Effects of Copper, Nickel and Boron on Mechanical Properties of Low-Alloy Steel Weld Metals Deposited at High Heat Input

Boron appears to have a detrimental effect on Charpy V-notch toughness when the alloy mix also includes copper and nickel

BY A. O. KLUKEN, T. A. SIEWERT AND R. SMITH

ABSTRACT. We studied the effects of copper, nickel and boron on the mechanical properties of low-alloy steel weld metals deposited at high heat input (4.8 kJ/mm) by the submerged arc process. The copper and nickel contents of the welds were systematically varied within the ranges of 0.03 to 0.89 wt-% Cu and 0.01 to 1.54 wt-% Ni. In addition, several of these copper and nickel combinations were duplicated with welds containing 36 to 44 ppm B. Tensile testing revealed yield strengths, $R_{p0.2}$, in the range from 462 to 546 MPa, and ultimate tensile strengths, R_m, from 638 to 869 MPa. The weld metal Charpy V-notch (CVN) data showed a 35 J transition temperature, ITT_{35J}, ranging from –28° to 55°C. Upon adding boron, patches of intergranular fracture were present on the CVN fracture surfaces. This was particularly true when boron was added at the higher copper and nickel levels. Qualitative metallographic examinations revealed the presence of extensive amounts of particles at the prior austenite grain boundaries in boron containing welds. Furthermore, adding boron at the higher copper and nickel levels promoted the formation of the MAC (martensite-austenite-carbide) microconstituent. Also nickel seemed to promote formation of MAC in reheated weld metal. At this high heat input (4.8 kJ/mm), additions of copper, nickel and boron did not improve the mechanical properties.

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Introduction

The major impetus for development of high-strength low-alloy (HSLA) steels has been the need for: 1) higher strength; 2) improved toughness, ductility and formability; and 3) increased weldability. In order to meet these contradictory requirements, the carbon content of steel was progressively lowered to below 0.10 wt-% C. The desired strength is largely achieved by adding alloying elements such as chromium, molybdenum and copper, combined with a refinement of the ferrite grain size. The ferrite grain size is refined through additions of microalloying elements such as aluminum, vanadium, niobium and titanium in combination with various forms of thermomechanical processing (Refs. 1, 2). In addition, nickel is added to improve the ferrite cleavage resistance. Quenched and tempered HSLA steels are currently produced with a minimum yield strength in the range 680 to 884 MPa, combined with good toughness (Ref. 2).

KEY WORDS

Submerged Arc Welds
Mechanical Properties
Weld Metals
Low-Alloy Steels
High Heat Input
Effects of Copper
Effects of Boron
Effects of Nickel
Tensile Strength
Fracture Resistance

However, the use of these steels in structures such as bridges, and in offshore and arctic installations, calls for better welding consumables that produce weld metal deposits with mechanical properties matching those of the base metal. Over the past decade, the bulk of research on weld metals has been directed toward achieving a maximum toughness and ductility for a yield strength level below about 600 MPa, through control of the weld metal microstructure (Ref. 3). Microstructures consisting primarily of acicular ferrite provide optimum weld metal mechanical properties, for both strength and toughness (Refs. 4–7).

The objective of the present investigation was to clarify the effects of copper, nickel and boron on the strength and toughness of low-alloy steel weld metals deposited at high heat input. These effects were studied through systematic variation of the electrode compositions.

Experimental Details

Materials and Welding Performance

For the present investigation, twelve 4-mm (0.16-in.) diameter tubular electrodes were formulated, filled, and drawn to their final diameters. Test plates, ASTM A36, were nominally 400 mm long by 150 mm wide by 25 mm thick

Table 1 — Chemical Compositions of Experimental Welds. All Elements in wt-%

Weld No.	C	Mn	Si	P	S	Al	Ti	Cu	Ni	B	Cr	Mo	V	O	N	Pcm[a]
1	0.07	1.72	0.53	0.016	0.012	0.021	0.024	0.03	0.01	0.0007	0.03	0.01	0.01	0.0176	0.0054	0.182
2	0.08	1.84	0.59	0.016	0.013	0.024	0.023	0.38	0.01	0.0008	0.05	0.01	0.01	0.0155	0.0054	0.219
3	0.08	1.86	0.62	0.016	0.013	0.022	0.023	0.81	0.01	0.0010	0.05	0.01	0.01	0.0174	0.0054	0.244
4	0.08	1.83	0.58	0.016	0.013	0.023	0.023	0.03	0.45	0.0006	0.05	0.01	0.01	0.0171	0.0055	0.217
5	0.08	1.84	0.59	0.015	0.013	0.024	0.025	0.40	0.45	0.0009	0.05	0.01	0.01	0.0168	0.0049	0.228
6	0.08	1.85	0.62	0.017	0.013	0.024	0.022	0.81	0.45	0.0010	0.05	0.01	0.01	0.0152	0.0052	0.250
7	0.08	1.83	0.62	0.014	0.013	0.020	0.023	0.03	1.51	0.0007	0.05	0.01	0.01	0.0185	0.0067	0.227
8	0.07	1.85	0.59	0.014	0.013	0.020	0.025	0.42	1.50	0.0009	0.05	0.01	0.01	0.0146	0.0057	0.237
9	0.08	1.92	0.66	0.015	0.013	0.023	0.026	0.89	1.54	0.0013	0.05	0.01	0.01	0.0148	0.0051	0.279
10	0.09	1.90	0.63	0.015	0.013	0.019	0.025	0.04	0.01	0.0036	0.05	0.01	0.01	0.0165	0.0056	0.230
11	0.08	1.94	0.67	0.015	0.013	0.022	0.028	0.43	0.51	0.0040	0.05	0.01	0.01	0.0142	0.0057	0.254
12	0.08	1.94	0.68	0.015	0.013	0.027	0.029	0.85	1.51	0.0044	0.05	0.01	0.01	0.0132	0.0055	0.294

