Study of welding characteristics of 0.3C–CrMoV(ESR) ultrahigh strength steel

M. R. Suresh · P. P. Sinha · D. S. Sarma · N. B. Ballal · P. Krishna Rao

Abstract A new ultrahigh strength low alloy steel 0.3C–CrMoV(ESR), having an ultimate tensile strength and 0.2% proof strength of above 1,700 and 1,500 MPa, respectively, in quenched and tempered condition, was developed primarily as a cost effective material for space launch vehicle applications. Welding is a major step in the fabrication of most of the pressure vessels, structures and equipments. Steels with carbon equivalent in excess of 0.40 wt% show a tendency to form martensite on welding, and therefore are considered difficult to weld. 0.3C–CrMoV(ESR) steel has a carbon equivalent value of nearly 1.0 that classifies it as a 'very difficult to weld' steel. In addition it has a niobium content of about 0.10% and a vanadium content of 0.25%. It is known that niobium content of more than 0.02 wt% has a deleterious effect on the toughness properties of low carbon welds. It has also been reported that the effect of niobium on weld metal toughness is more deleterious in the presence of vanadium. Hence, in the present study, the properties of the weldment of this new steel under different heat treatment conditions (HT-1 and HT-2) have been studied. In HT-1 condition, the plates were welded in hardened and tempered condition and no further heat treatment was given after welding, while in HT-2 condition, the annealed plates were subjected to welding followed by hardening and tempering heat treatments. For HT-1 condition, only tensile properties were evaluated. The welded plates under HT-2 condition were evaluated for tensile properties, fracture toughness, residual strength and microstructure features.

Introduction

0.3C–CrMoV(ESR) steel is a new ultrahigh strength low alloy steel developed primarily for aerospace pressure vessel applications such as booster motorcases. Pressure vessels meant for space applications are thin walled with wall thickness of around 10 mm. Its nominal chemical composition and average tensile properties are given in Tables 1 and 2, respectively. The present materials used for such pressure vessels world over are D6AC steel, Maraging steel, etc. The new material is a low alloy steel and does not contain any costly alloying elements. Further, its manufacturing route does not require any virgin element or vacuum melting process. Hence, in comparison to existing space materials, this new steel is very much cost effective. During alloy design, care had been taken to keep carbon content below 0.31% since carbon is known to affect the toughness and weld properties adversely. To further improve the welding behaviour of the steel, its processing was carried out through electroslag remelting (ESR) technique by improving cleanliness and solidification structure of the metal. In addition, inoculation using niobium was also carried out for grain refining, which contributes favourably

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Table 1 Nominal chemical composition of 0.3C–CrMoV(ESR) steel

Element	C	Mn	Mo	Cr	V	S	P	Nb	Fe
wt%	0.28	0.90	0.90	1.35	0.25	<0.015	<0.02	0.10	Bal.

Table 2 Average tensile properties of steel austenitized at 925 °C/oil quenched

Tempering temperature (°C)	UTS (MPa)	0.2%PS (MPa)	%El.
200	1,726	1,472	9.60
450	1,530	1,408	11.20

to welding and toughness characteristics. The steel has an ultimate tensile strength of more than 1,700 MPa, in quenched and low temperature-tempered condition, and its 0.2% proof strength is in the vicinity of 1,500 MPa.

Welding is a major step in the fabrication of most of the equipments, including pressure vessels for launch vehicles. The limiting stress in ultrahigh strength steels is usually fixed by the onset of brittle fracture in the presence of crack-like defects. This type of situation is encountered in materials that have the net failure stress (residual strength) lower than the yield strength. It is also important that the weld has strength and toughness comparable to those of the parent metal, since the design allowances and factor of safety are kept to a minimum.

The quality of a weld is governed by the quality of the weld metal (fusion zone) and the heat affected zone (HAZ). Major welding parameters that affect the quality are: specific heat input, composition of the parent metal and the weld metal, impurity levels in the metal, plate thickness, pre-heating and post-heat treatment, number of passes and amount of restraint during welding.

