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Ductility and Toughness of Stainless Steel Welds

The ductility and toughness of autogenous gas tungsten-arc welds in 18Cr-2Mo ferritic stainless steels were influenced by the level of interstitials and stabilizing elements

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ABSTRACT. A series of 18Cr-2Mo ferritic stainless sheet steels was produced to evaluate the influence of interstitial content (carbon and nitrogen) and stabilizing elements (titanium and columbium) on the ductility and toughness of autogenous gas tungsten-arc (GTA) welds. The aluminum content was also varied in some steels. The weld ductility was evaluated by the Olsen cup test and conventional bend tests. The weld metal impact toughness was evaluated using quarter-size Charpy V-notch specimens.

With stabilizing additions near the minimum required to prevent intergranular corrosion, and about 0.03% total carbon plus nitrogen, excellent ductility and toughness were obtained in 18Cr-2Mo welds. Compared to the columbium alloyed steels, the titanium alloyed steels exhibited a wider range of stabilizing element content over which the alloys were not susceptible to intergranular corrosion or loss in weld ductility as evaluated by the Olsen cup test. The columbium stabilized weld metals exhibited good impact toughness at all levels of columbium investigated, and the titanium stabilized weld metals exhibited good impact toughness with up to about 0.50% Ti. Aluminum additions significantly lowered the impact transition temperature of a columbium stabilized weld but caused a slight decrease in the ductility. A post-weld heat treatment improved the ductility of the high titanium welds but had little effect on the columbium stabilized welds.

Titanium promoted the solidification of equiaxed grains in the center of the fusion zone. By contrast, the titanium-free weld metals solidified by conventional epitaxial growth. High columbium contents caused micro hot cracking which initiated brittle fracture during Olsen cup testing. Directional properties in the titanium-free weld metals were related to the preferred orientation developed during weld metal solidification.

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Introduction

It has been recognized for some time that the ductility and toughness of ferritic stainless steels can be improved by lowering the carbon and nitrogen contents to very low levels (Refs. 1,2) and that stabilizing elements such as titanium and columbium can be used to improve the ductility of steels having moderately low interstitial contents (Ref. 3). The development of new steel refining practices such as the AOD process (Ref. 4) has opened the way for the economic production of ferritic stainless steels that have moderately low interstitial contents, e.g., 0.02 to 0.06% total carbon plus nitrogen (Ref. 5).

Because of their excellent resistance to chlorides and stress corrosion cracking, the new ferritic stainless steels have been used initially for tubular applications that frequently require the use of autogenous gas tungsten-arc (GTA) welds (Ref. 6). This investigation was undertaken to determine the effects of composition on the ductility and toughness of autogenous GTA welds in stabilized 18Cr-2Mo ferritic stainless steels. Although molybdenum has been shown to have little effect on the toughness of ferritic stainless steels (Ref. 3), increasing the chromium content decreases the tolerance of the steels for carbon and nitrogen in base metals (Ref. 1) and welds (Ref. 7); therefore, the mechanical properties obtained from this investigation may not be directly applicable to the other types of ferritic stainless steels (e.g., the stabilized 26%Cr-1% Mo alloys), but the same general trends with respect to the effects of stabilizing additions and interstitial levels should hold for the higher chromium alloys.

Experimental Procedure

Base Metal Preparation

The experimental steels were induction melted and cast into four 2½ in. (64 mm) diam ingots. Buttons chill cast on each ingot were used for chemical analyses. Alloy additions were made between each split of the four heats yielding a total of 16 steels listed in Table 1. The ingots were hot forged at 2000 F (1090 C) to 1 in. (25 mm) thick slabs and then hot rolled at 2000 F (1090 C) to 0.5 in. (13 mm) thick plates. The plates were then cold rolled at 500 F (260 C) to 0.125 in. (3.2 mm) thick sheets. Steels stabilized only with titanium (1A-1D, 2A and 3B, Table 1) were annealed at 1550 F (840 C) for 5 minutes (10 minutes total time in the furnace). The other steels that contained columbium were annealed at 1800 F (980 C) for 5 minutes. All steels were water quenched from the annealing temperature. Pieces were cut from the 0.125 in. (3.2 mm) thick sheet and rolled further at 500 F (260 C) to 0.055 in. (1.4 mm) thickness. The steels were then annealed after rolling as described before. All steels were cleaned by grit blasting prior to welding, and portions of some welds were given a postweld anneal for the same time and temperatures given above for the base metals.

Table 1 — Chemical Analyses of the Steels Investigated

Heat	Steel	C	N	Mn	Si	Cr	Mo
		Element, wt % [(a)]					
1	1A	0.018	0.012	0.45	0.53	18.9	2.03
	1B	0.017	0.017	—	—	—	—
	1C	0.016	0.013	—	—	—	—
	1D	0.029	0.015	—	—	—	—
2	2A	0.017	0.016	0.43	0.48	18.2	1.87
	2B	0.016	0.018	—	—	—	—
	2C	0.019	0.017	—	—	—	—
	2D	0.030	0.021	—	—	—	—
3	3A	0.016	0.016	0.43	0.48	18.3	2.03
	3B	0.016	0.018	—	—	—	—
	3C	0.018	0.019	—	—	—	—
	3D	0.014	0.030	—	—	—	—
4	4A	0.017	0.016	0.46	0.48	18.2	1.97
	4B	0.016	0.018	—	—	—	—
	4C	0.014	0.017	—	—	—	—
	4D	0.016	0.032	—	—	—	—

(a) For all elements not analyzed, the composition was assumed to be essentially equivalent to the analysis of the nearest preceding split for which an analysis is given.
(b) Amount of stabilizer above the minimum required to prevent intergranular corrosion in the Strauss test.

Welding

Full penetration autogenous welds were performed in each steel of both 0.125 in. (3.2 mm) and 0.055 in. (1.4 mm) thickness using an automatic, single-pass, gas tungsten-arc (GTA) welding process. The rolling direction was always perpendicular to the welding direction. Welding conditions are given in Table 2, and the parameters required to obtain full penetration in each steel are listed in Tables 3 and 4.

A trailing shield was constructed because, when travel speeds above 25 in./min (64 cm/min) were used, the weld puddle became extremely long and required the use of additional shielding to prevent heavy oxidation of the weld metal. To eliminate shielding as a variable, a trailing shield was used for all welds. In preliminary work, welds were deposited in 18Cr-2Mo steels without the use of a trailing shield but at slower travel speeds. The impact toughness of these welds was similar to that obtained from welds deposited with a trailing shield.

