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Notch Toughness of Low Oxygen Content Submerged Arc Deposits

The use of Al, (RE), Mn, Si and Mg for deoxidizing welds below 220 ppm oxygen impairs notch toughness; Ti and V promote acicular ferrite microstructures and improve toughness in deposits with 200 ppm oxygen

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ABSTRACT. Factors affecting the notch toughness properties of CMn and CMnCb deposits containing < 250 ppm oxygen have been examined.

If Al, (RE), Mn, Si or Mg are employed to deoxidize single pass deposits below 200 ppm oxygen, notch toughness properties are impaired. If Zr is used, deposits containing 100 ppm oxygen can be produced without an impairment of notch toughness properties. If Ti and V are employed, acicular ferrite formation is promoted and low oxygen content deposits can be produced with an improvement in notch toughness properties.

The effect of a deoxidant such as (RE) on notch toughness is cooling-rate dependent. At high cooling rates (RE) additions are detrimental. At low cooling rates (RE) additions are beneficial.

Aluminum additions in multipass CMn deposits (containing 200 ppm oxygen) are beneficial. Aluminum additions in CMnCb deposits of similar composition are detrimental to notch toughness properties.

Introduction

The oxygen content of submerged arc deposits has a considerable bearing on the microstructure and notch toughness properties. It is well known that high oxide inclusion contents are associated with decreased notch toughness properties. Recent research has suggested a relationship between oxide inclusion type and/or size distribution and deposit microstructure. For example, deposits with oxygen contents around 600 ppm produced using manganese silicate-type fluxes exhibit predominantly ferrite sideplate microstructures.[1] However, deposits produced using low oxygen potential fluxes containing around 300 ppm oxygen exhibit microstructures corresponding to Levine's Class I designation, viz., grain boundary ferrite and acicular ferrite.[2]

Different explanations have been put forward to explain the oxygen (oxide inclusion)-deposit microstructure relationship, viz., oxide inclusion nucleation of ferrite plates[3-6] and oxide inclusion control of austenite grain size which affects the microstructure of high oxygen deposits.[7] The type of inclusions present—not the total inclusion content—is a key feature, since alumina-based fluxes can deposit welds containing 600 ppm oxygen and exhibit acicular ferrite microstructures. Inclusion size distribution *per se* is not a major effect in determining deposit microstructure since agglomerated and fused fluxes depositing welds of similar composition have comparable inclusion size distributions but quite different microstructures.[8]

Electron beam welds and laser-remelted submerged arc deposits containing approximately 110 ppm oxygen preferentially form bainitic microstructures rather than acicular ferrite.[6] However, in conventional submerged arc welding, the deposition of welds with less than 200 ppm oxygen can only be achieved using low oxygen potential fluxes (such as 45 wt-%CaF_2-35 wt-%Al_2O_3-20 wt-%CaO[9] or by adding deoxidants.[9-11] Consequently, the role of the deoxidant itself must also be taken into account when considering the factors determining the microstructure and toughness of low oxygen content deposits.

Consider the addition of a deoxidant (M) producing low oxygen content submerged arc deposits. It is apparent

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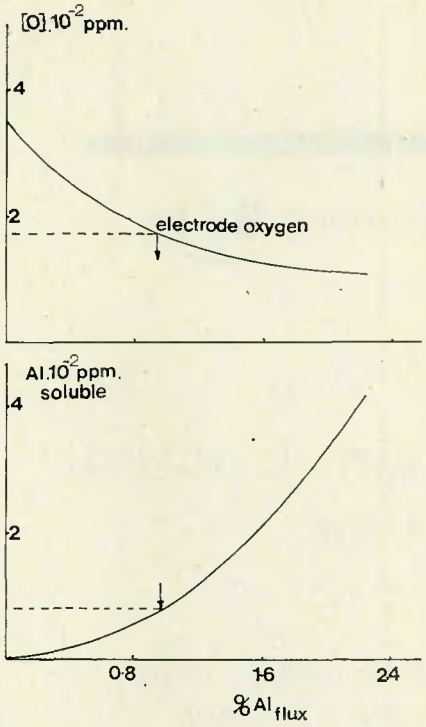


Fig. 1—Deoxidation during submerged arc welding using deoxidant (M). In this case (M) is aluminum

Table 1—Plate and Electrode Compositions Employed, %

Designation	C	Mn	Si	S	P	Cb	Ti	[O]
Plate P1	0.18	1.29	0.28	0.026	0.016	–	–	–
Plate P2	0.055	0.48	0.05	0.026	0.005	–	0.10	0.007
Plate P3	0.22	1.37	0.23	0.012	0.012	0.034	–	–
Plate P4	0.20	1.21	0.36	0.007	0.009	0.030	–	0.009
Electrode E1	0.11	2.10	0.23	0.03	0.03	–	–	0.017
Electrode E2	0.12	1.55	0.13	0.016	–	–	–	0.0125
Electrode E3	0.11	1.98	0.23	0.02	0.018	–	–	0.017
Electrode S1	0.08	0.47	0.18	0.012	0.002	–	–	0.029
Electrode S2	0.10	0.98	0.17	0.008	0.002	–	–	0.013
Electrode S3	0.12	1.59	0.29	0.012	0.012	–	–	0.008

from Fig. 1 that lowering the deposit oxygen content below that of the electrode will result in increased contents of deoxidant in the final deposit. The form this excess deoxidant will take and its influence on deposit microstructure and notch toughness will depend on the characteristics of the deoxidant considered.

If a stable deoxidant carbide is not formed, excess deoxidant will be in solution (e.g., soluble aluminum in excess of oxide and nitride requirements, or soluble (RE) in excess of oxide, oxysulphide and sulphide requirements). In the case of titanium, excess deoxidant is soluble Ti, TiN and Ti(CN). Soluble aluminum contents in wrought steels and in weld deposits impair notch toughness.[9] Nitride and/or carbonitride precipitation can affect deposit microstructure, yield strength and notch toughness properties. It follows that, in low oxygen content deposits, the deoxidant itself will have an important effect on deposit microstructure and toughness independently of any effect it has on deposit oxygen content.

The influence of aluminum and of (RE) additions during submerged arc welding has been examined. Aluminum additions in CMn deposits promote microstructural changes (decreased pro-eutectoid ferrite and increased (M-A) phase contents) and are beneficial to stress-relieved toughness properties when added up to a critical level.[9,10] In the case of (RE) additions, the amounts of (M-A) constituents and sideplate ferrite are increased, and as-welded toughness is impaired.[11] An important side effect of aluminum additions in CMnCb and in CMnCbV deposits is its effect in increasing the rate of precipitation of Cb(CN) during stress-relieving (with consequent impairment of notch toughness properties). In this connection large alumina contents in CaF_2-Al_2O_3-CaO flux formulations are associated with the poorest as-welded and stress-relieved notch toughness properties.[10]

Aims of Research Program

A number of lines of investigation were followed:

1. An examination was made of the effects of aluminum additions in multipass CMn and CMnCb deposits. Variables such as welding parameters, electrode and base material compositions, and flux formulation were kept constant. Columbium was added via the flux for effective comparison between CMn and CMnCb deposits.

In multipass deposits the notch toughness depends on the relative properties of refined and as-deposited weld materials. Since aluminum is a well-established grain-refining addition in wrought steels, the influence of aluminum on normalized and tempered deposits was examined.

2. An investigation was carried out on the effects of deoxidant additions on the microstructure and toughness of low oxygen content welds produced at widely different cooling rates. In these tests deposit composition was held constant, and the effects of (RE) additions were monitored.

3. The microstructures of deposits containing 250 to 100 ppm oxygen were examined. Electron beam welds and laser remelted submerged arc deposits having low oxygen contents preferentially form bainitic microstructures.[6] Since acicular ferrite is a desirable constituent from the point of view of notch toughness, it is important to evaluate whether this occurs in conventional submerged arc welding. In these tests low oxygen content deposits were produced using a low oxygen potential flux formulation (45 wt-%CaF_2-35 wt-%Al_2O_3-20 wt-%CaO) and by adding a range of deoxidants during welding.

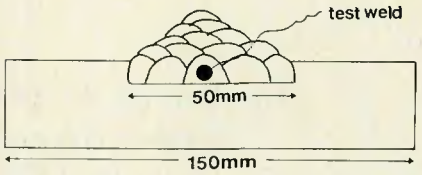


Fig. 2—Setup for multipass welding

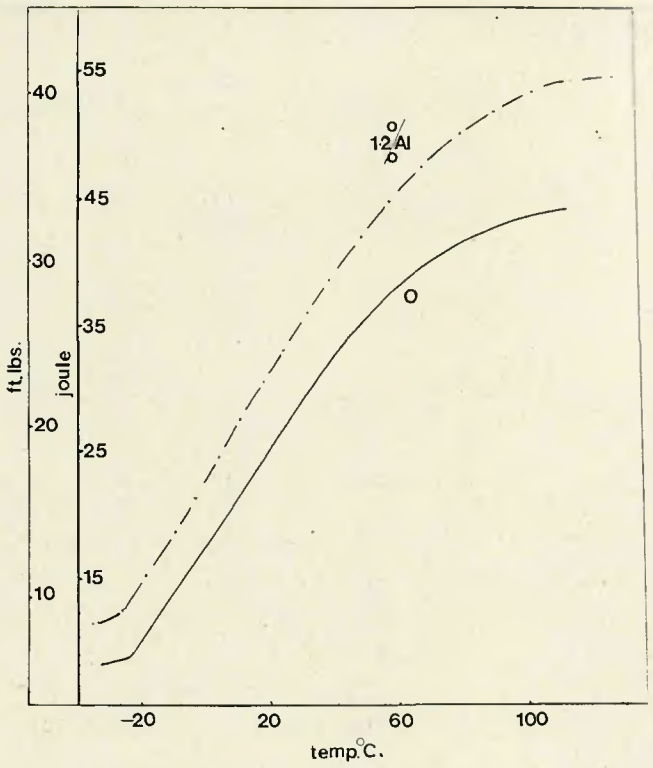
4. The effects of aluminum (RE) and titanium additions on the as-welded and stress-relieved toughness properties of CMnCb deposits containing < 200 ppm oxygen were compared.