(a) $Pcm = C + \frac{Si}{30} + \frac{Mn + Cr + Cu}{20} + \frac{Mo}{15} + \frac{V}{10} + \frac{Ni}{60} + 5B$

Table 2 — Weld Metal Mechanical Properties

Weld No.	Yield Strength $R_{p0.2}$ (MPa)	Tensile Strength R_m (MPa)	Elongation (%)	Reduction in Area (%)	35 J Transition Temperature ITT_{35J} (°C)
1	462	638	31	67	−28
2	507	694	30	58	9
3	523	714	30	53	−5
4	522	709	30	54	−11
5	500	731	28	49	11
6	504	743	27	46	18
7	515	737	26	48	0
8	515	810	24	41	30
9	542	829	26	51	33
10	466	761	29	44	37
11	489	807	27	42	51
12	546	869	21	28	55

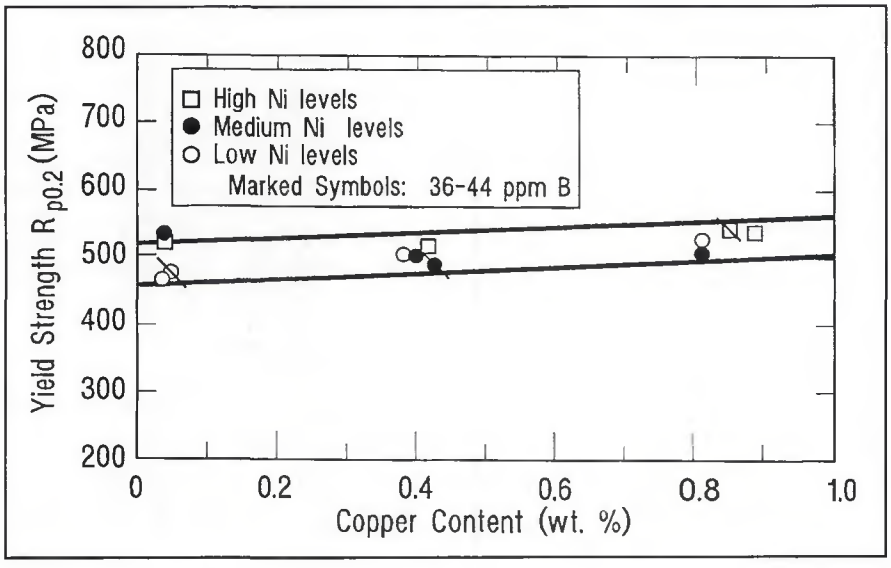


Fig. 1 — Effect of copper, nickel and boron on weld metal yield strength.

(15.7 x 6 x 1 in.), and the joint geometry was as specified in AWS A 5.23-80, *Specification for Low Alloy Steel Electrodes and Fluxes for Submerged Arc Welding*, except the included angle was opened to 50 deg (from 30) to reduce the dilution. Each test plate was mounted on two strongbacks to ensure high restraint. The beads were deposited with a mechanized submerged arc system under a commercial basic flux with basicity index, BI, equal to 3.1 (Ref. 8). Six layers were required to fill the joint, and each layer contained two beads, except the top layer, which contained three beads.

The current and the voltage were set at 700 A and 32 V, respectively, and the weld travel speed was held constant at 4.7 mm/s (11 in./min), giving a nominal heat input of 4.8 kJ/mm (122 kJ/in.). The interpass temperature was 200°C (392°F).

Chemical Analysis

The weld metal composition was determined with a combination of specific element analyzers (carbon, sulfur, oxygen and nitrogen) and an emission spectrometer for the remaining elements.

Mechanical Testing

Full-size Charpy V-notch (CVN) specimens were machined transverse to the welding direction with their upper surfaces approximately 2 mm (0.08 in.) below that of the base plate surface and the notches close to the weld centerline. These specimens were used to establish the full transition curve for each weld and to determine the 35-J impact transition temperature, ITT_{35J}.