Some welds were performed using a square butt joint geometry; however, curvature in the sheets made alignment and clamping extremely difficult. Since preliminary work demonstrated that a full penetration bead-on-plate weld had identical toughness to a square butt weld, all welds tested were of the bead-on-plate type which minimized the clamping and alignment problems.

Welds in the 0.055 in. (1.4 mm) thick sheet performed at 25 in./min (64 cm/min) were extremely smooth and flat — a useful feature during Olsen cup testing (described later). Welds performed at 100 in./min (250 cm/min) with this procedure exhibited some undercut; therefore, only the welds performed at 25 in./min (64 cm/min) were used for Olsen cup testing. All welds appeared sound upon visual inspection.

Modified Strauss Tests

Welds performed in 0.055 in. (1.9 mm) thick sheet of Steels 1A, 2A, 4A and 4B were tested for susceptibility to intergranular corrosion using the modified Strauss test (ASTM A262, Practice E). All other steels contained a higher amount of stabilizing element (see Table 1) and, thus, were less likely to be susceptible to intergranular corrosion.

After polishing and cleaning, the specimens were immersed in a solution containing 16% sulfuric acid and 6% copper sulfate by weight and were in good electrical contact with heavy copper turnings. The test was conducted in a boiling solution for 24 hours, and after exposure, the specimens were bent through 180 degrees with the weld at the apex of the bend. If no cracks could be detected at a magnification of X20, the weld was considered to have passed the test.

Olsen Cup Tests

The Olsen cup test is conventionally used to evaluate the formability of materials in biaxial tension, but, in this investigation, it was found to be a useful tool for evaluating weld ductility. The welds that were deposited with 25 in./min (64 cm/min) travel speed in the 0.055 in. (1.4 mm) thick sheet had an extremely smooth surface permitting cup testing without surface preparation. Welds in all steels were tested in the "as-welded" condition. In addition, some welds were tested after a postweld anneal, and most steels were also tested in the unaffected base metal.

The following procedure was used during Olsen cup testing. The sheets were clamped between smooth circular die plates with a load of 10 tons, and a 7/8 in. (22 mm) diam steel ball was then forced into the sheet using a ram speed of 1 in./min (2.5 cm/min). Welds were aligned so that the weld centerline contacted the top center of the steel ball. A 0.004 in. (0.1 mm) thick polyethylene sheet, coated on each side with an additive-free SAE 50-weight oil, was placed over the ball for lubrication. Draw-in was not observed in any of the specimens. A plot of load vs cup depth was automatically recorded and used to determine the fracture load and fracture depth for each test. At least five cups were tested for each steel in each condition.

Bend Tests

Bend tests were performed in selected columbium stabilized welds to provide a correlation between the Olsen cup test and a commonly used test for weld quality. Specimens measuring 1.0 in. (25 mm) by 1.5 in. (38 mm) were selected from welds in both 0.125 in. (3.2 mm) and 0.055 in. (1.4 mm) thick sheet. The weld was centered and aligned parallel to the long direction for longitudinal bends

S	P	Ni	Cu	Al	Ti	Cb	Excess Stabil. wt % [b]	Ti+Cb C+N
0.010	0.021	0.28	0.11	0.04	0.34	0	0.08	11.3
—	—	—	—	—	0.59	—	0.31	17.4
—	—	—	—	—	0.78	—	0.52	26.9
—	—	—	—	—	—	—	0.47	17.7
0.013	0.017	0.26	0.11	0.02	0	0.24	0.01	7.3
—	—	—	—	—	—	0.42	0.18	12.3
—	—	—	—	—	—	0.64	0.39	17.8
—	—	—	—	—	—	—	0.28	12.5
0.013	0.018	0.28	0.11	0.08	0.37	0	0.10	11.6
—	—	—	—	0.14	—	—	0.09	10.6
—	—	—	—	—	—	0.26	0.28	17.0
—	—	—	—	0.12	0.41	—	0.29	15.2
0.013	0.018	0.26	0.11	0.08	0	0.25	0.02	7.6
—	—	—	—	0.15	—	—	0.01	7.4
—	—	—	—	—	0.28	0.23	0.19	16.4
—	—	—	—	0.13	0.32	—	0.14	11.4

Table 2 — Welding Conditions

Torch
- Type: water cooled
- Cup size: 0.75 in. (19 mm) ID

Electrode
- Type: 2% thoriated tungsten
- Diameter: 0.125 in. (3.2 mm) for 0.055 in. (1.4 mm) thick sheet; 0.156 in. (4.0 mm) for 0.125 in. (3.2 mm) thick sheet
- Tip angle: 90 deg included
- Tip-to-work dist. (arc length): 0.050 in. (1.3 mm)
- Stickout: 0.18 in. (5 mm)

Shielding
- Type gas: argon
- Torch: 30 cfh (14 l/min)
- Back-up: 15 cfh (7 l/min)
- Tr. Shield: 25 cfh (12 l/min)

and parallel to the short direction for the transverse bends. Duplicate specimens for each condition were surface ground to remove irregularities and formed in the long direction with a guided bend jig through a 180 degree bend with a 2T radius (twice the sheet thickness). If no cracking was detected, the specimen was formed to a smaller bend radius by clamping until cracking occurred.

Impact Tests

Blanks for Charpy V-notch impact specimens were sectioned from the welds performed in 0.125 in. (3.2 mm) thick sheet. The specimens were machined to one-quarter the standard thickness (2.5 mm) and notched in the center of the weld metal. The specimens were tested over a range of temperatures and both impact energy and lateral expansion were recorded.

Hardness Tests

A hardness traverse was performed in each weld deposited in 0.125 in. (3.2 mm) thick sheet and for selected welds performed in 0.055 in. (3.2 mm) thick sheet. The average Vickers hardness was calculated for those impressions lying in the fusion zone.

Table 3 — Welding Parameters for 0.055 In. (1.4 mm) Thick Sheet

Heat	Steel	Type	Current A	Voltage V	Heat input kJ/in. (kJ/cm)
		Welds Performed at 25 ipm (64 cm/min)			
1	1A-1D	Ti	140-155	9.5-10.5	3.2-3.9 (1.3-1.5)
2	2A-2D	Cb	130-145	9.5-10.5	3.0-3.7 (1.2-1.5)
3	3A, 3B	Ti-Al	135-145	9.5-10.5	3.1-3.7 (1.2-1.5)
	3C, 3D	Ti-Cb-Al	130-140	9.5-10.5	3.0-3.5 (1.2-1.4)
4	4A, 4B	Cb-Al	130-140	9.5-10.5	3.0-3.5 (1.2-1.4)
	4C, 4D	Ti-Cb-Al	130-140	9.5-10.5	3.0-3.5 (1.2-1.4)
		Welds Performed at 100 ipm (250 cm/min)			
1	1A-1D	Ti	390-410	16-18	3.7-4.4 (1.5-1.7)
2	2A-2D	Cb	370-400	16-18	3.6-4.3 (1.4-1.7)
3	3B	Ti-Al	395-405	16-18	3.8-4.4 (1.5-1.7)
4	4B	Cb-Al	365-375	16-18	3.5-4.1 (1.4-1.6)

