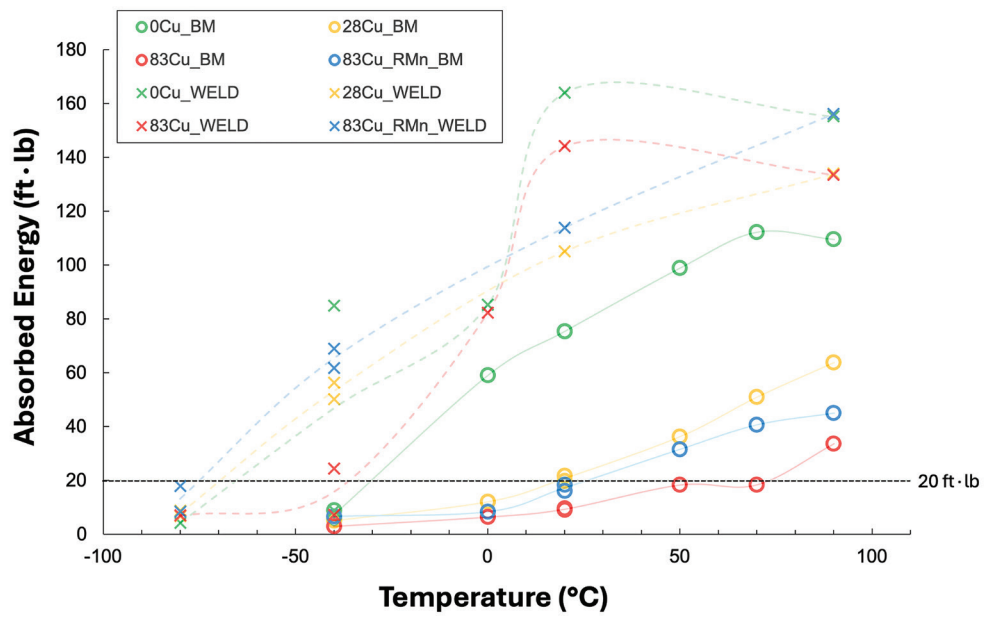


Fig. 8 — Charpy impact energy for base metal and GMAW HAZ samples.