Experimental Procedures

Materials

The chemical compositions of plate and electrode materials are given in Table 1.

A number of laboratory prepared fluxes were employed (10 wt-%CaF_2-45 wt-%Al_2O_3-45 wt-%CaO) flux was used in some normalized and tempered tests. A flux comprised of 45 wt-%CaF_2-35 wt-%Al_2O_3-20 wt-%CaO was used in multipass tests, and in almost all tests aluminum, (RE), titanium, zirconium and magnesium were added during welding. The only exception to this occurred when the effects of (RE) additions in alumina-bearing and in alumina-free fluxes were compared. The fluxes employed in this comparison were 45 wt-%CaF_2-35 wt-%Al_2O_3-20 wt-%CaO and



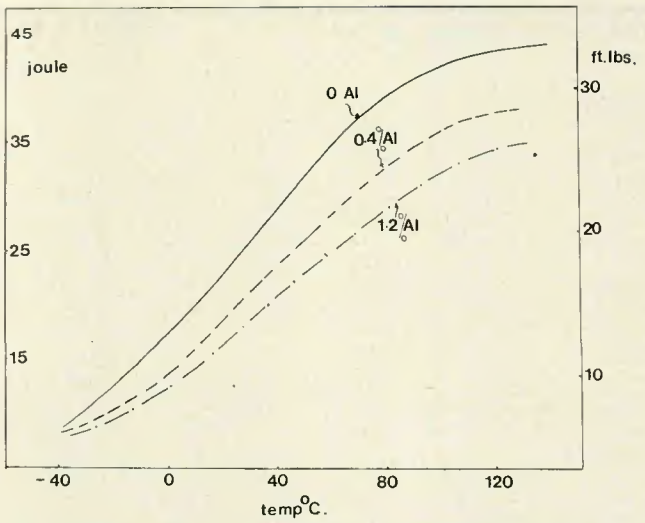


Fig. 4—Effect of aluminum additions on the notch toughness of multipass CMnCb deposits: —— 0% Al addition in flux; – – – 0.4 wt-% Al in flux; —·—·— 1.2 wt-% Al in flux

Fig. 3 (left)—Effect of aluminum additions on the notch toughness of multipass CMn deposits: —— 0% Al additions in flux; —·—·— 1.2 wt-% Al addition in flux

80 wt-%CaF_2-20 wt-%CaO respectively.

Aluminum and titanium were added by mixing pure metal powders with fluxes prior to welding. Mixing was carried out in a rotating drum. (RE) additions were made using ferro-(RE) powder. Ferro-(RE) powder was prepared by melting 60 parts of electrolytic iron with 40 parts of mischmetal in an argon atmosphere. This material comprised approximately 18-20% cerium.

In multipass CMn and CMnCb deposits, columbium additions were made by adding 0.2 wt-% ferro-columbium via the 45 wt-%CaF_2-35 wt-%Al_2O_3-20 wt-%CaO flux. In this way the effect of aluminum additions in CMn and CMnCb deposits of comparable deposit compositions could be compared easily.

Welding Conditions

Multipass CMn and CMnCb deposits were produced as shown in Fig. 2. A 5 × 50 mm (0.197 × 1.97 in.) groove was cut in 25 mm (1 in.) thick mild steel plate (P1). The electrode employed was (E1). Welding conditions were 600A, 32V, 250 mm/min (9.8 ipm). The test weld was deposited along the centerline of the groove, and other weld beads were deposited around it. In this way CMn and CMnCb deposits of similar compositions could be subjected to the same thermal cycle of multipass welding.

Normalized and tempered tests were made on the final pass of ten-high pads. The pads were deposited by laying four beads, followed by three, then by two, and finally by the test weld. Proprietary electrodes with different manganese levels were employed (0.5%, 1.0% and 1.5%Mn). The normalizing and tempering treatment comprised 2 hours (h) at 925 C (1697 F), followed by 2 h at 650 C (1202 F), and then by air-cooling.

The influence of cooling rate on the role of (RE) additions in low oxygen content deposits was examined as follows. Bead-on-plate deposits were made on 20 mm (0.79 in.) thick mild steel plate (P2) using electrode (E2). These deposits had a cooling time of 110 s for the 800-500 C (1472-932 F) temperature range. When the same specimens were chilled prior to welding using solid CO_2, cooling times of 50 seconds (s) were achieved. In these tests 3 wt-% of pure manganese powder was added to the flux in order to raise the final deposit manganese content to 1.3%. The welding conditions were 600A, 32V, 250 mm/min (9.8 ipm).

Tests comparing the effects of aluminum and titanium additions on notch toughness properties were carried out using single-pass deposits. In the tests, 7 mm (0.276 in.) deep 90 deg V notches were cut in 30 mm (1.22 in.) thick CMnCb plate (P3). Electrode (E3) was used (see Table 1). Both as-welded and stress-relieved conditions were examined—stress-relief being carried out at 600 C (1112 F) for 1 h.

In tests where (RE) additions were made to CMnCb deposits, the plate employed was (P3) and the electrode used was (E2). Bead-on-plate welds were employed in this case.

Two types of Charpy specimens were employed in this work. Small 8 mm (0.315 in.) diameter × 45 mm (1.77 in.) long specimens in normalized and tempered multipass, and in single-pass welds compared the effects of cooling rate on the role of (RE) additions, and in (RE) tests on CMnCb deposits. Also, 10 mm (0.394 in.) × 10 mm (0.394 in.) specimens were used in single-pass tests comparing aluminum and titanium additions in CMnCb deposits. Longitudinal Charpy specimens were employed in normalized and tempered tests, and in tests comparing the role of (RE) additions in welds made at different cooling rates. In all other tests, transverse Charpy specimens were used and had notches cut in the direction parallel to the welding axis.

Gas tungsten arc remelt tests were carried out as follows. Submerged arc deposits free of cerium and with 0.03% cerium were deposited using electrode (E2) and plate (P2). Welding conditions were 620A, 32V, 250 mm/min (9.8 ipm). Specimens for remelting 50 × 75 mm (1.97 × 2.95 in.) were cut from the submerged arc welds and were set up on a roller-bearing system which facilitated workpiece movement during welding. Gas tungsten arc remelts were made in a specially constructed chamber which was evacuated and back-filled with argon prior to welding. In these tests the tungsten electrode remained stationary, and the weld metal cooling rate was modified by changing the traverse

Table 2—Deposit Analyses in Tests, %

Designation		C	Mn	Si	S	P	Cb	Al (flux)	Al deposit	Oxy-gen	Ce	Ti
		0.14	1.15	0.16	0.007	0.015	–	–	0.008	0.023	–	–
Multipass		0.18	1.23	0.21	0.006	0.014	–	1.20	0.017	0.017	–	–
deposits		0.18	1.05	0.16	0.007	0.014	0.028	–	0.011	0.023	–	–
		0.16	1.09	0.19	0.009	0.013	0.027	0.4	0.015	0.018	–	–
		0.18	1.12	0.21	0.006	0.016	0.029	1.20	0.031	0.0175	–	–
(RE)	a	0.08	1.42	0.11	0.011	0.007	–	–	–	0.015	–	–
Additions	b	0.085	1.35	0.11	0.010	0.007	–	–	–	0.011	0.015	–
and	c	0.09	1.33	0.12	0.009	0.006	–	–	–	0.010	0.021	–
cooling	r	0.08	1.36	0.11	0.009	0.006	–	–	–	0.018	–	–
rate.	s	0.085	1.30	0.12	0.008	0.007	–	–	–	0.012	0.028	–
	t	0.09	1.38	0.13	0.007	0.008	–	–	–	0.012	0.035	–
Normalized	S1	0.09	0.39	0.11	0.01	0.01	–	–	0.006	0.043	–	–
and	S1	0.09	0.38	0.10	0.010	0.014	–	0.8	0.029	0.030	–	–
tempered	S2	0.09	0.41	0.05	0.005	0.015	–	–	0.007	0.032	–	–
deposits	S2	0.15	0.49	0.11	0.004	0.012	–	0.4	0.022	0.030	–	–
	S3	0.105	1.04	0.18	0.012	0.015	–	–	–	0.022	–	–
	S3	0.11	1.07	0.26	0.008	0.016	–	0.4	–	0.019	–	–
CMnCb		0.195	1.42	0.15	0.011	0.010	0.020	–	0.0125	0.016	–	–
deposits		0.175	1.46	0.15	0.010	0.009	0.021	0.40	0.014	0.011	–	–
		0.17	1.48	0.21	0.005	0.009	0.024	–	0.012	0.012	–	0.011
	AW	0.19	1.28	0.19	0.008	0.009	0.024	–	0.037 soluble	0.013	–	–
	AW	0.20	1.32	0.21	0.007	0.009	0.021	–	0.039 soluble	0.011	0.01	–
	AW	0.20	1.34	0.22	0.007	0.010	0.023	–	0.048 soluble	0.009	0.027	–
	AW	0.21	1.35	0.25	0.006	0.008	0.027	–	0.052 soluble	0.009	0.040	–
	SR	0.16	1.33	0.22	0.005	0.013	0.027	–	–	0.014	–	–
	SR	0.18	1.33	0.24	0.005	0.013	0.029	–	–	0.011	0.02	–
Submerged arc deposits		0.08	1.01	0.06	0.011	0.008	–	–	0.021 soluble	0.0195	–	0.035
for remelting		0.08	0.97	0.06	0.009	0.007	–	–	0.028 soluble	0.015	0.03	0.035

speed of the workpiece. Conditions for the GTA remelting were 500A, 10V, 140-190 mm/min (5.51-7.48 ipm). Weld metal cooling rates were measured using a platinum/platinum-rhodium thermocouple.