Table 4 — Welding Parameters for 0.125 In. (3.2 mm) Thick Sheet [a]

Heat	Steel	Type	Current A	Voltage V	Heat input kJ/in. (kJ/cm)
1	1A-1D	Ti	390-410	16-18	15.0-17.7 (6.0-7.0)
2	2A-2D	Cb	365-385	15-17	13.1-15.7 (5.2-6.2)
3	3A, 3B	Ti-Al	400-420	16-18	15.4-18.1 (6.0-7.1)
	3C, 3D	Ti-Cb-Al	440-460	16-18	16.9-19.9 (6.7-7.8)
4	4A, 4B	Cb-Al	460-480	16-18	17.7-20.7 (7.0-8.2)
	4C, 4D	Ni-Cb-Al	470-490	17-19	19.2-22.3 (7.5-8.8)

(a) All welds performed at 25 ipm (64 cm/min)

Metallographic Examination

Welds in all steels were sectioned for optical examination of both a top surface and a transverse cross section. Weld metal inclusions that were observed on as-polished sections of 18Cr-2Mo-Nb steels were analyzed with an energy dispersive spectrometer in an electron microprobe. Carbon extraction replicas were taken from selected weld metals by the following procedure. Polished specimens were etched electrolytically in an aqueous 50% HNO_3 solution (specimen cathode). The carbon film was deposited, and the film was stripped by electrolytically etching in a solution of 12.5% H_2SO_4 in methanol (specimen anode). The replicas were observed in a transmission electron microscope and electron diffraction patterns were obtained from the extracted particles. The fracture surfaces of impact and Olsen cup specimens were also observed in a scanning electron microscope.

Results

Chemical Analyses

The results of chemical analyses are presented in Table 1. The excess of titanium and/or columbium above the minimum required for stabilization against intergranular corrosion is also presented in Table 1. These minimums were determined from the following formulas that are reportedly valid up to at least 0.04% carbon plus nitrogen (Ref. 8):

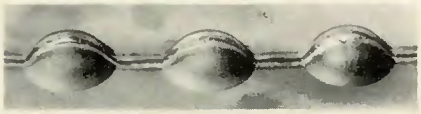
$$\%Ti\,(min) = 0.15 + 3.7\,(\%C + \%N)$$
$$\%Cb\,(min) = 7\,(\%C + \%N)$$

For alloys containing both titanium and columbium, it was assumed for this calculation that half of the interstitials combined with titanium and half with columbium although ti-

tanium is less soluble and probably combined with more than half of the interstitials. The weld metals were assumed to have the same average composition as the base metals because, in preliminary work, little change in the analysis was observed in this type of weld.

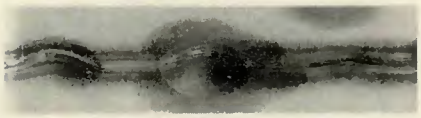
Modified Strauss Teats

Welds deposited in Steels 1A, 2A, 4A, and 4B were tested in the modified Strauss test, and no cracks could be detected after exposure; therefore all welds were considered to be resistant to intergranular corrosion.

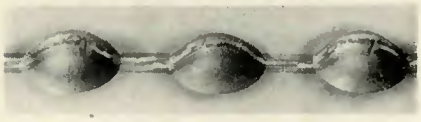


(a) Steel 1A (0.018% C and 0.34% Ti)

(b) Steel 1B (0.017% C and 0.59% Ti)



(c) Steel 1C (0.019% C and 0.78% Ti)



(d) Steel 1D (0.029% C and 0.78% Ti)

Fig. 1 — Olsen cup tests of as-deposited 18Cr-2Mo welds in 0.055 in. (1.4 mm) thick sheet showing the effect of titanium

Mechanical Tests

The results of base metal Olsen cup tests are presented in Table 5. There was little variation in the properties of the base metals. The Olsen cup fractures were all ductile and aligned parallel to the rolling direction. Generally, greater fracture depths corresponded to higher fracture loads.

The results of the Olsen cup tests performed on as-welded specimens are presented in Table 6, and typical Olsen cup fractures are shown in Fig. 1. Some columbium stabilized welds exhibited large deviations in cup depth (Steels 2B, 4A, and 4B — Table 6) and both ductile and brittle cup fractures in the same weld. Ductile fractures in the welds were aligned both parallel and transverse to the welding direction. Brittle fractures in the titanium stabilized steels were oriented parallel to the welding direction (Fig. 1c). Brittle fractures in the columbium stabilized steels were usually aligned transverse to the welding direction, and longitudinal cracks, when they occurred, were frequently associated with a crack oriented transverse to the welding direction. The brittle transverse cracks did not extend beyond the welded region.

An increase in the titanium or columbium content caused a decrease in the ductility as measured by the Olsen cup fracture depth. Figure 2 summarizes these results. These curves show that welds containing up to about 0.60% Ti exhibited an average Olsen cup fracture depth of over 0.3 in. (7.6 mm); however, the same ductility was obtained in welds containing up to only about 0.35% Cb. The difference is greater if atomic

Table 5 — Results of Olsen Cup Tests Performed on 18Cr-2Mo Base Metals

Steel	Additions, %	Average fracture depth in. (mm)	Std. dev. fracture depth in. (mm)
1A	0.34 Ti	0.40 (10.2)	0.02 (0.5)
1B	0.59 Ti	0.41 (10.4)	0.01 (0.3)
1C	0.78 Ti	0.37 (9.4)	0.01 (0.3)
1D	0.78Ti+0.03C	0.37 (9.4)	0.01 (0.3)
2A	0.24 Cb	0.39 (9.9)	0.01 (0.3)
2B	0.42 Cb	0.41 (10.4)	0.01 (0.3)
2C	0.64 Cb	0.40 (10.2)	0.02 (0.5)
2D	0.64Cb+0.03C	0.42 (10.7)	0.01 (0.3)
3A	0.37Ti+0.08Al	0.41 (10.4)	0.02 (0.5)
3B	0.37Ti+0.14Al	0.40 (10.2)	0.02 (0.5)
3C	0.37Ti+0.14Al+0.26Cb	0.43 (10.9)	0.02 (0.5)
3D	0.41Ti+0.12Al+0.26Cb+0.03N	0.42 (10.7)	0.01 (0.3)
4A	0.25Cb+0.08Al	0.42 (10.7)	0.02 (0.5)
4B	0.25Cb+0.15Al	0.39 (9.9)	0.01 (0.3)
4C	0.23Cb+0.15Al+0.28Ti	0.42 (10.7)	0.01 (0.3)
4D	0.23Cb+0.13Al+0.32Ti+0.03N	0.42 (10.7)	0.01 (0.3)