In addition to impact testing, the weld metal strength and ductility were assessed using tensile specimens taken from the central region of the welds, with the longitudinal axes of the specimens parallel to the welding direction. Tensile testing was performed at a constant

cross-head speed of 0.5 mm/min (0.02 in./min) and a gauge length of 25.4 mm (1 in.).

Fractographic and Metallographic Examinations

Selected broken CVN specimens were examined with a scanning electron microscope (SEM) to determine the fracture mode. In addition, one unbroken CVN specimen from each weld was prepared according to standard techniques for optical metallographic examination. The microstructure of the notch area was subsequently revealed by etching in a 2% nital solution.

Results and Discussion

Weld Metal Chemical Composition

The chemical compositions of the experimental welds are given in Table 1. As seen from this table, the copper and the nickel contents of the welds were systematically varied within the following ranges:

Cu: 0.03 to 0.89 wt-%, and Ni: 0.01 to 1.54 wt-%.

Only minor variations were observed for the other elements, except for welds 10 through 12, which had intentionally higher boron contents (36 to 44 ppm B). Contribution from the base plate (ASTM A36) was minimal because of the 13-mm (0.5-in.) root opening and the greater included angle.

Based on previous investigations (Refs. 9, 10), it is somewhat surprising to note the low oxygen content of the welds. However, this is attributed to the higher deoxidation potential of the flux used in the present study than the flux used in these previous studies. The submerged arc flux used in this investigation had the approximate composition 21% Al_2O_3, 26% CaF_2, 21% MgO, 14% SiO_2, 7% CaO, and 6% $CaCO_3$.

We also tried to relate the results to the hardenability of the weld metal. From the various empirical measures of hardenability or susceptibility to hydrogen cracking, we selected P_{cm}. As a result of the variations in the weld metal copper and nickel contents, the carbon equivalent (P_{cm}) values are also seen to vary within relatively wide limits (0.182 to 0.294). This seems to imply that transformation behavior further complicates the comparison of the mechanical properties in the present investigation.

Weld Metal Tensile Properties

The tensile test results are summarized in Table 2. A closer inspection of the data reveals a yield strength, $R_{p0.2}$, in the range from 462 to 546 MPa, and a

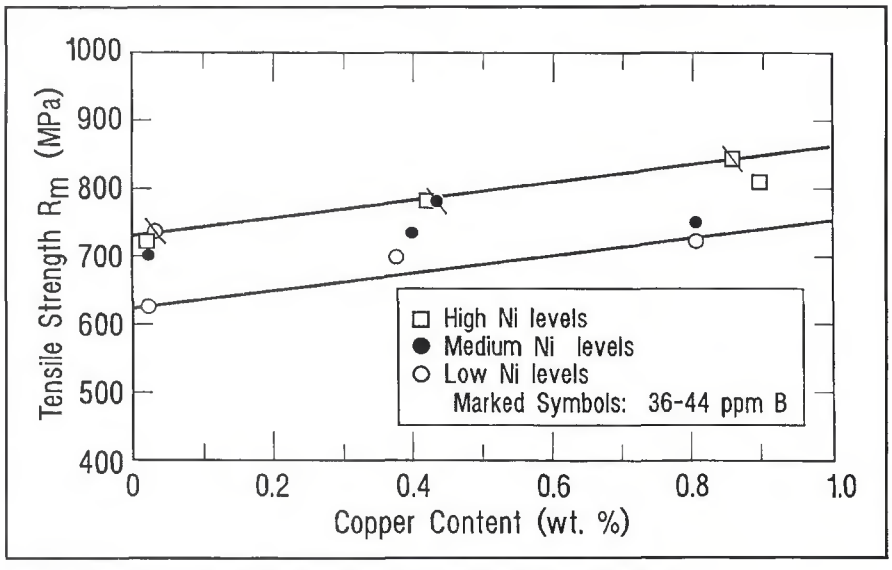


Fig. 2 — Effect of copper, nickel and boron on weld metal tensile strength.