Water-quenching experiments were carried out by welding plates in a specially-constructed tray. Immediately after arc termination, the tray was flooded with water from a reservoir. Great care was taken to keep each test condition comparable with prior tests.

Chemical Analysis

Oxygen analysis was carried out using a Balzer Exalograph EA1 machine which gave a reproducibility of ± 0.001% on any deposit. Aluminum analysis was carried out using the eriochrome-cyanine method. Cerium analysis was carried out using an ion-exchange analysis method and had an accuracy of ± 0.002%.

Results

Multipass CMn and CMnCb Deposits Containing Aluminum

The effect of aluminum additions on the notch toughness of CMn and CMnCb deposits are presented in Figs. 3 and 4. It is apparent that aluminum additions are detrimental in the case of CMnCb deposits and are beneficial in the case of CMn deposits. The compositions of these deposits are given in Table 2.

The notch toughness of multipass deposits depends on the relative properties of unrefined and refined regions of weld deposits. In CMn deposits increasing aluminum contents are associated with decreased deposit oxygen content (from 230 to 180 ppm), with lower inclusion contents and with finer-grained ferrite and pearlite microstructures in reheated regions of weld beads. An addition of 1.2% aluminum to the flux decreased the refined zone grain size from 13.5μ to 9.8μ.

In CMn deposits aluminum had no effect on the proof-stress values of transverse tensile specimens. Other investigators have noted increased proof-stress values in aluminum-treated multipass submerged arc deposits,[12] in GMA-CO_2 deposits,[13] and in shielded metal arc deposits.[14]

The beneficial effects of aluminum additions in multipass CMn deposits are similarly present in normalized and tempered weld deposits—Fig. 5. The effectiveness of aluminum additions depends on the manganese content. Both upper shelf and transition temperature values improve at higher manganese contents. In low manganese content deposits (containing 0.38-0.42% Mn) upper shelf values are improved, but there is only a slightly beneficial effect on transition temperature. In deposits containing 0.85-0.87% Mn, aluminum additions are only slightly beneficial. In deposits containing 1.04 to 1.23% Mn, only transition temperature values are improved.

Aluminum additions in multipass CMnCb deposits are detrimental to both upper shelf and transition temperature values—Fig. 4.

Cooling Rate and (RE) Additions

Figure 6 shows the influence of (RE) additions on the as-welded toughness of single-pass CMn deposits produced at different cooling rates. Cooling times between 800 C (1472 F) and 500 C (932 F) were 50 and 110 s, respectively. It is clear that (RE) additions are beneficial in 110 s cooling time tests and are detrimental in lower cooling

time tests. An interesting feature of Fig. 6 is that the presence of large (RE) contents in deposits makes the notch toughness properties independent of cooling rate. Table 2 shows that comparable deposit analyses were produced at both cooling rate conditions.

The beneficial effect of (RE) additions in 110 s cooling time tests is due to grain refinement, large primary ferrite grains being replaced by polygonal ferrite surrounding a finer ferrite network—Fig. 7.

In tests where the cooling time between 800 C (1472 F) and 500 C (932 F) is 50 s, (RE) additions are detrimental to as-welded notch toughness properties. This results from a small increase in the sideplate ferrite content of deposits (from 2 to 8%). Also, (RE) additions reduce alumina from the molten slag and, consequently, part of the detrimental effect is due to increased contents of aluminum in solution.

(RE) promotion of sideplate ferrite formation during welding was confirmed in water-quenching tests. Figure 8 shows that, at identical positions behind the rear of the weld pool, (RE)-bearing deposits exhibit appreciable contents of ferrite sideplates and (RE)-free deposits do not.

The relationship between microstructure, cooling rate and (RE) content in CMn deposits was clarified over a wide range of conditions using gas tungsten arc remelting of submerged arc deposits. At any cooling time between 7 and 140 s for the 800–500 C (1472–932 F) temperature range, (RE) additions refined deposit microstructure. (For equivalent (RE) contents, decreased cooling times progressively refined deposit microstructure.)

In all GTA remelt tests, deposit oxygen content decreased to a constant level of 90 ppm. Where no deoxidant was added to the flux, oxygen content fell from 180 ppm to 90 ppm; this decrease was due to removal of oxide inclusions trapped in the original submerged arc welds. Characteristic microstructures of submerged arc welds deposited at any cooling rate could not be observed in remelt tests carried out at the same cooling rate. In fact, the 7 s cooling time remelt test had an identical microstructure to a 45 s cooling time submerged arc weld. However, the welding parameters during remelting were different from those in submerged arc welding and, consequently, different weld pool dimensions and fusion zone solidification microstructures were produced. As a result, it is not possible to wholly ascribe the microstructural differences between equivalent cooling rate remelt and submerged arc deposits to the marked fall in oxygen content during remelting.

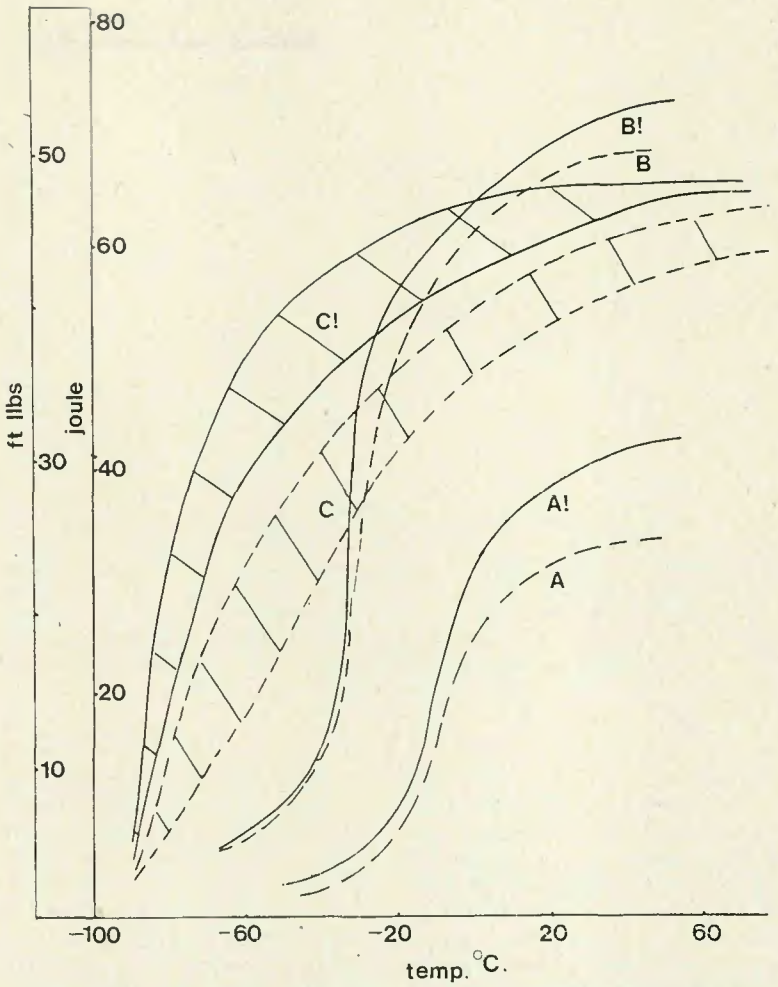


Fig. 5—Effect of aluminum additions on the notch toughness of normalized and tempered CMn deposits (after Nowicki[25] and Chakravarty[24]): A—0% Al addition in 10 wt-% CaF_2-45 wt-% Al_2O_3-45 wt-% CaO flux, S1 electrode wire; A!—0.8 wt-% Al addition in 10 wt-% CaF_2-45 wt-% Al_2O_3-45 wt-% CaO flux, S1 electrode wire; B—0% Al addition in 45 wt-% CaF_2-35 wt-% Al_2O_3-20 wt-% CaO flux, S2 electrode wire; B!—0.4 wt-% Al in 45 wt-% CaF_2-35 wt-% Al_2O_3-20 wt-% CaO flux, S2 electrode wire; C—0 wt-% Al addition in 45 wt-% CaF_2-35 wt-% Al_2O_3-20 wt-% CaO flux, S3 electrode wire; C!—0.4 wt-% Al in 45 wt-% CaF_2-35 wt-% Al_2O_3-20 wt-% CaO flux, S3 electrode wire

Microstructures of Deposits Having Oxygen Contents Between 250-100 ppm.

Deposits Free of Deoxidant Additions. Figure 9 shows typical microstructures of the final runs of ten-high pad deposits. Apart from manganese content differences (1.15% *cf.* 1.48%), deposit compositions are similar and were deposited with identical welding conditions. Figure 9A comprises pro-eutectoid ferrite, sideplate ferrite and a small content of acicular ferrite. However, Fig. 9B exhibits higher acicular ferrite contents. In effect, the higher manganese content of Fig. 9A restricts acicular ferrite formation. Choi and Hill noted a similar effect when manganese was increased to high levels in CMn deposits.[5]

Deposits Containing Deoxidants. The presence of titanium in CMn deposits containing 180 ppm oxygen promoted formation of a microstructure wholly comprising acicular ferrite and pro-eutectoid ferrite—Fig. 10A. Such an effect has been noted previously.[3] The employment of other deoxidants such as aluminum, (RE)—and even silicon and magnesium[15]—to obtain such low oxygen levels was not accompanied by such a microstructure.