Table 6 — Results of Olsen Cup Tests Performed on 18Cr-2Mo Welds

				As-deposited welds		Welds postweld annealed [a]	
Steel	Additions, %	Excess stabilizer, wt %	Ti+Cb C+N	Average fracture depth, in. (mm)	Std dev. fracture depth, in. (mm)	Average fracture depth, in. (mm)	Std dev. fracture depth, in. (mm)
1A	0.34 Tl	0.08	11.3	0.37 (9.3)	0.04 (1.0)	—	—
1B	0.59 Ti	0.31	17.4	0.31 (7.8)	0.04 (1.0)	0.34 (8.6)	0.02 (0.5)
1C	0.78 Ti	0.52	26.9	0.22 (5.5)	0.07 (1.8)	0.37 (8.6)	0.02 (0.5)
1D	0.78Ti+0.03C	0.47	17.7	0.36 (9.1)	0.04 (1.0)	0.35 (8.9)	0.01 (0.3)
2A	0.24 Cb	0.01	7.3	0.35 (9.0)	0.02 (0.5)	—	—
2B	0.42 Cb	0.18	12.3	0.22 (5.5)	0.12 (3.0)	0.12 (3.0)	0.13 (3.3)
2C	0.64 Cb	0.39	17.8	0.06 (1.7)	0.04 (1.0)	0.15 (3.9)	0.13 (3.3)
2D	0.64Cb+0.03C	0.28	12.5	0.08 (2.1)	0.03 (0.8)	0.18 (4.6)	0.04 (1.0)
3A	0.37Ti+0.08Al	0.10	11.6	0.38 (9.7)	0.04 (1.0)	—	—
3B	0.37Ti+0.14Al	0.09	10.9	0.38 (9.8)	0.01 (0.3)	—	—
3C	0.37Ti+0.14Al+0.26Cb	0.28	17.0	0.36 (9.2)	0.02 (0.5)	—	—
3D	0.41Ti+0.12Al+0.26Cb+0.03N	0.29	15.2	0.38 (9.6)	0.03 (0.8)	—	—
4A	0.25Cb+0.08Al	0.02	7.6	0.20 (5.1)	0.12 (3.0)	0.26 (6.6)	0.15 (3.8)
4B	0.25Cb+0.15Al	0.01	7.4	0.23 (5.8)	0.13 (3.3)	0.36 (9.1)	0.02 (0.5)
4C	0.23Cb+0.15Al+0.28Ti	0.19	16.4	0.36 (9.2)	0.02 (0.5)	0.42 (10.7)	0.02 (0.5)
4D	0.23Cb+0.13Al+0.32Ti+0.03N	0.14	11.4	0.36 (9.1)	0.02 (0.5)	0.40 (10.2)	0.01 (0.3)

(a) Welds 1B, 1C, and 1D were annealed at 1550 F (840 C) and the other welds at 1800 F (980 C) for 5 minutes then water quenched.

percent is considered because 0.60% Ti is 5.5 times the stoichiometric equivalent of Ti(C,N) whereas 0.35% Cb is only 1.7 times the stoichiometric equivalent of Cb/C,N) in steels containing 0.03% total carbon plus nitrogen.

A postweld anneal caused a significant improvement in the ductility of the high titanium welds as shown in Table 6. The high columbium welds exhibited a slight improvement in ductility after the postweld heat treatment. but the deviation in cup height was large indicating brittle behavior in some regions of the welds.

The results of bend tests performed on columbium stabilized weld metals are presented in Table 7. Cracking, when it occurred, generally extended across the entire weld fusion zone in a direction perpendicular to the principal tensile stress. Cracks in the longitudinal bend specimens arrested in the heat-affected zone. Note that, in the thicker specimens, the minimum bend radius was larger for the longitudinal bend specimens indicating lower ductility when the strain was applied parallel to the welding direction.

The impact test results are summarized in Table 8. An increase in the amount of stabilizing element caused a decrease in the notch toughness although columbium additions were less detrimental than titanium. Higher aluminum contents in the columbium stabilized steel caused a large improvement in the transition temperature. Impact specimens from the 18Cr-2Mo-Cb-Al welds that were tested in the transition region occasionally exhibited stepped fracture surfaces (discussed later). Combined additions of titanium, columbium and aluminum were detrimental to the impact toughness.

The columbium stabilized welds exhibited higher hardness levels than the titanium stabilized welds (Table 9). Increasing titanium slightly increased the hardness, but increasing columbium decreased the hardness. To determine whether the hardness reading varied with the indenter load, a second traverse was performed in the 0.125 in. (3.2 mm) thick welds in Steels 1A and 2A using a 500 gram load. Average weld metal hardness values of 187 HV 0.5 for Steel 1A and 198 HV 0.5 for Steel 1B were only slightly higher than the corresponding values determined with a 10 kg load (Table 9). Consequently, the difference in hardness between the 0.055 in. (1.4 mm) and 0.125 in. (3.2 mm) welds was real and not the result of the different indenter loads.

Metallographic Observations

The weld metal grain structure was strongly influenced by the presence of titanium as shown in Figs. 3 and 4. All weld metals that contained titanium including those that contained both titanium and columbium exhibited grain structures typical of that shown in Fig. 3a. Near the center of the fusion zone the titanium alloyed steels exhibited an equiaxed dendritic structure (Figs. 3a and 4a), but the titanium-free steels exhibited epitaxial growth throughout the entire fusion zone (Figs. 3b and 4b). Weld metals in the 0.125 in. (3.2 mm) thick steels exhibited the same structures observed

Table 7 — Bend Test Results

		Limit of Bend Radius [a]			
		0.050 in. (1.3 mm) thick welds		0.095 in. (2.4 mm) thick welds	
Steel	Additions, %	Longit.	Transv.	Longit.	Transv.
2A	0.24 Cb	<1/4T	<1/4T	1/3T	<1/4T
2B	0.42 Cb	<1/4T	<1/4T	1/2T	<1/4T
2C	0.64 Cb	2T	Cracked at start	2T	1/2T
2D	0.64Cb+0.08Al	<1/4T	<1/4T	1/4T	<1/4T

(a) Bend radii are normalized with respect to the thickness (T).