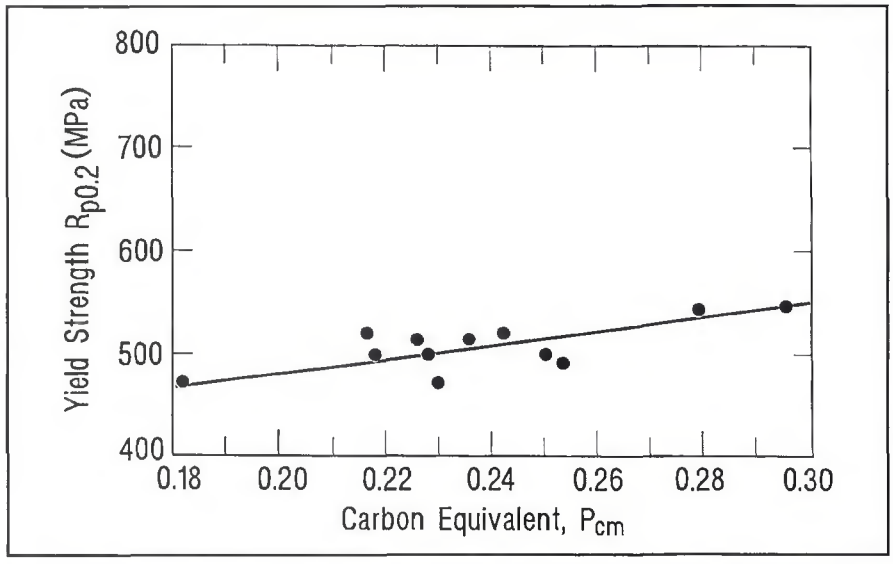


Fig. 3 — Relationship between weld metal yield strength and the P_{cm} carbon equivalent.

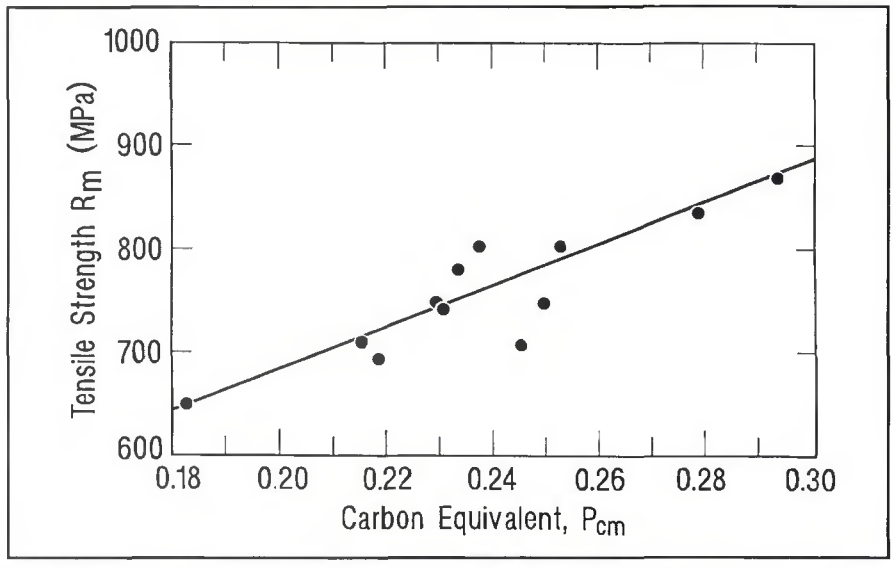


Fig. 4 — Relationship between weld metal tensile strength and the P_{cm} carbon equivalent.

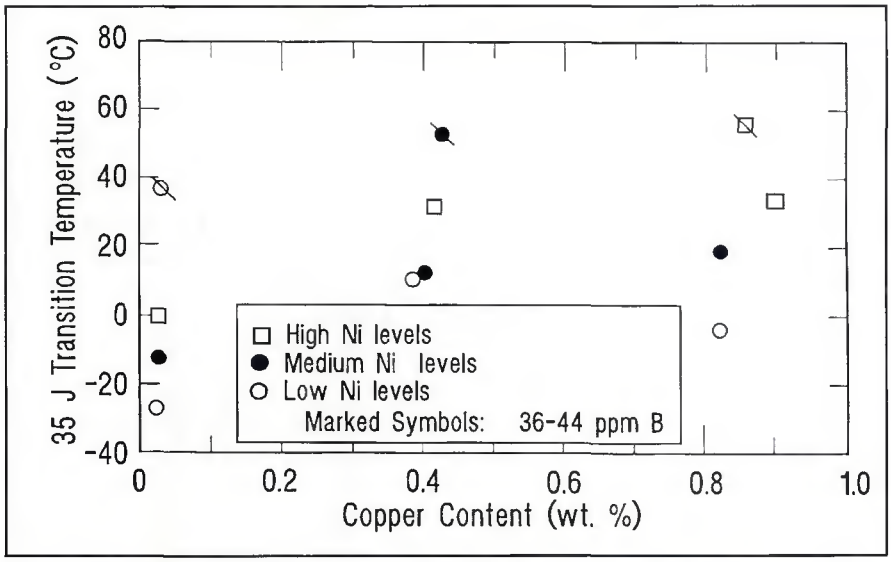


Fig. 5 — Effect of copper, nickel and boron on the 35-J impact transition temperature.