In CMnCb deposits titanium additions also produced a microstructure comprising pro-eutectoid ferrite and acicular ferrite. However, aluminum and (RE) additions did not modify the characteristic Type II microstructure of CMnCb deposits,[2] i.e., parallel ferrite

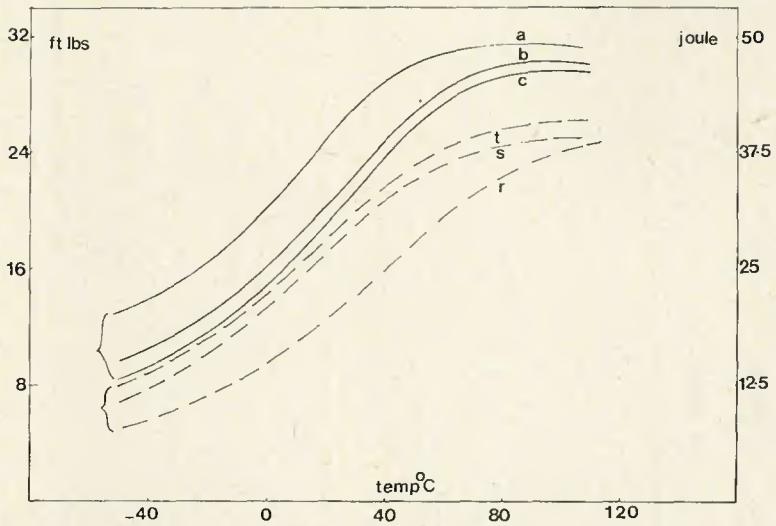


Fig. 6—Effect of (RE) additions on the notch toughness of CMn deposits subjected to different cooling times for the 800-500 C (1472-932 F) temperature range (percentages are wt-%): a—0% (RE) addition in flux, 50 s cooling time; b—0.015% Ce in deposit, 50 s cooling time; c—0.021% Ce in deposit, 50 s cooling time; r—0% Ce in flux, 110 s cooling time; s—0.028% Ce in deposit, 110 s cooling time; t—0.035% Ce in deposit, 110 s cooling time

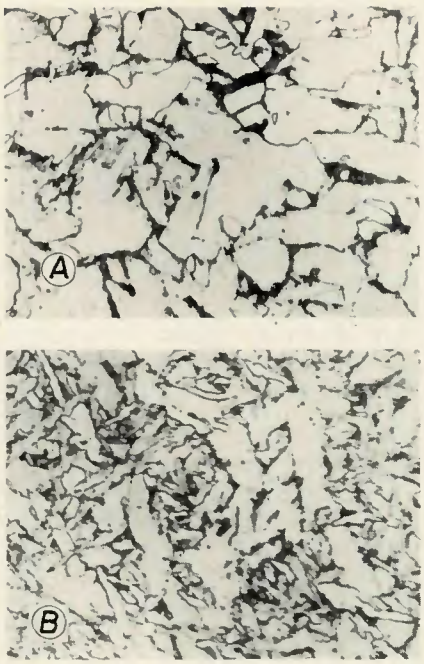


Fig. 7—Microsections of 110 s cooling time-CMn deposit: A—no (RE) addition; B—0.035 wt-% Ce in deposit. ×500 (reduced 50% on reproduction)

laths emanating from prior austenite grain boundaries—Figs. 10B and 10C.

Ito and Nakanishi[3] have suggested that the promotion of acicular ferrite microstructures in titanium-bearing deposits is due to TiN particles acting as nuclei for growth of ferrite plates. However, aluminum has a strong affinity for nitrogen, and low oxygen content CMnCb deposits do not exhibit acicular ferrite microstructures. Also, open-arc, flux-cored electrodes containing large aluminum contents exhibit bainitic rather than acicular ferrite microstructures.

Since it has been suggested that TiN formation modifies the deposit microstructure, it is of considerable interest to examine the influence of other nitride-forming additions, e.g., zirconium and vanadium. Zirconium has a considerable affinity for nitrogen, and additions of up to 1.4% in 45 wt-%CaF_2-35 wt-%Al_2O_3-20 wt-%CaO flux have no observable effect on the typical microstructure of CMnCb deposits. However, vanadium additions in CMnCb deposits containing < 200 ppm oxygen promote acicular ferrite formation—Fig. 10D. In effect, titanium and vanadium promote acicular ferrite microstructures, but zirconium and aluminum do not.

Notch Toughness of Deoxidized CMnCb Deposits Containing Aluminum, (RE) and Titanium

Figures 11 to 14 compare the effects of aluminum, (RE) and titanium on the notch toughness of CMnCb deposits. The deposit compositions are given in Table 2.

Only titanium lowered the deposit oxygen content to 100 ppm with an improvement in as-welded toughness properties. Aluminum additions were detrimental to notch toughness properties. In the case of (RE), concentrations greater than approximately 0.01 wt-% (RE) impaired the as-welded toughness properties.

The notch toughness of stress-relieved welds was reduced by the addition of aluminum, (RE) and titanium. In particular, the notch toughness

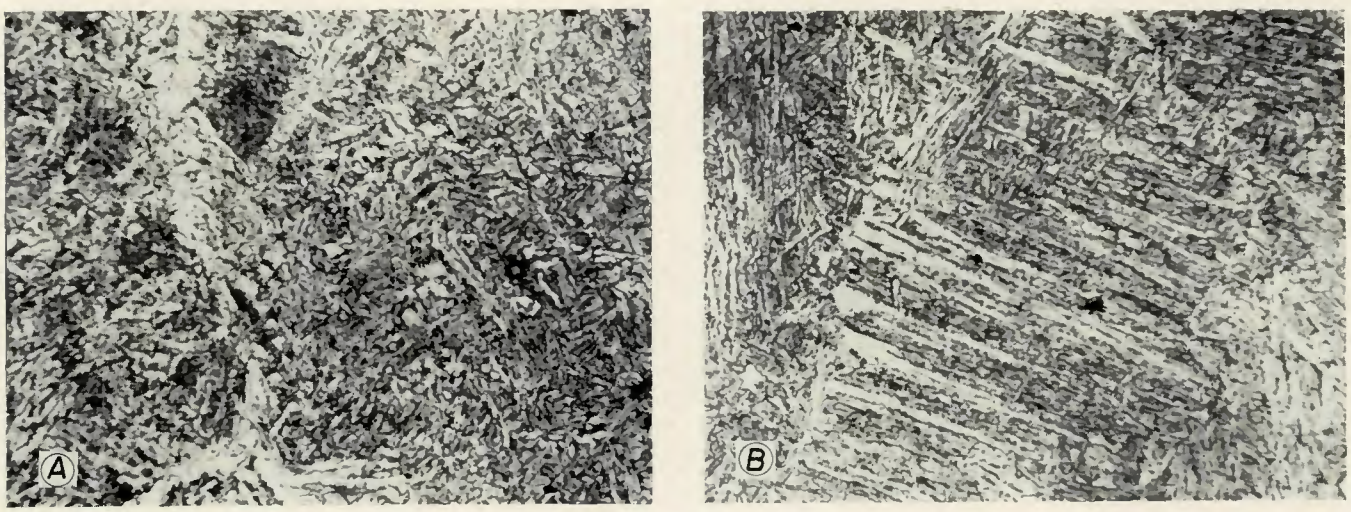


Fig. 8—Microsections of CMn submerged arc weld after water quenching—40 mm (1.57 in.) behind the weld pool: A—no (RE) addition; B—0.02 wt-% Ce in deposit. ×500 (reduced 50% on reproduction)

of titanium-bearing deposits was severely impaired by stress-relieving.

The tensile strength of titanium-bearing deposits was markedly increased on stress-relieving and was symptomatic of precipitation hardening. (RE) additions had a negligible effect on both as-welded and stress-relieved tensile strength properties. Aluminum additions had little effect on the as-welded tensile properties but did increase stress-relieved tensile strength by increasing the rate of precipitation of Cb(CN) particles.

Discussion

Multipass Welds

Aluminum additions were beneficial in CMn deposits and were detrimental in CMnCb welds. In multipass welds, toughness depends on the structure of regions reheated by subsequent runs and on the as-deposited structure. It has been shown that addition of up to 1.2 wt-% aluminum in the flux has little influence on as-welded toughness and is beneficial to the stress-relieved toughness properties of single-pass CMn deposits.[9] Also, although no clear relation between soluble nitrogen content and aluminum additions could be found in CMn deposits, chemical analysis showed appreciable insoluble nitride contents.[10]

As a result, some part of the beneficial effect of aluminum on transition temperature of CMn deposits may result from decreased soluble nitrogen contents. The finer grain size of reheated zones of aluminum-bearing CMn deposits would also promote improved transition temperature values. The lower inclusion content of multipass deposits accounts for the beneficial effect of aluminum on upper Charpy shelf values.