Table 8 — Results of Quarter-Size Impact Tests of 18Cr-2Mo Weld Metals

		Transition temperature, F(C)	
Steel	Additions, %	7.5 ft-lb (10.2J) energy absorption	0.015 in. (0.38 mm) lateral expansion
1A	0.34 Ti	32 (0)	35 (0)
1B	0.59 Ti	92 (33)	102 (39)
1C	0.78 Ti	100 (38)	96 (36)
1D	0.78Ti+0.03C	138 (59)	132 (56)
2A	0.24 Cb	25 (-4)	28 (-2)
2B	0.42 Cb	50 (10)	52 (11)
2C	0.64 Cb	48 (9)	50 (10)
2D	0.64Cb+0.03C	52 (11)	100 (38)
3A	0.37Ti+0.08Al	68 (20)	58 (14)
3B	0.37Ti+0.14Al	64 (18)	66 (19)
3C	0.37Ti+0.14Al+0.26Cb	180 (82)	180 (82)
3D	0.41Ti+0.12Al+ 0.26Cb+0.03N	166 (74)	168 (76)
4A	0.25Cb+0.08Al	-52 (-47)	-52 (-47)
4B	0.25Cb+0.15Al	-52 (-47)	-52 (-47)
4C	0.23Cb+0.15Al+0.28Ti	182 (83)	162 (72)
4D	0.23Cb+0.13Al+ 0.32Ti+0.03N	170 (77)	168 (76)

Table 9 — Results of Hardness Tests Performed in Weld Metals

Steel	Additions, %	Avg hardness, HV-10 0.125 in. (3.2 mm) thick sheet	Avg hardness, HV0.5 0.055 in. (1.4 mm) thick sheet
1A	0.34Ti	180	218
1B	0.59 Ti	186	—
1C	0.78 Ti	187	221
1D	0.78Ti+0.03C	188	—
2A	0.24 Cb	196	247
2B	0.42 Cb	194	250
2C	0.64 Cb	191	213
2D	0.64Cb+0.03C	199	225
3A	0.37Ti+0.08Al	188	—
3B	0.37Ti+0.14Al	188	—
3C	0.37Ti+0.14Al+0.26Cb	198	—
3D	0.41Ti+0.12Al+ 0.26Cb+0.03N	199	220
4A	0.25Cb+0.08Al	186	225
4B	0.25Cb+0.15Al	186	—
4C	0.23Cb+0.15Al+0.28Ti	180	—
4D	0.23Cb+0.13Al+ 0.32Ti+0.03N	193	—

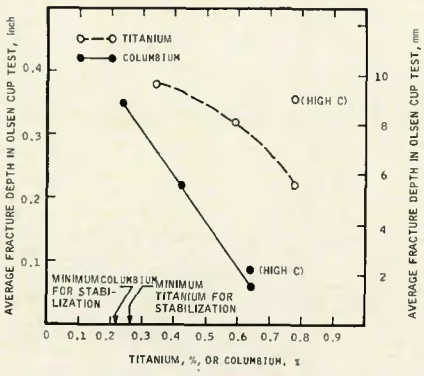


Fig. 2 — Effect of titanium and columbium on the ductility of welds in 18Cr-2Mo steels containing 0.03% total carbon plus nitrogen

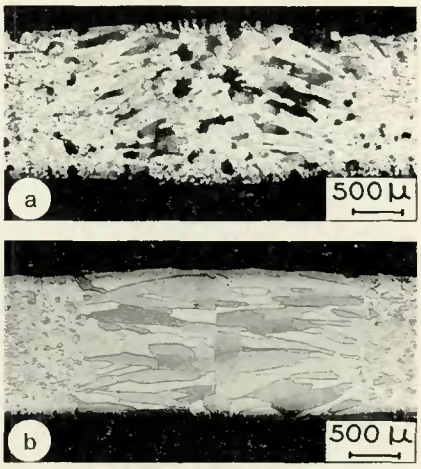


Fig. 3 — Photomacrographs showing a cross section of welds in 0.055 in. (1.4 mm) thick sheet; X20, Hydrochloric-nitric-glycerol etch. (a) 18Cr-2Mo-Ti steel (1D); (b) 18Cr-2Mo-Cb steel (2B)

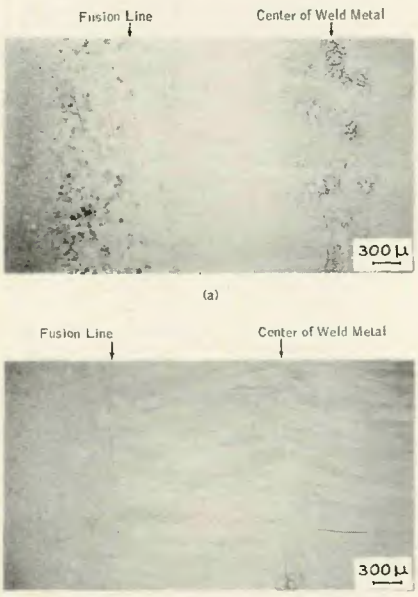


Fig. 4 — Photomacrographs showing the top surface of welds in 0.005 in. (1.4 mm) thick sheet; X30, 50% HNO_3 electrolytic etch. (a) 18Cr-2Mo-Ti steel (1A); (b) 18Cr-2Mo-Cb steel (2A)

in the 0.055 in. (1.4 mm) thick steels. All weld metals solidified from teardrop shaped puddles. Slower welding speeds that would have promoted an elliptically shaped puddle were not used.

Precipitates and other particles observed in the weld metals are shown in Fig. 5. The weld metals exhibited large ferrite grains with no evidence of martensite. Carbonitride precipitates were distributed throughout the grains and aligned at the grain boundaries. Compared with the base metal or heat-affected zone, the weld metal precipitates were more numerous and smaller in size. The weld metals also contained small spherical oxide inclusions and, in the case of weld metals with higher columbium contents, elongated inclusions at the grain boundaries as shown in Figs. 6a and 6b. Silicon was detected in the inclusion to the left of Fig. 6a and aluminum, iron, chromium, and sulfur were detected in the inclusion shown in Fig. 6b. In the dark region of Fig. 6b, a significant amount of carbon was present, however, in other areas of this particle and in other particles, carbon was not present; therefore, it can be assumed that the large grain boundary inclusions were predominantly complex oxides and sulfides. Smaller cube-shaped particles were also observed at the grain boundaries of the columbium stabilized weld metals. Areas containing a large concentration of these particles at grain boundaries are shown in Figs. 6c and 6d.