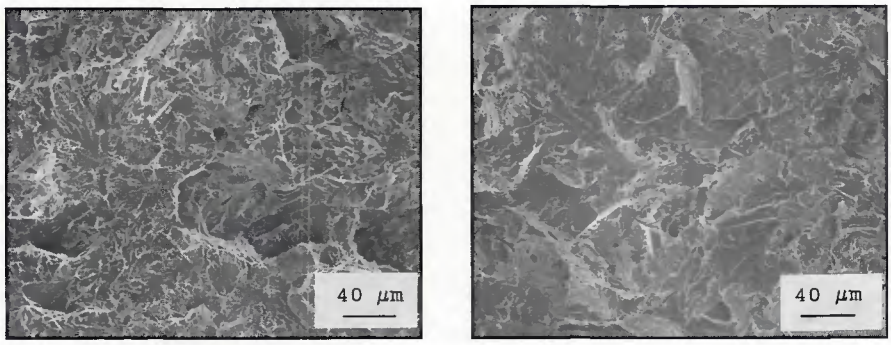


Fig. 6 — Fracture appearance of: A — weld 1 (0.03 wt-% Cu, 0.01 wt-% Ni, 7 ppm B); B — weld 10 (0.04 wt-% Cu, 0.01 wt-% Ni, 36 ppm B).

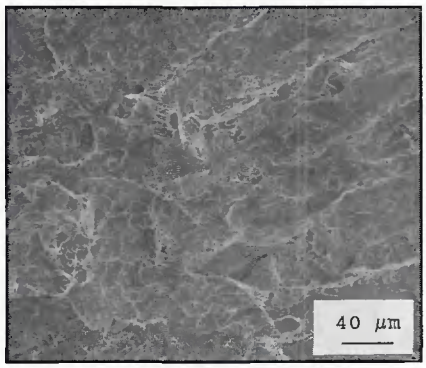


Fig. 7 — Fracture appearance of weld 9 (0.89 wt-% Cu, 1.54 wt-% Ni, 13 ppm B).

tensile strength, R_m, from 638 to 869 MPa. In addition, elongation at fracture and reduction in area remained above 21% and 28%, respectively.

The yield and the tensile strengths are plotted vs. the weld metal copper content in Figs. 1 and 2, respectively. The yield strength did not significantly increase with increasing copper content. Furthermore, the interactive effects of copper, nickel and boron are seen to be weak. On the other hand, the tensile strength increased strongly (about 15%) with increasing copper content. Additions of nickel and/or boron at a given copper content are also seen to increase the tensile strength.

In Figs. 3 and 4, the yield and the tensile strengths are plotted vs. P_{cm}, the carbon equivalent. Although there is considerable scatter in the data, these plots exhibit the same general trends observed in Figs. 1 and 2.

Weld Metal Toughness

The results from the CVN testing are summarized in Table 2. As seen from the table, the 35-J transition temperature, ITT_{35J}, ranges from –28° to 55°C (18.4° to 131°F).

Figure 5 shows a plot of the transition temperature vs. the weld metal copper content. It is obvious from the plot that copper additions are harmful to toughness. Also, nickel is seen to increase the transition temperature, particularly at low and high copper levels. This observation is somewhat surprising considering the beneficial effect of nickel on the ferrite cleavage resistance, but may be due to the high manganese content (compared to many submerged arc electrodes) of these welds (Ref. 11). It is speculated that the higher than usual manganese content may overshadow the expected beneficial effect of nickel on the ferrite cleavage resistance. Boron is especially detrimental to toughness at all copper and nickel levels examined in the present investigation. At low copper and nickel levels, an increase in the boron content from 7 to 36 ppm shifted the transition temperature from –28° to 37°C (-18.4°–98.6°F). This shift in transition temperature was accompanied by a shift in the fracture mode, as shown in Fig. 6. It is evident from the figure that the weld with more boron exhibits some patches of intergranular fracture compared to the predominantly quasi-cleavage fracture mode of the weld with less boron. Also, the welds with more copper and nickel showed patches of intergranular fracture — Fig. 7. Especially pronounced was the intergranular fracture appearance of the weld which contained the most copper, nickel and boron — Fig. 8. These observations indicate a local decrease in fracture resistance along the prior austenite grain boundaries, combined with a strengthening of the matrix (Ref. 12). Particles on the fracture surface are also evident in Fig. 8B.

In Fig. 9, the 35-J transition temperature is plotted vs. the carbon equivalent, P_{cm}. The scatter in the data indicates that the overall microstructure does not solely control the toughness.

Weld Metal Microstructure

As was previously pointed out, no attempt was made to quantify the amount of the different microconstituents. However, from the qualitative optical examinations of the welds, some general remarks can be made.

As-deposited weld metal microstructure. Examples of as-deposited weld metal microstructures are shown in Fig. 10. As expected, weld 1 (low Cu and Ni) contained a large fraction of acicular ferrite — Fig. 10A. Upon adding boron, the microstructure became more blocky and the acicularity disappeared. In addition, particles (probably borocarbides) were present at the prior austenite grain boundaries — Fig. 10B. Adding copper refined the microstructure and increased the amount of grain boundary ferrite — Fig. 10C. These observations are consistent with previous investigations (Refs. 13, 14). Nickel also seemed to increase the amount of primary ferrite, and the microstructure did not exhibit a typical acicularity — Fig. 10D. This is contrary to the findings of Evans (Ref. 11), but may be due to the higher heat input applied in the present investigation. Simultaneous

additions of copper and nickel refined the microstructure. Furthermore, the content of primary ferrite decreased and the MAC constituent was present — Fig. 10E. This latter observation is more pronounced when boron, copper and nickel were added — Fig. 10F.