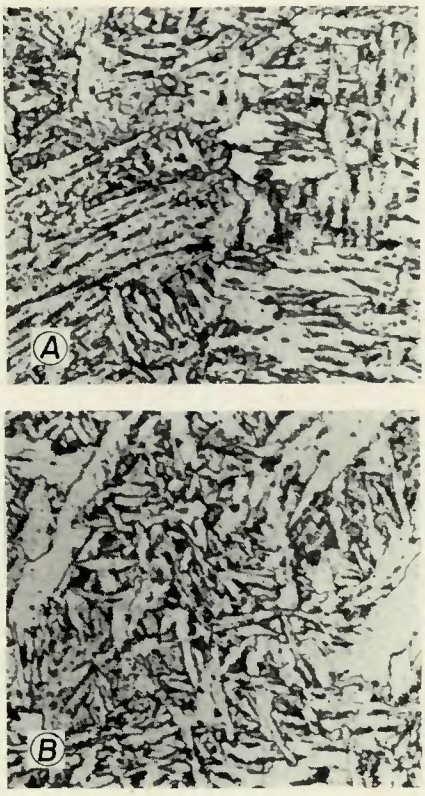


Fig. 9—Microsections of CMn deposits with varying compositions: A—0.13% C, 1.48% Mn, 0.017% S, 0.009% P, 0.0218% oxygen; B—0.13% C, 1.15% Mn, 0.08% Si, 0.012% S, 0.011% P, 0.026% oxygen. ×500 (reduced 50% on reproduction)

The detrimental effect of aluminum in CMnCb deposits occurs since the beneficial finer grain size of reheated regions is outweighed by detrimental effects on other weld bead areas. Such effects include promotion of (M-A) phase formation, and increased rates of Cb(CN) precipitation during the thermal cycle of multipass welding.[10]

The beneficial role of aluminum on the notch toughness of normalized and tempered CMn deposits results from lower oxygen and inclusion contents (producing improved upper shelf values); it also results from finer austenite grain sizes during normalizing (producing improved transition temperature values).[28,29]

Cooling Rate and (RE) Additions in Low Oxygen Content Deposits

The changing role of (RE) additions with change in cooling rate is not unexpected. It is well established that a change of deposit composition may be beneficial or detrimental to notch toughness properties, depending on the particular deposit composition, cooling rate and microstructural situation considered.[15] For example, increased manganese content in CMn deposits made at 1.6 to 3.3 kJ/mm (40.60 to 83.8 kJ/in.) is only beneficial up to a certain level; at higher manganese levels, notch toughness is impaired.

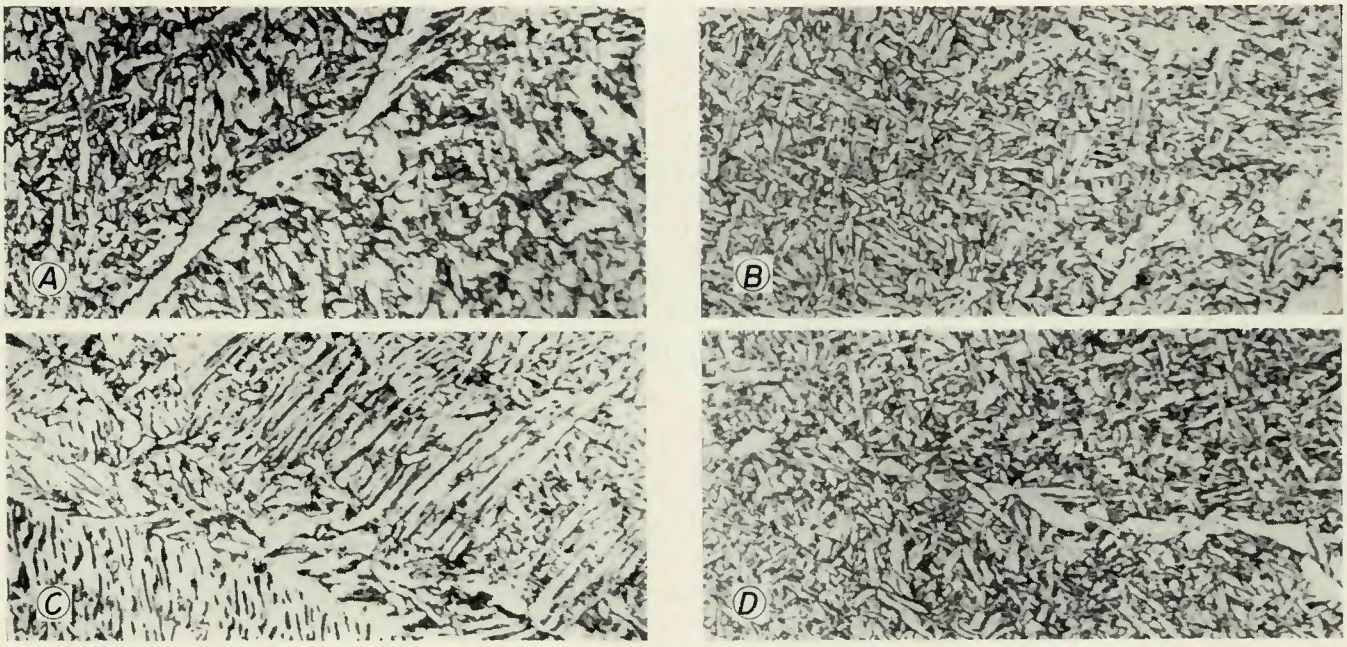


Fig. 10—CMn/CMnCb deposit microsections: A—titanium-bearing 0.08%C, 0.11% Si, 1.42% Mn, 0.011% S, 0.007% P, 0.035% Ti, 0.018% oxygen; B—titanium-bearing 0.17% C, 0.21% Si, 1.48% Mn, 0.005% S, 0.009% P, 0.024% Cb, 0.011% Ti, 0.012% oxygen; C—0.195% C, 0.15% Si, 1.42% Mn, 0.011% S, 0.010% P, 0.020% Cb, 0.016% oxygen; D—vanadium-bearing 0.17% C, 0.28% Si, 1.28% Mn, 0.006% S, 0.008% P, 0.018% Cb, 0.09% V, 0.010% oxygen. ×500 (reduced 50% on reproduction)

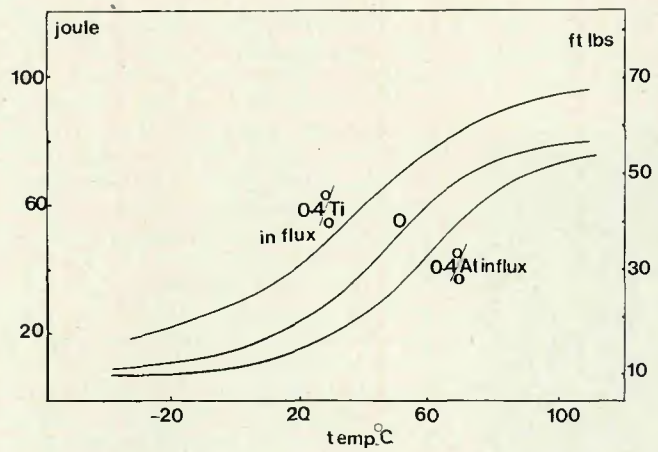


Fig. 11—Effect of aluminum and titanium additions on the as-welded notch toughness of low oxygen content CMnCb deposits

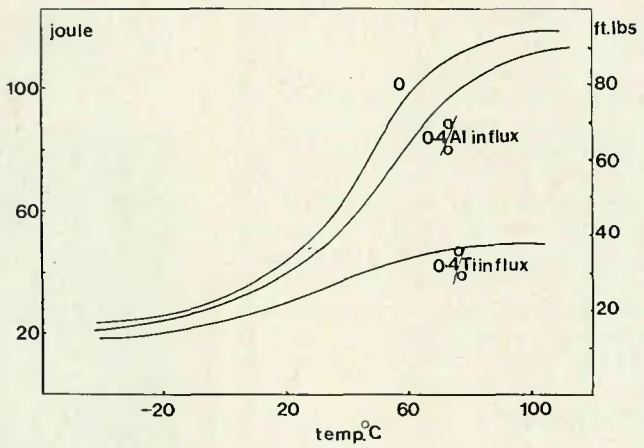


Fig. 12—Effect of aluminum and titanium additions on the stress-relieved notch toughness of low oxygen content CMnCb deposits

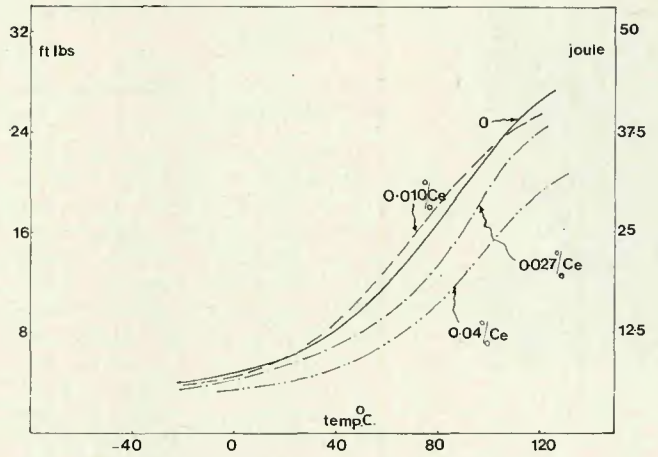


Fig. 13—Effect of (RE) additions on the as-welded notch toughness of low oxygen content CMnCb deposits: ——— 0 wt-% (RE) addition; – – – – – 0.01 wt-% Ce deposit; –·–·– 0.027% wt-% Ce in deposit; –··–··– 0.04 wt-% Ce in deposit

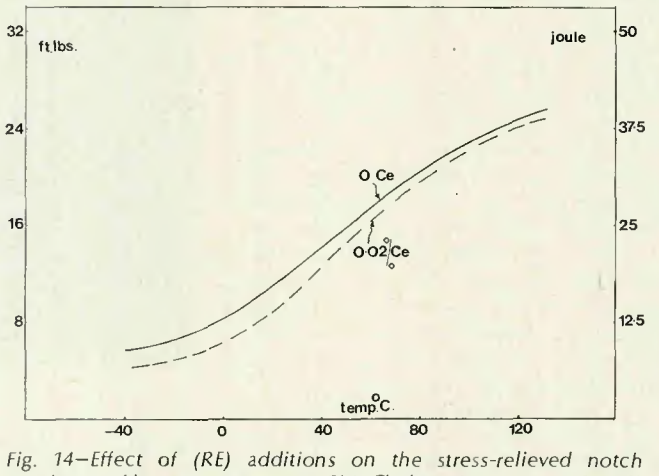


Fig. 14—Effect of (RE) additions on the stress-relieved notch toughness of low oxygen content CMnCb deposits: ——— 0 wt-% (RE) addition; – – – – 0.02 wt-% Ce in deposit

The slow cooling time of 110 s for the 800-500 C (1472-932 F) temperature range is possibly more typical of electroslag welding situations. However, the results emphasize that low oxygen content deposits (< 200 ppm) can be produced with improved toughness properties in some cooling rate situations.