Extraction replicas taken from the titanium stabilized steels contained small unidentified particles less than 100 angstroms in diameter as well as larger cuboid Ti(C,N) precipitates. Annealing reduced the number of small precipitates and those observed were predominantly large particles that were identified by electron diffraction as Ti(C,N) (Fig. 7a). The columbium stabilized weld metals contained cubic precipitates identified by electron diffraction as Cb(C,N) (Fig. 7b). These precipitates were aligned in rows and assumed to be identical in composition to the cubic particles observed optically at the grain boundaries (Fig. 6d).

Fracture surfaces of welds that had exhibited good Olsen cup performance appeared ductile as shown in Fig. 8a. Fracture surfaces of high titanium welds that had performed poorly in the Olsen cup test exhibited classic brittle cleavage characteristics. It was noted previously that the welds with higher columbium contents usually fractured transverse to the welding direction in the Olsen cup test. Figure 8b shows a transverse fracture which is intersected by a longitudinal fracture. Note that the large facets are aligned on preferred cleavage planes. Optical metallography revealed that fractures in the high columbium welds were predominantly transgranular.

Other transverse weld metal fractures from Olsen cup tests of the high columbium steels are shown in Fig. 9. These Olsen cup fractures exhibited low ductility. A smooth area in Fig. 9a is shown at higher magnification in Fig. 9b. A high magnification view of another transverse weld metal fracture is also shown in Fig. 9c. These areas appear to be grain or subgrain boundaries and contain particles or impressions of extracted particles that in some cases are cube shaped (Fig. 9c) and similar in appearance to the Cb(C,N) particles identified by electron diffraction (Fig. 7b).

Discussion

Comparison of Olsen Cup and Bend Tests

The results of the bend tests (Table 7) show a loss in weld ductility at a high columbium level as was determined by the Olsen cup tests (Table 6); however, the bend tests of the 0.050 in. (1.3 mm) specimens exhibited lower ductility only at the highest level of columbium tested (0.64% Cb). The Olsen cup test, therefore, was more sensitive to differences in alloy content that cause a loss in weld ductility. It should also be noted that the longitudinal bend specimens of the columbium stabilized steels were more susceptible to cracking than the transverse specimens. The Olsen cup test, by comparison, imposed an equal tensile stress in all directions and, consequently, in one test, forced the specimen to fail in the weakest direction. Because of these advantages, the Olsen cup test was considered to give the best measure of weld ductility during slow loading.

Comparison of Base Metal and Weld Ductility

Base metal Olsen cup results (Table 5) were not significantly affected by composition. However, the ductility of the welds in the as-deposited condition was greatly influenced by the composition (Table 6). The ductility of the titanium alloyed welds that were given a postweld anneal approached that of the base metals; however, the welds with higher columbium contents exhibited lower ductility than the comparable base metals even after a post-weld anneal (Tables 5 and 6).

Titanium Stabilized Welds

Titanium promoted the formation of equiaxed grains in preference to the normal columnar type during the solidification of the 18Cr-2Mo weld

metals. This effect of titanium has also been observed in duplex (ferrite-austenite) stainless alloys (Ref. 9) and low alloy steels (Ref. 10). It has been suggested that stable titanium and aluminum nitrides form in the liquid metal and provide nucleation sites for the growth of individual dendrites ahead of the advancing solid-liquid interface (Ref. 9). Steel 1D (Fig. 3a), which contained high titanium (0.78%) and carbon (0.029%) contents, exhibited the greatest proportion of equiaxed grains. In contrast to previous work with duplex alloys (Ref. 9), the addition of aluminum above 0.05% did not significantly increase the number of equiaxed grains in the titanium stabilized weld metals.

The weld ductility as evaluated by the Olsen cup test was lowered at the highest titanium levels. In 26% Cr alloys, Pollard (Ref. 11) found that increasing the titanium content from 6 to 17 times the total carbon plus nitrogen caused a marked decrease in the weld ductility during tensile testing (to 0% reduction of area). Wright (Ref. 5) observed in 26%Cr-1%Mo alloys that the effect of titanium on weld ductility depended on the interstitial content. The results of this investigation (Fig. 6) generally support the trends observed in the literature, although the 18Cr-2Mo steels in this investigation with Ti/(C + N) ratios up to about 17 (1B and 1D) exhibited reasonably good ductility in the Olsen cup test (Table 6).

The low ductility in the weld that contained the highest titanium content cannot be explained by the presence of any precipitates or phases observed in the weld. It seems likely, therefore, that the loss in ductility was caused by precipitation strengthening from submicroscopic titanium-rich particles that had precipitated during the rapid thermal cycle of the weld. A similar embrittlement mechanism for 26% Cr alloys over stabilized with titanium has been suggested by Pollard (Ref. 12). A postweld anneal that promoted the growth of more coarse Ti(C,N) particles (Fig. 7c) caused an improvement in the ductility of the high titanium welds.

A carbon addition to the high titanium steel resulted in a significant increase in the weld ductility (Fig. 1) indicating that the titanium and interstitial content must be balanced to obtain good as-welded ductility. In the range of compositions investigated (Table 1), reasonably good ductility was obtained in the Olsen cup test [e.g., greater than 0.3 in. (7.6 mm) Olsen cup depth] with titanium contents less than 17 times the total carbon plus nitrogen and excess titanium less than 0.47%. This level corresponds to about 0.60% Ti for a steel containing 0.03% total carbon plus nitrogen.

The impact toughness (Table 8) was more sensitive to titanium additions, but good impact toughness [e.g., a 7.5 ft-lb (10.2 J) transition temperature below room temperature] could be obtained with about 0.03% total carbon plus nitrogen and less than about 0.50% Ti. The weld with a higher carbon content (1D) exhibited good Olsen cup performance but