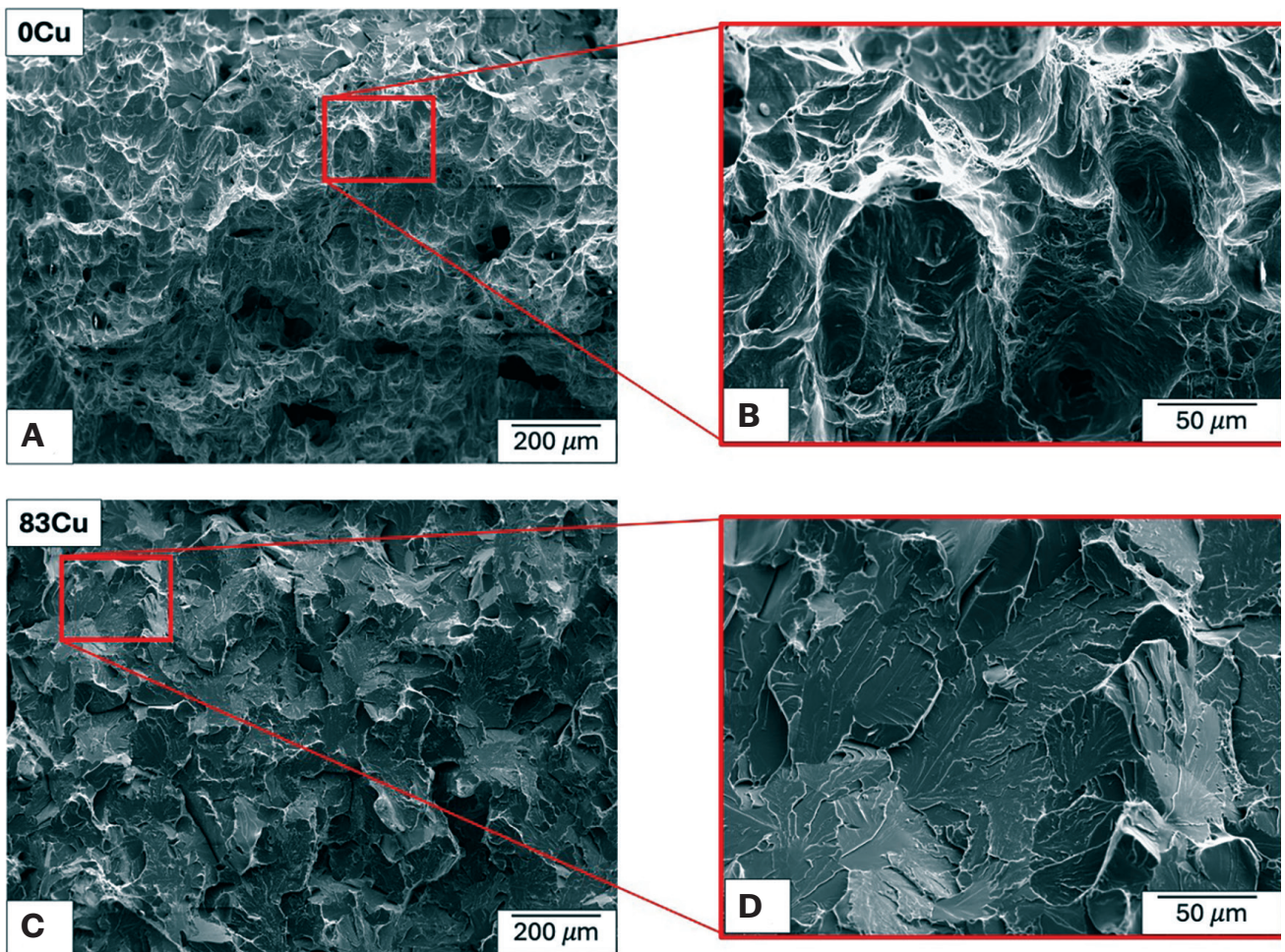


Fig. 9 — SEM SE fractographs from the central flat region of broken base metal Charpy specimens: A–B — 0Cu; C–D — 83Cu; both tested at 20°C.

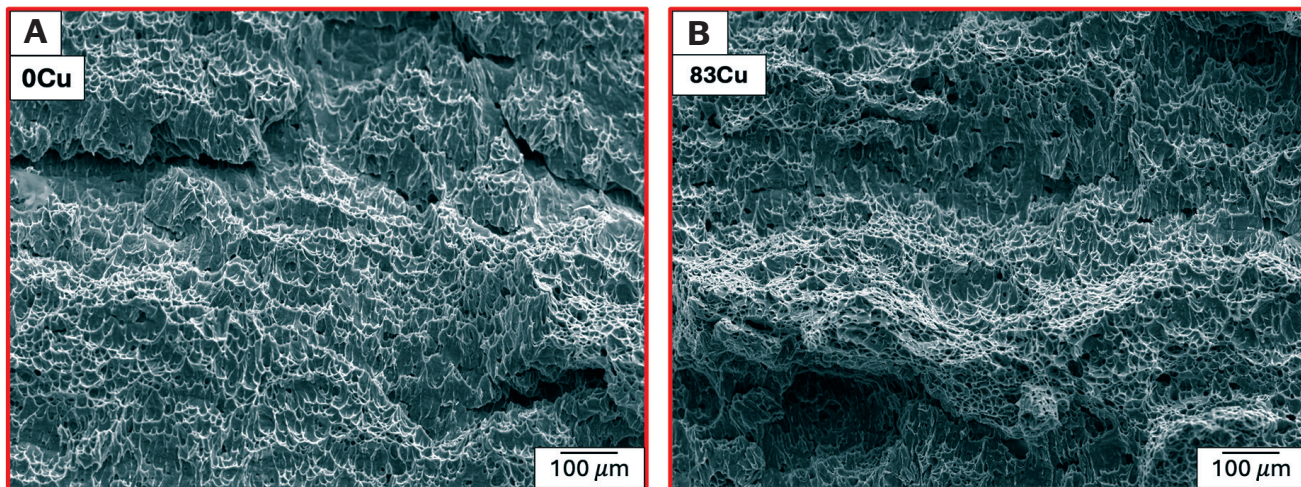


Fig. 10 — SEM SE fractographs from the central flat region of broken GMAW HAZ Charpy specimens: A — 0Cu; B — 83Cu; both tested at 20°C.

decreasing amounts of proeutectoid ferrite are observed with increasing copper content. Furthermore, the austenite present at the eutectoid temperature transformed to pearlite with 0Cu, but formed more brittle constituents, such as bainite, martensite, and potentially retained austenite with higher copper content. Clearly, furnace cooling from the hot rolling temperature provided a cooling rate such that differences in hardenability of the austenite present at the eutectoid temperature manifested as different transformation products.

It should be noted that there may be greater differences in local hardenability than are considered in Table 2. The non-ferritic constituents in the base metal appear in bands as shown in Fig. 3, which is often attributed to residual manganese segregation from solidification (Ref. 10). Copper also segregates to the liquid during solidification, which means the bands of non-ferritic microconstituent are likely regions of residual copper and manganese segregation. The areas of non-ferritic constituent that were formerly the austenite present above the eutectoid temperature may be more enriched in manganese and copper than predicted from equilibrium simulations, so larger differences in hardenability may be present between the alloys. Unintended variations in other alloying elements, such as silicon or molybdenum, while not as large in magnitude, may also influence microstructure evolution upon cooling.