Microstructures of Low Oxygen Content Deposits

Titanium and vanadium promote acicular ferrite formation in CMnCb deposits, but aluminum, (RE) and zirconium do not. It is worth examining how titanium and vanadium promote acicular ferrite formation in low oxygen content deposits. This might be accomplished by:

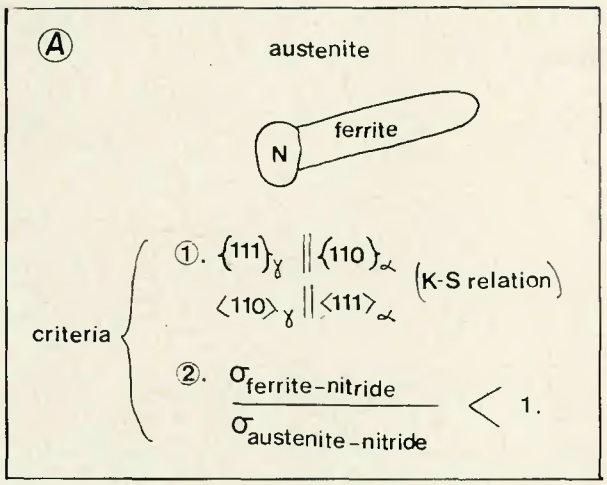
1. Alteration of the oxide type and/or inclusion size distribution so that prior austenite grain size or intragranular nucleation of ferrite laths is affected, or
2. TiN and VN affecting austenite grain size or intragranular nucleation of ferrite plates.

The role of oxide inclusions acting as nuclei within grains for acicular ferrite formation has been emphasized.[3-6] Since (RE) additions have a much greater affinity for oxygen than titanium, these were made to examine their effect on the microstructure of titanium-bearing deposits. Acicular ferrite formation was not suppressed. It follows that the role of titanium in promoting acicular ferrite was not due to alteration of oxide inclusion parameters.

Zirconium has a greater affinity for nitrogen than titanium (or vanadium). When zirconium was added to titanium and vanadium-bearing deposits, the acicular ferrite content was markedly decreased. In titanium-bearing deposits, it was decreased from approximately 70% to less than 20%. In vanadium-bearing deposits, acicular ferrite formation was almost completely prevented. Since zirconium only slightly increases hardenability in steels containing columbium,[17] these results suggest that zirconium affects acicular ferrite formation by interfering with TiN and VN formation.

Measurement of prior austenite grain size using SASPA-NANSA showed that CMnCb deposits containing deoxidants had finer grain size than deposits made with fluxes free of deoxidants, e.g., 53μ in titanium-bearing and 62μ in deoxidant-free fluxes. CMnCb deposits containing Ti, V, Zr and (RE) had comparable prior austenite grain sizes. Also, those containing V and (V + Zr) had equivalent prior austenite grain sizes. In effect, titanium and vanadium-bearing welds did not exhibit acicular ferrite microstructures as a direct result of changes in prior austenite grain size.

However, the finer prior austenite grain size of titanium and vanadium-bearing deposits was indicative to TiN and VN particles acting at prior austenite grain boundaries. The presence of such particles together with finer prior austenite grain size promote ferrite formation. Since the typical ferrite sideplate microstructure of Cb-bearing



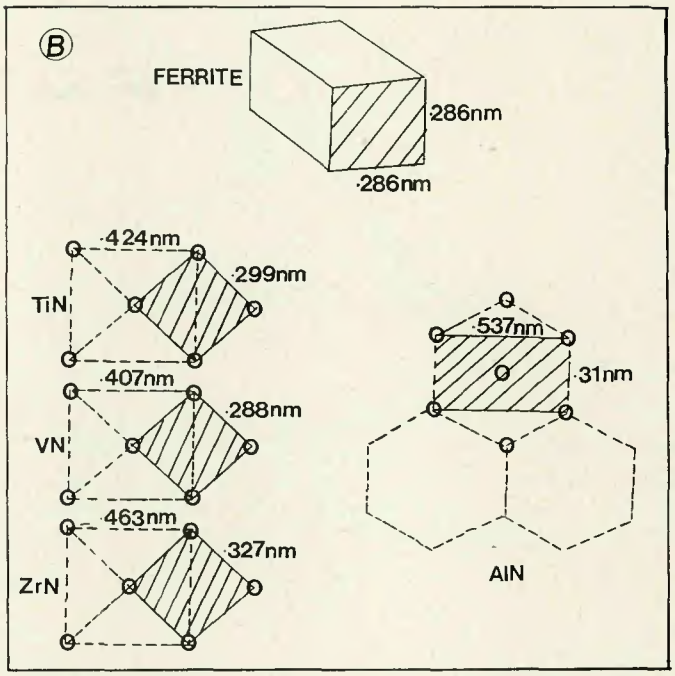


Fig. 15—Nucleation of ferrite plates on intragranular nitride particles in austenite. A—schematic representation; B—lattice dimensions

deposits was suppressed when titanium and vanadium were present, such additions must modify the γ/α transformation in some way so that ferrite sideplate formation was inhibited (or replaced by a more energetically favorable means of ferrite lath formation).

Ito and Nakanishi[3] have pointed out that both TiN and ZrN particles could promote acicular ferrite formation; this is because they were insoluble in austenite and provided intragranular nucleation sites during the austenite/ferrite transformation. However, in the present work zirconium additions did not promote acicular ferrite microstructures in CMnCb deposits.

Both TiN and ZrN may be formed as cuboid particles in steel melts, and are virtually insoluble in austenite.[18,19] On the other hand AlN and VN cannot precipitate at 1600 C (in the concentrations normally encountered) and have only limited solubility in austenite.[19] Since TiN, VN, ZrN and AlN particles have been confirmed in submerged arc deposits (using internal friction measurements[3] and by chemical analysis techniques[10,20]), one might speculate why TiN and VN particles might promote intragranular ferrite plate formation and AlN and ZrN do not.

The principal driving forces for growth of a ferrite crystal from austenite comprise the difference in chemical free energy between α and γ, and the tendency for the system to minimize the total interfacial free energy of interface boundaries.[21] Figure 15A illustrates ferrite plate nucleation from nitride particles in austenite. Such ferrite plate formation will be promoted if two requirements are fulfilled:

1. The ferrite plates must take up the Kurdjumov-Sachs orientation relationship with the parent austenite, i.e.,

$$(111)_\gamma \;//\; (110)_\alpha$$

$$< 110 >_\gamma \;//\; < 111 >_\alpha$$

(growth direction)

This matching of close-packed planes in f.c.c. and b.c.c. lattices provides the lowest degree of atomic misfit.

2. The interfacial energy of the ferrite/nitride boundary must be less than the interfacial energy of the austenite/nitride interface:

$$\frac{\sigma_\alpha\text{-nitride}}{\sigma_\gamma\text{-nitride}} < 1$$

Close matching of important planes of the ferrite and nitride lattices would promote this effect.

The energy of the austenite-nitride interface will depend markedly on the mode of formation of such particles. TiN particles may comprise those formed in the melt prior to solidification, and those precipitated from the austenite phase. In the case of TiN particles formed prior to solidification, the austenite/nitride interface will be incoherent. However, since vanadium nitride is also important in acicular ferrite formation, the factors determining nitride precipitation from austenite must be considered.

The crystal structures and lattice parameters of nitride particles are:

- TiN cubic. NaCl type; a = 0.424 nm.
- VN cubic. NaCl type; a = 0.407 nm.
- ZrN cubic. NaCl type; a = 0.463 nm.
- AlN close-packed-hexagonal; a = 0.31 nm, c = 0.497 nm

Cubic titanium, vanadium and zirconium carbides and nitrides precipitate in austenite and in ferrite as plates on { 100 } planes.[22,23] Such planes are favorable since the bulk modulus is low in $< 100 >$ directions.[22,24] As these precipitates grow in size, the interface between austenite and nitride particles will become incoherent. If coherent precipitates formed in austenite, the energy of the austenite/precipitate interface would be low and this would make the $(\sigma_{\alpha\text{-nitride}}/\sigma_{\gamma\text{-nitride}}) < 1$ criterion more difficult to achieve.

The two interfacial energy conditions for ferrite plate nucleation on nitride particles can be easily satisfied if the { 100 } planes of ferrite and nitride particles match—Fig. 15B. In the case of TiN and VN close matching between the 100 planes of ferrite and nitrides occurs and the mismatch parameter (δ = difference of nitride and ferrite lattice parameters/ferrite lattice parameter) is 0.0045 (TiN) and 0.007 (VN).

In the case of ZrN particles, the conditions for matching between {100} planes of nitride and ferrite lattices are not so favorable and δ is 0.143. For $\delta < 0.05$ the interface is coherent, and for $0.05 < \delta < 0.25$ a semi-coherent boundary is produced.[24] Since the interfacial energy of a coherent interface is much less than that of a semi-coherent interface, TiN and VN would provide more energetically favorable sites for ferrite lath nucleation than ZrN.