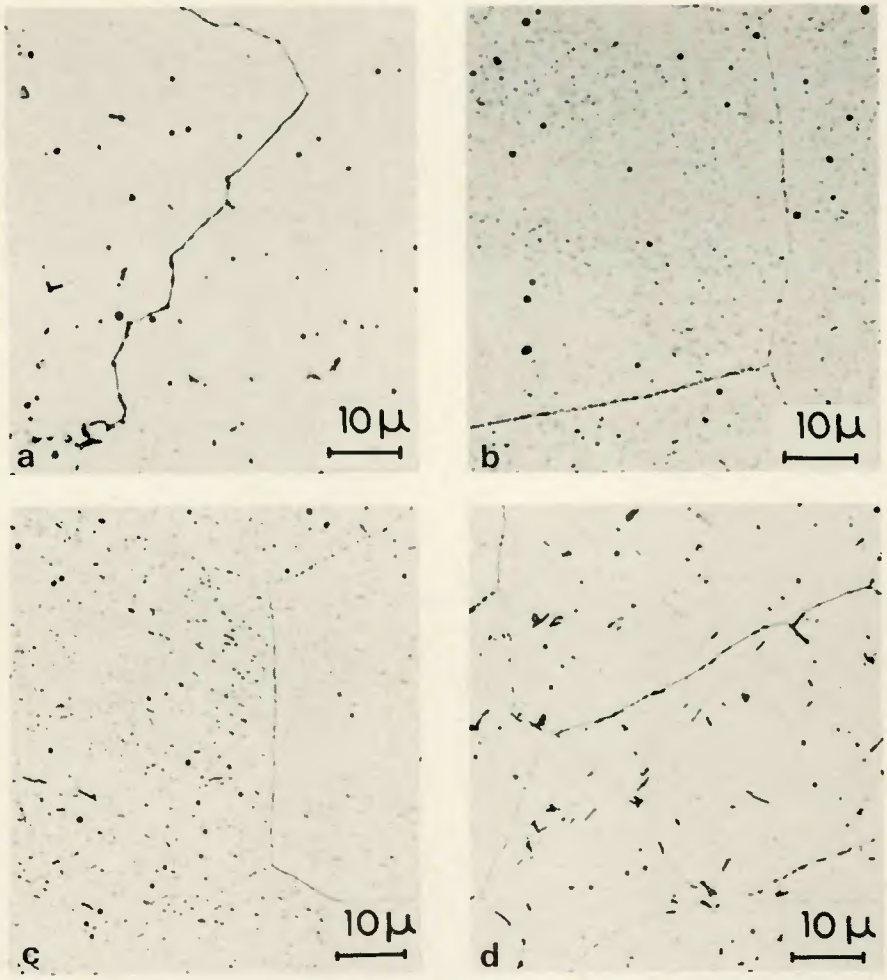


Fig. 5 — Weld metal precipitates in 18Cr-2Mo steels; X1000, 50% HNO_3 electrolytic etch. (a) 18Cr-2Mo-Ti steel (1A); (b) 18Cr-2Mo-Cb steel (2A); (c) 18Cr-2Mo-Cb steel (2C); (d) 18Cr-2Mo-Ti-Cb steel (4C)

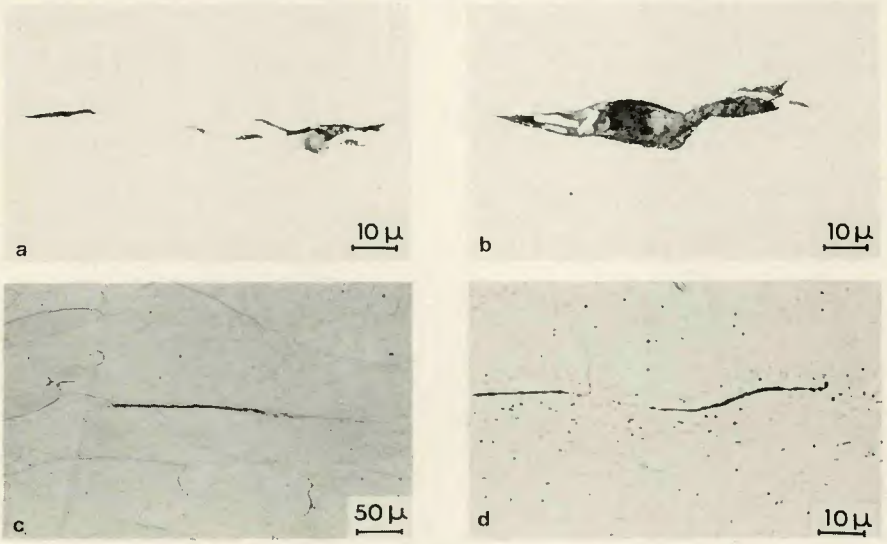


Fig. 6 — Grain boundary particles in the weld metal of an 18Cr-2Mo-Cb steel (2B). (a) and (b) X1000, as polished; (c) X100, 50% HNO_3 electrolytic etch; (d) X1000, 50% HNO_3 electrolytic etch

relatively poor impact toughness. This result can be explained partly by noting that a higher concentration of Ti(C,N) particles larger than about 1 micron were observed in the weld metal of Steel 1D. Particles of this size can lower the impact energy absorption but have little effect at slow strain rates.

Columbium Stabilized Welds

The columbium stabilized weld metals which did not contain titanium exhibited a solidification structure characteristic of epitaxial growth as shown in Figs. 3b and 4b. This structure is normal for weld metals that solidify from a tear-drop shaped puddle (Ref. 13). If a weld metal chemistry is susceptible to hot cracking, then cracking is likely to occur when solidification occurs by this mechanism. In the case of weld metals containing higher levels of columbium, there was some evidence of grain boundary micro hot cracking (Figs. 6 and 9), and the presence of these defects probably contributed to the poor Olsen cup performance of the high columbium welds. A high concentration of cuboid Cb(C,N) precipitates at the grain boundaries (Fig. 6d) may have also promoted the embrittlement of these welds. Brittle cup fractures exhibited grain or subgrain boundary surfaces (Fig. 9) that may have contained these particles or a low melting columbium-rich phase.

Precipitation strengthening by an $(Fe,Cr)_2Cb$ Laves phase is possible in Fe-Cr-Cb alloys (Ref. 14), and the Fe-Cb phase diagram (Ref. 15) indicates that Fe_2Cb can precipitate from alpha iron containing less than 1% Cb. However, the 1800 F (980 C) anneal after welding would have overaged a weld strengthened by $(FeCr)_2Cb$ precipitation (Ref. 14). The low ductility even after a postweld anneal (Table 8) indicates that precipitation strengthening by a Laves phase did not play a major role in the high columbium welds.

An increase in the carbon content to 0.03% had little effect on the ductility of the high columbium weld (Fig. 2); however, the 0.03% C weld had 0.28% excess columbium — a level that would have caused ductility problems at the lower carbon level (Table 6).

Columbium stabilized weld metals with higher aluminum contents exhibited extremely good toughness but the impact fracture surfaces and Olsen cup fractures indicated some weakness transverse to the impact fractures. The lower resistance to crack propagation perpendicular to the welding direction did not affect the impact energy absorption because the impact fracture propagated in a direction parallel to the welding direction. All of the titanium free steels solidified by epitaxial growth from a tear-drop shaped puddle, and, thus, developed a preferred orientation in the fusion zone by the mechanism reported by Savage and Aronson (Ref. 16). Since the preferred growth direction in body centered cubic materials is the <100> direction, the titanium-free weld metals developed a preferred orientation in which the <100> direction was transverse to the welding direction. Body centered cubic metals fracture by cleavage on {100} planes (Ref. 17); consequently, the titanium-free weld metals exhibited large cleavage facets on {100} planes perpendicular to the welding direction (Fig. 8b).