In the case of the HAZ of the GMAWs, there are much smaller differences in microstructure as a function of copper content. The higher cooling rate in the CGHAZ likely led all the alloys to transform to martensite upon initial cooling, followed by tempering during the thermal cycles of subsequent welding passes. In other words, the cooling rate in the CGHAZ was fast enough that differences in hardenability do not manifest in different transformation products. The peak hardness values in the HAZ do appear to increase with increasing copper content, as shown by Fig. 7. The differences in peak hardness observed are larger than the differences in copper solid solution strengthening estimated in prior work (~15 HV) of the same materials (Ref. 11). Because the same welding parameters were used for all the materials and similar thermal profiles are expected, differences in peak HAZ hardness are likely attributable (in part) to differences in tempering behavior. The higher-cop-

per-containing materials showed some degree of tempering resistance, leading to higher hardness in the tempered martensite in the HAZ (Ref. 12).

Effects of Transformation Behavior on Charpy Impact Toughness

The Charpy impact curves for each base metal shown in Fig. 8 highlight a substantial shift with the addition of any copper, compared to 0Cu. Namely, the 20 ft-lb transition temperature shifts from ~ -30°C with the ferrite + pearlite microstructure of the 0Cu alloy to 30 – 50°C with the higher copper alloys with ferrite + MBA microstructures. It is likely that the more brittle behavior of the higher-copper alloys is due to pearlite being replaced by more brittle MBA constituents. As expected from the subtle differences in microstructure, the GMAW HAZ impact energies changed negligibly with increasing copper content. All of the alloys showed tempered martensite in the HAZ, resulting in lower transition temperatures and higher absorbed energy at each temperature. The only minor difference in impact behavior observed with the GMAW HAZ samples was that the 83Cu alloy showed a transition temperature of ~ -40°C, whereas all the other alloys showed a transition temperature of ~ -70°C. Additionally, not all GMAW HAZ samples were tested at 0°C, which impacted the final appearance of the DBTT curves in the transition region. The peak hardness in the HAZ in the 83Cu alloy was higher than the other alloys, which means the lesser degree of tempering experienced may have contributed to the slightly higher transition temperature.

Outlook

The influences of increasing residual copper on austenite stability and hardenability were clearly manifested in the base metal microstructures. The higher hardenability of 0.28–0.83 wt-% Cu resulted in more brittle microconstituents (compared to pearlite) forming during furnace cooling after hot rolling. Consequently, slower cooling rates must be utilized to achieve ferrite + pearlite microstructures with elevated

residual copper. Alternatively, this work also explored alloy modification to counteract the effects of increasing copper. Other alloying elements, such as manganese, can be reduced, guided by the carbon equivalent, to modify transformation behavior during base metal processing. However, more drastic changes than those evaluated here are needed to maintain a ferrite + pearlite microstructure after furnace cooling from hot rolling. Although increasing copper showed negligible effects on microstructure and toughness of the HAZ from GMAW in this work, higher heat input welding that produces slower cooling rates may result in less desirable microstructural features in the ICHAZ with increasing copper content. Overall, copper content does influence transformation behavior that can lead to lower toughness microstructures, so base metal composition, base metal processing, and potentially welding parameters may need to be adjusted to produce high copper-containing scrap-based steels for structural applications.

Conclusion

The goal of this work was to assess the effects of increased copper content on the microstructure and toughness of base metals and HAZs of GMAWs on low-carbon steel plate. Alloys were designed to have increasing copper content from 0–0.83 wt-%. Carbon equivalent calculations were made to evaluate reduced Manganese content to counteract the increase in hardenability from increasing copper. Changes in microstructural constituents were characterized by using microscopy for both the base metal and GMAW HAZ. Hardness profiles of the heat-affected zone were measured at both the root and cap sections of the V-groove weld. Variations in Charpy impact fracture toughness of both the base metal and the HAZ were assessed and interpreted in terms of the effects of increasing copper on microstructure evolution. The conclusions of this work are as follows:

- Increases in copper content resulted in a shift from a ferrite + pearlite to a ferrite + MBA constituents in the base metal due to increased hardenability. The amount of non-ferritic constituents also increased with copper content, due to greater stabilization of the austenite phase.
- Charpy impact testing showed that the shift from ferrite + pearlite to ferrite + MBA in the base metal from increasing copper content resulted in lower toughness and increased 20 ft-lb transition temperature.
- The HAZ of multi-pass GMAWs contains primarily tempered martensite in the CGHAZ and a mixture of ferrite and tempered martensite in the ICHAZ regardless of copper content. The HAZ hardness was higher with higher copper contents, which may be due to the tempering resistance imparted by higher copper content.
- Charpy specimens with notches containing the full HAZ showed overall higher toughness and lower 20 ft-lb transition temperature than their base metal counterparts, owing to the tempered martensite microstructure. Increasing copper content had minor influence on the Charpy impact behavior of the HAZ, where only the highest copper content alloy (also with nominal manganese) showed a higher 20 ft-lb transition temperature.
- The reduction of manganese in the 0.83 wt-% copper alloy was shown to lower peak hardness in the HAZ, as well as lower the 20 ft-lb transition temperature in alignment with the lower copper content alloys.

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