In the case of AlN particles, close matching between the {100}, {110 and {111} planes of ferrite and important planes in the c.p.h., AlN lattice

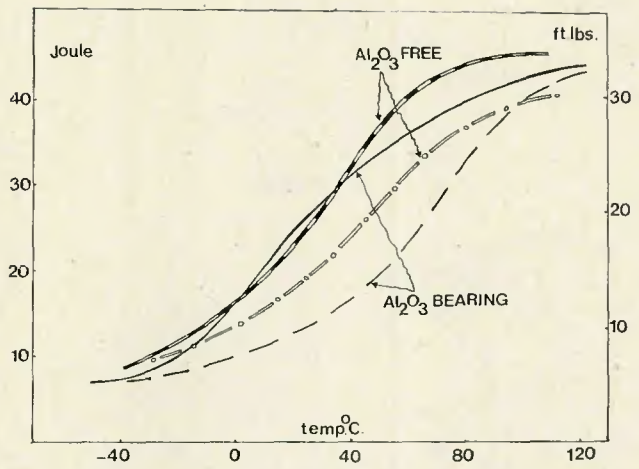


Fig. 16—Effect of (RE) additions on the as-welded notch toughness of CMnCb deposits made using Al_2O_3-free, and Al_2O_3-bearing fluxes. Plate composition (P4). Electrode (E2)

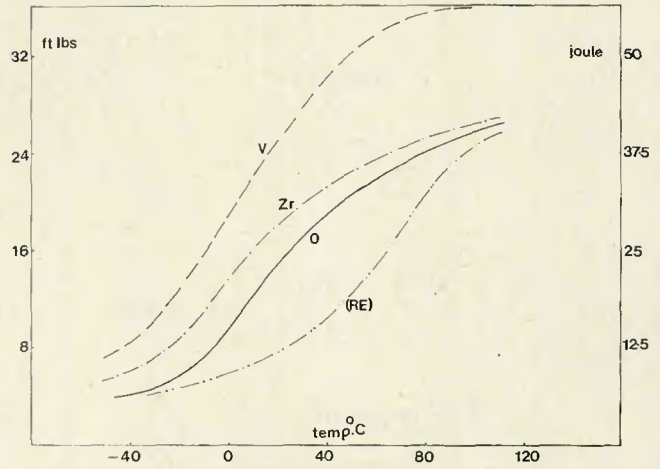


Fig. 17—Effect of zirconium, vanadium and (RE) additions on the as-welded notch toughness of CMnCb deposits. Electrode (E2) employed. 1 wt-% Zr, 1 wt-% FeV and 5 wt-% FeCe in flux. Plate composition P4

({0001}, {1010} and {1020}) cannot be achieved. As a result AlN particles in austenite would be unlikely sites for intragranular ferrite plate nucleation.

It is well established that increased cooling rates can suppress precipitation. In submerged arc deposits the fine oxide inclusion dispersion might act as sites for TiN and VN precipitation. In effect, ferrite laths could be nucleated by TiN or VN particles or possibly via nitride particles which have precipitated on oxide inclusions. In this connection Funakoshi[23] has observed that boron-nitride particles promote formation of fine ferrite plates in the HAZ of (RE)-treated steels by precipitating on dispersed oxide inclusions (of 0.3-1μ diameter). Such nitride-oxide sites were much more effective in nucleating ferrite plates than similarly dispersed oxide inclusions in boron-free steels containing rare-earths.

Notch Toughness of CMnCb Deposits Containing Aluminum, (RE) and Titanium.

As-Welded Properties. Only titanium deoxidized CMnCb deposits to 100 ppm oxygen with an improvement in notch toughness. Both aluminum and (RE) additions impaired toughness properties.

The beneficial role of titanium was partly related to the promotion of an acicular ferrite and pro-eutectoid ferrite microstructure. Such a microstructure has superior notch toughness properties to the characteristic microstructure of CMnCb deposits (parallel ferrite plates emanating from prior austenite grain boundaries[2]). In this connection Jesseman[20] observed that titanium additions in deposits containing 0.05% columbium promoted acicular ferrite formation and improved toughness. Jesseman's deposits were produced using a flux having a much higher oxygen potential than in the present work; consequently, his results apply to a much higher deposit oxygen regime.

In the case of aluminum and (RE) additions, no such beneficial microstructural change was observed, and notch toughness properties were impaired due to increased (M-A) phase and soluble aluminum contents in deposits. In (RE)-treated deposits, increased soluble aluminum contents resulted from alumina reduction (from the flux). Figure 16 clearly shows that the detrimental influence of similar (RE) additions in alumina-free and in alumina-bearing fluxes was greater when fluxes contained alumina.

Decreased soluble nitrogen contents in submerged arc deposits improve notch toughness properties. Ito and Nakanishi[3] showed that titanium reduced soluble nitrogen levels in submerged arc deposits and partly accounted for improved toughness properties. Although Craig, North and Bell[10] could find no clear relationship between soluble nitrogen content and aluminum additions, chemical analysis confirmed appreciable insoluble nitrogen contents in deposits.

It follows that the detrimental influence of aluminum in CMnCb deposits (Fig. 11) must result from factors which outweigh any beneficial effect on soluble nitrogen content. These factors were increased (M-A) phase and soluble aluminum contents. It is important to note that, although zirconium additions did not promote acicular ferrite microstructures, they were slightly beneficial to the as-welded toughness properties of low oxygen content CMnCb deposits—Fig. 17. This was possibly due to zirconium affecting the soluble nitrogen content.[15] However, in Fig. 17 it is important to point out that the acicular ferrite structure of vanadium-bearing deposits had superior as-welded toughness properties compared to the zirconium-treated welds.

Stress-Relieved Properties. Titanium-bearing deposits were most impaired by stress-relieving. The increased tensile strength of stress-relieved deposits was symptomatic of precipitation hardening.

In aluminum-bearing deposits the increased rate of Cb(CN) precititation during stress-relieving accounted for the impaired notch toughness properties.[10]

Commentary. The notch toughness properties of CMn and CMnCb deposits having oxygen contents < 200 ppm will depend on how the low oxygen level is attained, and on the cooling rate after welding. This is clearly illustrated in Fig. 18 where the effects of a number of deoxidants (Al, (RE), Ti, Zr and V) on the notch toughness of low oxygen content CMnCb deposits are summarized.

When the cooling time between 800 C (1472 F) and 500 C (932 F) is approximately 50 s or less, the choice of deoxidant is critical. If aluminum and (RE) are employed, the as-welded notch toughness properties of CMnCb deposits may be impaired, and a large facet of this impairment results from the deoxidant itself.

In this connection Ito and Nakanishi[3] noted that deposition of CMn deposits with < 200 ppm oxygen using manganese and silicon deoxidants resulted in an impairment of notch toughness properties. Their tests were carried out on 19 mm (0.75 in.) thick plate at a heat input of 5 to 6 kJ/mm (127-152.4 kJ/in.).

Tests carried out using thicker plate

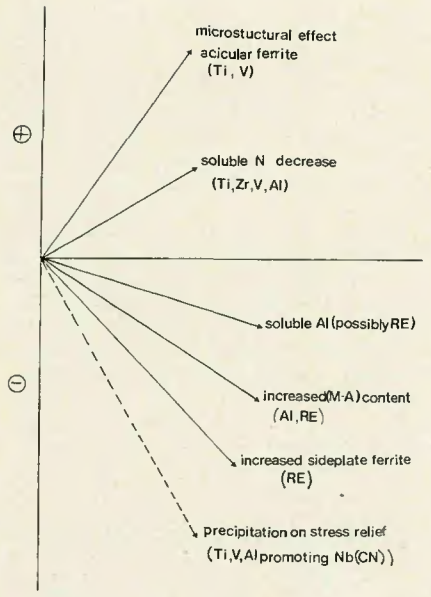


Fig. 18—Comparison of the effects of aluminum, (RE), titanium, vanadium and zirconium on the microstructure and notch toughness properties of low oxygen CMnCb deposits.

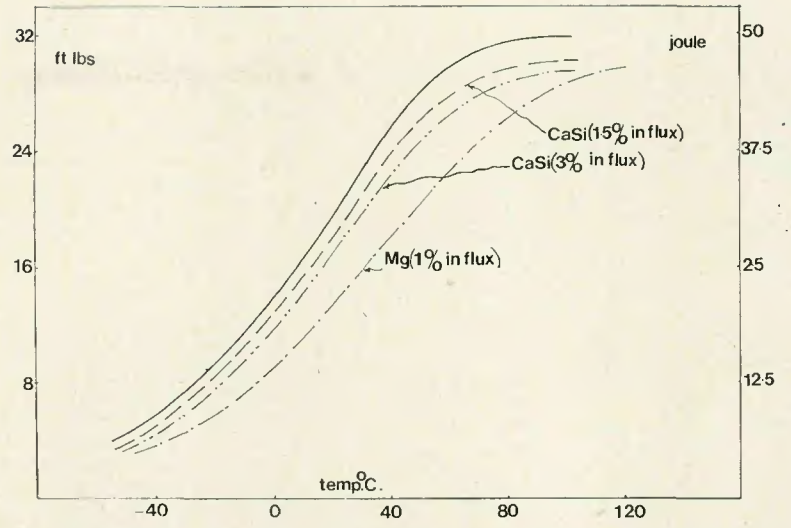


Fig. 19—Effect of magnesium and calcium silicide additions on the as-welded notch toughness of CMn deposits.

with the welding conditions employed in this study (600A, 32V, 250 mm/min (9.8 ipm)) have confirmed Ito and Najanishi's results and have shown that employment of other deoxidants such as magnesium also produce impaired toughness properties.[15,26] In these tests silicon variation was produced by making calcium silicide additions to fluxes, and magnesium was added by employing fluxes containing pure magnesium powder. The detrimental effect of CaSi and Mg additions (Fig. 19) resulted from increased contents of (M-A) phases, and from increased proof and ultimate strength values. Calcium silicide (3 wt-%) additions in fluxes raised the deposit silicon from 0.14 to 0.35% and consequently augmented solid-solution strengthening effects.