Combined Columbium, Titanium and Aluminum Additions

The addition of titanium to a columbium stabilized weld metal resulted in an improvement in the ductility. The titanium addition also caused the weld metal to solidify with equiaxed grains in the center and eliminated micro hot cracking. The impact toughness of the welds with combined titanium, columbium and aluminum additions was relatively low (Table 8). The rows of large precipitates shown in Fig. 5d were probably detrimental to impact toughness but had little effect on the ductility at slow strain rates. The same effect was observed in the weld with a high titanium and carbon content (Steel 1D).

Practical Implications

The columbium content must be controlled closely in 18Cr-2Mo-Cb steels that are welded at high speeds with an autogenous GTA process. Higher levels of columbium can be tolerated if other steps are taken to

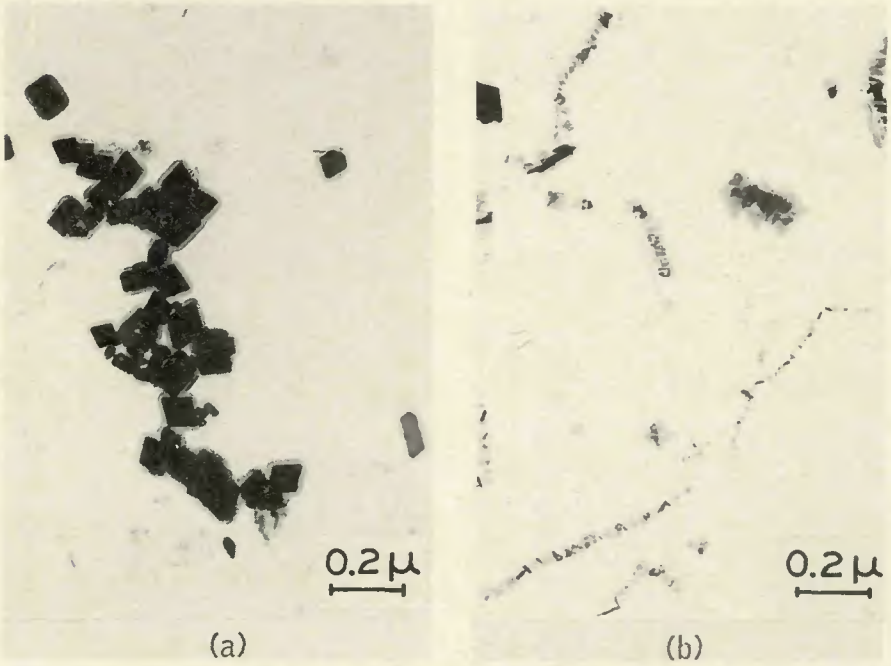


Fig. 7 — Particles observed in extraction replicas of 18Cr-2Mo weld metal, X50,000. (a) Ti(C,N) in weld metal of steel 1C that was annealed after welding; (b) Cb(C,N) in weld metal of steel 2A (as-welded)

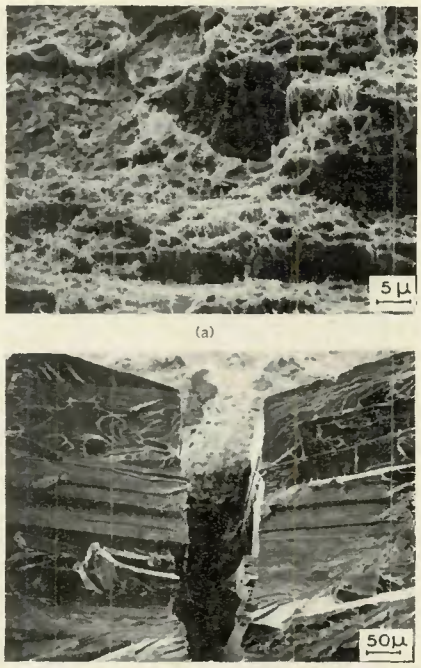


Fig. 8 — Scanning electron micrographs of Olsen cup fractures in 18Cr-2Mo weld metals of (a) steel 1A with 0.37% Ti, X2000, and (b) steel 2D with 0.64% Cb, X200

reduce the tendency for micro hot cracking or if forming operations on the weld are not severe. The tendency for micro hot cracking can be reduced by adding small amounts of other elements that stabilize the low-melting constituents or interact with carbon and nitrogen in preference to columbium (e.g., titanium). A decrease in the welding travel speed could be employed to improve the weld ductility. The slower speed would produce an elliptically shaped weld puddle and thereby decrease the degree of preferred orientation. A slower travel speed might also lower the degree of constitutional supercooling (Ref. 14) and decrease the tendency for micro hot cracking.

A broader range of titanium contents can be tolerated in the 18Cr-2Mo-Ti steels. Since the titanium stabilized welds were not affected by micro hot cracking, a change in the welding speed would not be expected to influence the ductility. Moreover, welds with higher titanium contents can be annealed after welding to improve the ductility.

Conclusions

1. With a proper level of stabilizing element, both titanium and columbium stabilized 18Cr-2Mo steels containing about 0.03% total carbon plus nitrogen exhibited excellent ductility and toughness.

2. The titanium stabilized steels exhibited good weld ductility in the Olsen cup test over a broader range of stabilizing element concentration than the columbium stabilized steels.

3. Weld metals with less than about 0.50% Ti and with all columbium levels investigated exhibited 7.5 ft-lb (10.2 J) transition temperatures that were below room temperature in quarter-size Charpy V-notch tests.

4. The ductility of the titanium stabilized weld metals was affected by the relative titanium and interstitial levels, but the impact transition temperature was increased by an increase in either the titanium or interstitial content.

5. Weld metals containing columbium, titanium, and higher aluminum contents exhibited good ductility during cup testing but relatively high impact transition temperatures.

6. An increase in the aluminum content to 0.07% caused a marked improvement in the impact toughness of a columbium stabilized weld, but slight reduction in the ductility as measured by the Olsen cup test.

7. A postweld anneal improved the ductility of the high titanium welds but had only a small effect on the high columbium welds.

8. Titanium promoted the solidification of equiaxed grains in the center of the weld fusion zone, whereas the titanium-free weld metals solidified completely by epitaxial growth and developed a preferred <100> orientation transverse to the welding direction.

9. High columbium contents caused micro hot cracking at the grain boundaries which initiated brittle fracture during Olsen cup testing.

10. The Olsen cup test proved to be an effective test for evaluating the weld ductility.

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Fig. 9 — Scanning electron micrographs of transverse weld metal fractures in Olsen cup specimens of 18Cr-2Mo-Cb steels, (a) steel 2C, X200; (b) steel 2C, X1000; (c) steel 2B, X2000