In addition, extensive amounts of particles were present at the prior austenite grain boundaries of this latter weld, as evidenced by the micrograph in Fig. 11. The extensive TEM studies necessary to determine the exact chemical composition of the precipitates could not be performed, but it is believed that the large particles were probably copper precipitates, based on the work of Hannerz (Ref. 14). Further, it is believed that the small ones may be borocarbides, $Fe_{23}(BC)_6$, based on the study by Kluken and Grong (Ref. 12). The above findings (interactions between the alloy additions and the microstructure) partly explain the presence of intergranular fracture previously observed, and in addition, due to the high heat input applied, precipitation of ε-Cu may have occurred, strengthening the matrix (Ref. 14). However, this remains to be verified.

Reheated weld metal microstructure. Figure 12 contains examples of reheated weld metal microstructures. A common feature of all welds contained in the present investigation was the lack of a well defined zone of reheated weld metal.

As with the as-deposited microstructure, additions of copper and/or nickel are seen to refine the reheated weld metal microstructure — Fig. 12C– E. In addition, nickel seems to promote formation of the MAC microconstituent —

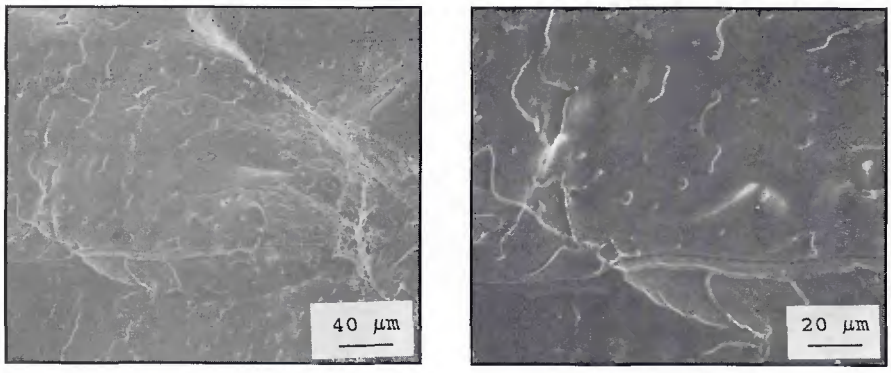


Fig. 8 — A — Fracture appearance of weld 12 (0.85 wt-% Cu, 1.51 wt-% Ni, 44 ppm B); B — close-up showing the presence of particles at the fracture surface.

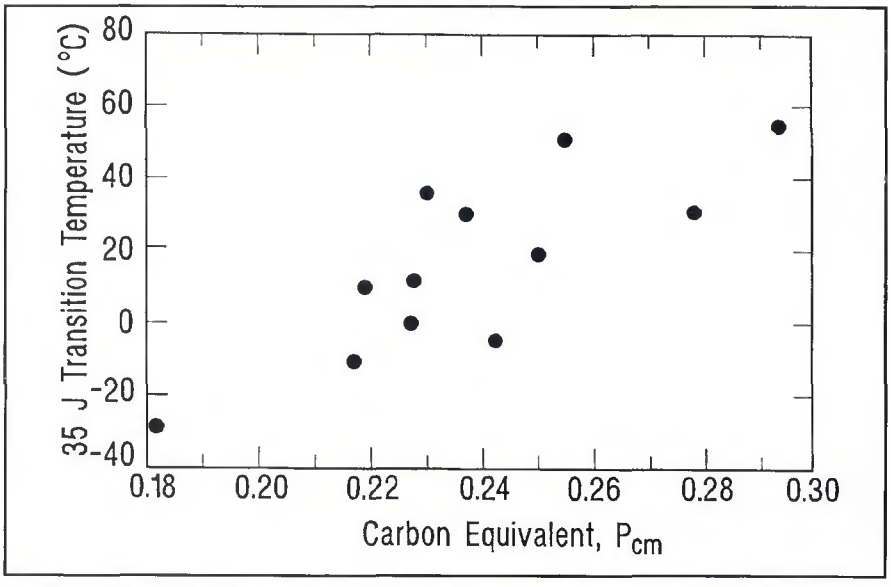


Fig. 9 — Relationship between the 35-J impact transition temperature and the P_{cm} carbon equivalent.