In effect, low oxygen content CMn and CMnCb deposits produced by means of Al, (RE), Mn, Si and Mg deoxidation will exhibit inferior notch toughness properties compared to higher oxygen content deposits. (RE), Mn, Si and Mg do not form stable nitrides. In the case of aluminum deoxidation, any beneficial role in decreasing soluble nitrogen content is outweighed by increasing contents of soluble aluminum and (M-A) phases. If the soluble nitrogen content is modified without adverse effects due to excess soluble deoxidant content (as in the case of zirconium additions), notch toughness will not be impaired in deposits containing < 200 ppm oxygen. Both titanium and vanadium additions lower soluble nitrogen content and promote acicular ferrite formation and, consequently, promote the best possible notch toughness in deposits < 200 ppm oxygen.

Proprietary fused fluxes generally do not contain alloy and deoxidant additions. Consequently, deposit composition is largely determined by manganese and silicon deoxidation reactions. When using such fluxes, a lower limit of approximately 200 ppm oxygen will represent a barrier to production of satisfactory notch toughness properties in single-pass deposits free of titanium and/or vanadium. If titanium and/or vanadium are present in single-pass deposits (due to dilution from the plate, from the electrode or from complex deoxidant additions in agglomerated fluxes), acicular ferrite microstructures can be produced and notch toughness may be improved in deposits containing much lower oxygen levels (~ 100 ppm).

In high heat input welding situations where the cooling time between 800 C (1472 F) and 500 C (932 F) is of the order of 110 s, the role of deoxidants in affecting notch toughness may be different from that in lower cooling rate situations. Addition of rare-earths can lower the oxygen content to 100 ppm with an improvement in notch toughness properties. In this case improvement is due to ferrite grain size refinement.

The results in this study apply to an unalloyed deposit situation. The role of additions such as titanium can be complex and may vary depending on the carbon, alloy and columbium content.[27-30] Also, the cooling rate below which an (RE) addition may be beneficial in low oxygen content deposits will depend on the alloy and carbon content (hardenability) of the deposit.

Conclusions

1. Aluminum additions were beneficial in multipass CMn deposits but were detrimental in CMnCb deposits.

2. The effect of (RE) on the notch toughness of CMn deposits depended on the cooling rate after welding. In high cooling rate situations it was detrimental, and in low cooling rate situations it was beneficial.

3. Acicular ferrite microstructures in low oxygen content deposits were promoted by titanium and vanadium additions. Nitride forming additions such as aluminum and zirconium did not promote such microstructures.

4. When aluminum and (RE) additions were employed to produce low oxygen content CMnCb deposits, as-welded notch toughness was impaired. Use of titanium or vanadium improved as-welded toughness properties. The notch toughness of titanium-bearing deposits was most impaired by stress-relieving.

5. When fused fluxes are employed to deposit single-pass welds free of titanium or vanadium, a lower limit of approximately 200 ppm oxygen will represent a barrier to production of satisfactory notch toughness properties. However, if titanium and/or vanadium are present, deposits with much lower oxygen contents (~ 100 ppm) can be produced without impairment of toughness properties.

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Bibliography

1. Cochrane, R. C., and Kirkwood, P. R., "The Effect of Oxygen on Weld Metal Microstructure," International Conf. on Trends in Steels and Consumables for Welding, London, 13-16 November 1978, Paper 35, pp. 103-122.

2. Levine, E., and Hill, D. C., "Structure/Property Relationships in Low C Weld Metal," *Metallurgical Trans.*, 8A(9), 1977, pp. 1453-1457.

3. Ito, Y., and Nakanishi, M., "Study on Charpy Impact Properties of Weld Metal in Submerged Arc Welding," IIW Doc. XII-A-113-75, 1975, pp. 1-50.

4. Shiga, A., Imura, H., and Tsuboi, J., "Effects on Niobium and Vanadium on Toughness of Submerged Arc Weld Metal," IIW Doc. IX-1049-77, XII-A-149-77, 1977, pp. 1-21.

5. Choi, C. L., and Hill, D. C., "A Study of Microstructural Progression in As-Deposited Weld Metal," *Welding Journal*, 57(8), Aug. 1978, Research Suppl., pp. 232-s to 236-s.

6. Abson, D. J., Dolby, R. E., and Hart, P. A. M., "The Role of Nonmetallic Inclusions in Ferrite Nucleation in Carbon Steel Weld Metals," International Conference on Trends in Steels and Consumables for Welding, London, 13-16 November 1978, Paper 25, pp. 75-89.

7. Cochrane, R. C., and Kirkwood, P. R., Corollary to paper (reference 1) presented at International Conference in London, Nov. 1978.

8. Widgery, D. J., "New Ideas in Submerged Arc Welding," International Conference on Trends in Steels and Consumables for Welding, London, 13-16 November 1978, Paper 26, pp. 217-229.

9. North, T. H., Bell, H. B., Nowicki, A., and Craig, I., "Slag/Metal Interaction, Oxygen and Toughness in Submerged Arc Welding," *Welding Journal*, 57 (3), March 1978, Research Suppl., pp. 63-s to 75-s.

10. Craig, I., North, T. H., and Bell, H. B., "Deoxidation during Submerged Arc Welding," International Conference on Trends in Steels and Consumables for Welding, London, 13-16 November 1978, Paper 21, pp. 249-263.

11. Koukabi, A., North, T. H., and Bell, H. B., "Flux Formulation, Sulphur, Oxygen, and Rare-Earth Additions in Submerged Arc Welding," International Conference on Trends in Steels and Consumables for Welding, London, 13-16 November 1978, Paper 22, pp. 281-297.

12. Lewis, W. J., Faulkner, G. E., and Rieppel, P. J., "Flux and Filler Wire Developments for Submerged Arc Welding of HY80 Steel," *Welding Journal*, 40 (8), Aug. 1961, Research Suppl., pp. 337-s to 345-s.

13. Ul'yanov, V. I., Parfessa, G. I., and Shevchuk, R. N., "Effects of Aluminum in Electrode Wire on the Strength of CO_2 Weld Metal in St. 3 Steel," *Automatic Welding*, (2) 1974, pp. 15-19 (Welding Institute translation from Russian).

14. Sakaki, H., "Effect of Alloying Elements on the Notch Toughness of Basic Weld Metals," *Journal of Japan Welding Soc.*, V. 28, 1959, pp. 858-863.

15. Koukabi, A., North, T. H., to be published.

16. Dolby, R. E., "Factors Controlling Weld Toughness—the Present Position. Part 2. Weld Metals," Welding Institute report 14/1976/M. 1976, pp. 1-17.

17. Falce, J., and Henry, J. M., "Influence of Zirconium on the Properties of Weldability of HSLA Steel Containing Niobium," joint ASM-AIM symposium—Welding of HSLA (Microalloyed) Structural Steels, 8-12 Nov. 1976, Rome, Paper 5.

18. McLean, A., and Kay, D. A. R., Control of Inclusions in High-Strength Low-Alloy Steels," *Microalloying* 75, Oct. 1975, Washington, pp. 215-231.

19. Gladman, T., Dulieu, D., and McIvor, I., "Structure-Property Relationships in High-Strength Microalloyed Steels," *Microalloying* 75, Oct. 1975, Washington, pp. 32-54.

20. Jesseman, R. J., "Columbium Pick-up in High Dilution Submerged Arc Welds," *Microalloying* 75, Oct. 1975, Washington, pp. 578-592.

21. Aronson, H. J., "The Proeutectoid Ferrite and Proeutectoid Cementite Reactions," Met. Soc. AIME Symposium on Decomposition of Austenite by Diffusional Processes, Oct. 1960, Philadelphia, pp. 387-546.

22. Silcock, J. M., and Denham, A. W., "Precipitation of NaCl-type Carbides in Austenite and Their Behaviour in the Neighbourhood of Grain Boundaries," The Institute of Metals Symposium on The Mechanism of Phase Transformations in Crystalline Solids, July 1968, Manchester, pp. 59-63.

23. Baker, R. G., and Nutting, J., "Precipitation Processes in Steels," Iron & Steel Institute Special Report, 1969, V. 64, pp. 1-40.

24. Chadwick, G. A., *Metallography of Phase Transformations*, Butterworth Publishers, London 1972, pp. 193-200.

25. Funakoshi, T., Tanaka, T., Ueda, S. Iskikawa, M., Koshizuka, N., and Kobayashi, K., *Trans. I.S.I.*, Japan, V. 7, 1977, pp. 419-427.

26. Craig, I., and North. T. H., to be published.

27. Stout, R. D., WRC Weldability Committee-Lehigh Program Progress Report, Feb. 1978, Bethlehem.

28. Nowicki, A., "Effect of Using CaF_2-Al_2O_3-CaO Based Fluxes on the Metallurgical Properties of Submerged Arc Weld Metal," Ph.D. Thesis, 1977, University of Strathclyde.

29. Chakravarty, S. K., "Study of the Effect of Flux Composition on the Composition and Properties of Submerged Arc Welds," M.Sc. Thesis, 1977, University of Strathclyde.

30. Baumann, S. F., Sawhill, J. M., and Nakabayashi, M., "Wire and Flux Development for Seam Welding of HSLA Line Pipe," WRC Monograph, *Welding of Line Pipe Steels*, 1977, pp. 56-86.

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Generalized Yield Surfaces for Plates and Shells

by D.B. Peterson, W.C. Kroenke, W.F. Stokey and W.J. O'Donnell

Complete expressions are derived for the lower bound loads which cause yielding of general plate and shell elements subjected to membrane, bending and shear loads. Stress limits based on yield surface equations are proposed to evaluate elastically calculated primary stresses obtained, using finite element or other methods. The proposed evaluation method provides a consistent and nearly uniform margin of safety against gross plastic yielding.

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