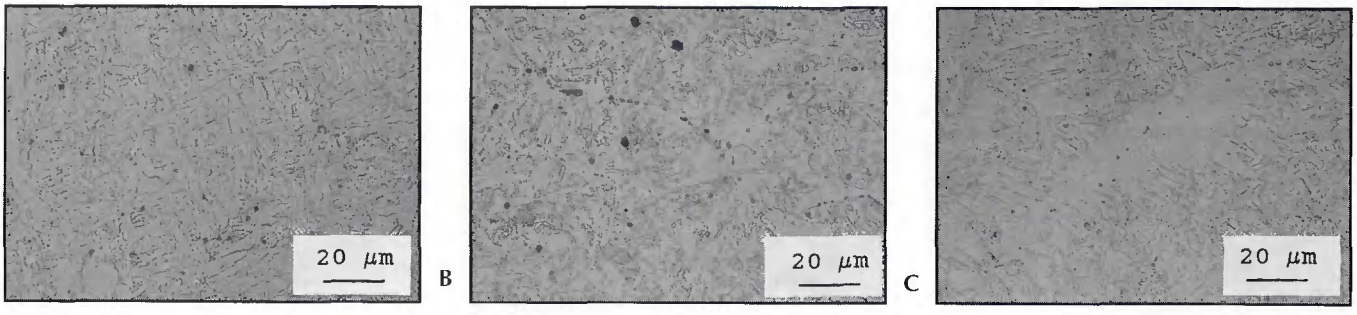
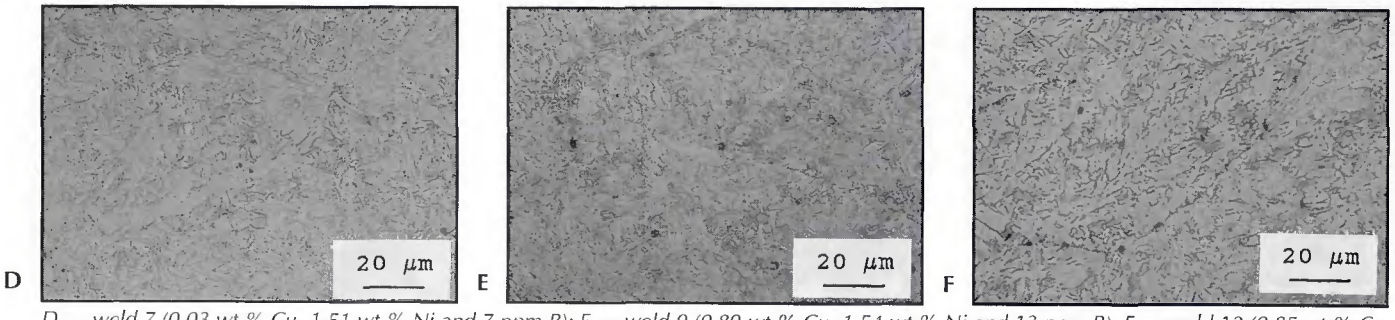


Fig. 10 — Examples of as-deposited weld metal microstructures. A — Weld 1 (0.03 wt-% Cu, 0.01 wt-% Ni and 7 ppm B); B — weld 10 (0.04 wt-% Cu, 0.01 wt-% Ni and 36 ppm B); C — weld 3 (0.81 wt-% Cu, 0.01 wt-% Ni and 10 ppm B);



D — weld 7 (0.03 wt-% Cu, 1.51 wt-% Ni and 7 ppm B); E — weld 9 (0.89 wt-% Cu, 1.54 wt-% Ni and 13 ppm B); F — weld 12 (0.85 wt-% Cu, 1.51 wt-% Ni and 44 ppm B).

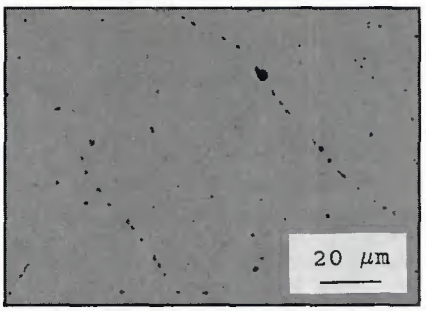


Fig. 11 — Micrograph of polished sample from weld 12 showing precipitates along prior austenite grain boundaries.

present in the microstructure — Fig. 12B and F.

Conclusions

1. At this high heat input (4.8 kJ/mm), the yield strengths of the welds showed no significant increase when copper, nickel and boron were added. The tensile strengths, on the other hand, exhibited a considerable increase with increasing contents of copper, nickel and boron in the welds, due to enhanced work hardening.

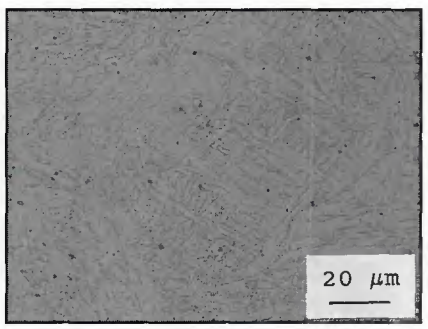
2. Both copper and nickel increased the 35-J CVN transition temperature. Boron was especially detrimental to toughness at all copper and nickel levels examined, primarily because it promoted the formation of precipitates at prior austenite grain boundaries.

3. Intergranular fracture was present on the CVN fracture surfaces of the high-boron containing welds, but also on the welds with the higher copper and nickel contents. Intergranular fracture was especially pronounced for the weld which contained the most copper, nickel and boron.

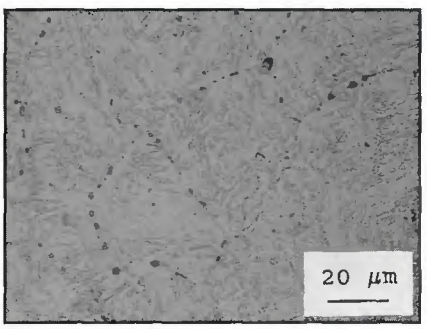
4. Substantial numbers of particles were present on the prior austenite grain boundaries when boron was added. For the welds with the highest copper and nickel contents, these particles were probably copper precipitates and borocarbides.

5. The presence of particles locally decreased the fracture resistance along the prior austenite grain boundaries and, in addition, the high heat input applied may have caused precipitation of ε-Cu, strengthening the matrix. Consequently, intergranular fracture may occur.

6. Additions of boron at high copper and nickel levels, tended to promote formation of the MAC microconstituent, further adding to the brittle behavior of the welds. Also nickel seemed to promote formation of MAC in reheated weld metal.

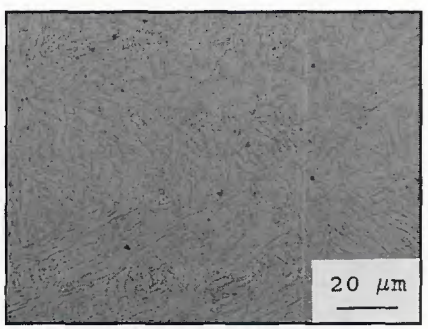


A

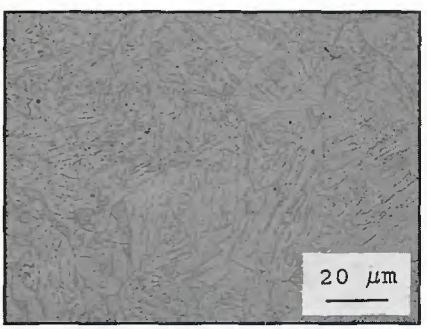


B

Fig. 12 — Examples of reheated weld metal microstructures. A — Weld 1 (0.03 wt-% Cu, 0.01 wt-% Ni and 7 ppm B); B — weld 10 (0.04 wt-% Cu, 0.01 wt-% Ni and 36 ppm B);

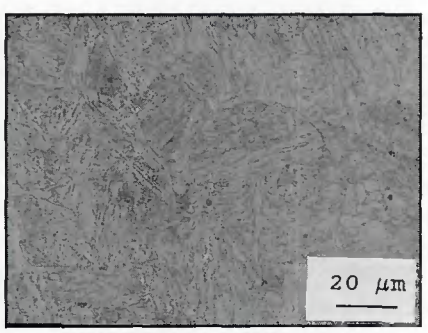


C

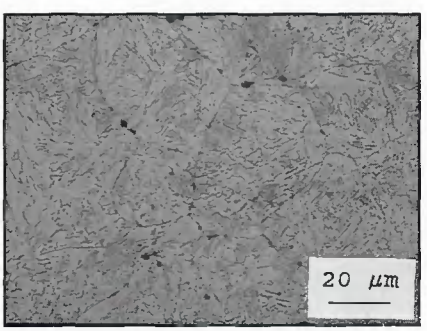


D

C —weld 3 (0.81 wt-% Cu, 0.01 wt-% Ni and 10 ppm B); D — weld 7 (0.03 wt-% Cu, 1.51 wt-% Ni and 7 ppm B);



E



F

E — weld 9 (0.89 wt-% Cu, 1.54 wt-% Ni and 13 ppm B); F — weld 12 (0.85 wt-% Cu, 1.51 wt-% Ni and 44 ppm B).

Fig. 12D. The effect of boron is also seen to be essentially the same as that occurring in as-deposited weld metal, the prior austenite grain boundaries are decorated with precipitates, and at high copper and nickel levels, a large amount of MAC is

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RECOMMENDATIONS PROPOSED BY THE PVRC COMMITTEE ON REVIEW OF ASME NUCLEAR CODES AND STANDARDS

Approved by the PVRC Steering Committee

The ASME Board on Nuclear Codes and Standards (BNCS) determined in 1986 that an overall technical review of existing ASME nuclear codes and standards was needed. The decision to initiate this study was reinforced by many factors, but most importantly to capture a pool of knowledge and "lessons learned" from an existing generation of technical experts with codes and standards background.

Project responsibility was placed with the Pressure Vessel Research Council, and activity was initiated in January 1988. Direction was vested in a Steering Committee which had overview of six subcommittees.

Recommendations were provided by nuclear utilities and industry and these recommendations were combined with independent considerations/recommendations of the PVRC Subcommittees and Steering Committees.

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