In welding, hardenability is often expressed in terms of "Carbon Equivalent". For alloys containing carbon in excess of 0.18%. The empirical formula recommended by IIW gives the carbon equivalent [1]:

$$CE = C + (Mn + Si)/6 + (Ni + Cu)/15 \\ + (Cr + Mo + V)/5 \, wt\%.$$

Steels with carbon equivalent in excess of 0.40 wt% show a tendency to form martensite on welding, and therefore are considered difficult to weld. 0.3C–CrMo-V(ESR) steel has a carbon equivalent value of nearly 1.0 that classifies it as a 'very difficult to weld' steel. In addition it has a niobium content of about 0.10% and a vanadium content of 0.25%. Levine and Hill [2] reported that niobium content of more than 0.02 wt% has a deleterious effect on the toughness properties of low carbon welds. It has also been reported that the effect of niobium on weld metal toughness is more deleterious in the presence of vanadium [2, 3].

Although the design of the alloy and processing technique was devised such that the even higher carbon content does not adversely affect the weld properties, an investigation was initiated to evaluate actual properties of this steel in welded condition.

Experimental

The steel was produced at the facility of M/s. Mishra Dhatu Nigam Limited, Hyderabad by electric arc melting followed by casting to primary electrodes and remelting of these electrodes by the ESR process. During electroslag remelting, the steel was inoculated with niobium to refine the grain size [4]. The electroslag refined ingots weighing around 4 tonnes were then forged to slabs and finally hot rolled to plates of size 1,000 mm × 1,000 mm having thicknesses of 7.8 and 12.5 mm, respectively.

The welding parameters for this steel were selected based on earlier studies on the weldability of this steel [5] through bead on plate and crack susceptibility tests such as varestraint test [6]. The optimum parameters finalised from the outcome of these studies are given in Table 3. Welding studies were conducted on 7.8 and 12.5 mm thick plates. Gas Tungsten Arc welding (GTAW) process was employed throughout the present investigation since this process is known to produce high quality low hydrogen weld joints, particularly in realizing thin walled pressure vessels. Moreover, it is the most preferred process for high strength steels with yield strength above 1,000 MPa [7]. The details of the weld set up are given in Fig. 1. The flow rate of argon shielding gas was 10 lpm while the purging gas flow rate was 6 lpm. The interpass temperature was maintained above 300 °C. No preheating of the weld joint was employed.

Initially several compositions and preparation procedures for the filler wire were studied, and the one giving the best properties was selected. Evaluation of the strength and toughness properties of the weld joints made using wires having different chemical compositions helped in selecting the appropriate filler wire for welding this steel. It has been found from these studies that filler wire of the same chemical composition as that of the parent metal is the most suitable one. The selected filler wire was also pro-

Table 3 Typical GTAW parameters used for plates

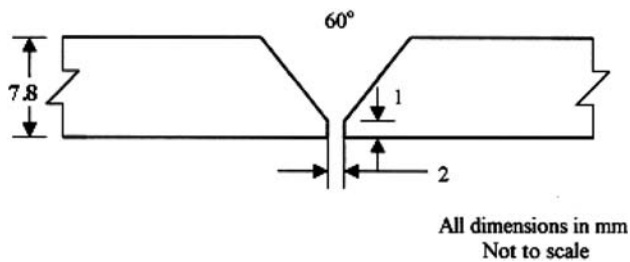| Pass	Voltage (V)	Current (A)	Wire feed rate (mm/m)
I	11–14	225	600
II	11–14	220	620
III	11–14	220	620
IV	11–14	225	620

Fig. 1 TIG welding set up of 7.8 mm thick plate

cessed through ESR and inoculated with niobium, and belonged to the same heat as that of the parent metal used. Low inclusion levels are important to decrease the possibility of weld cracking at high strength levels. Niobium in the wire was meant to refine the grain size of the weld metal. These two measures were expected to improve the toughness of the weldment.

Welding studies were carried out in two heat treatment conditions—(a) welding the plates after they were given the desired quench-and-temper treatment (HT-1 condition) and (b) welding the annealed plates and then heat treating the welded plates to obtain the desired properties (HT-2 condition). In both the cases, the weldments were characterized by testing for tensile properties, fracture toughness, residual strength in the presence of surface cracks, and optical and electron microscopy.

HT-1 condition: welding of plates in hardened and tempered condition

The plates to be welded were austenitized at 925 °C and tempered either at 200 or 450 °C. After welding no further heat treatment operation was carried out. The tensile properties of the steel were evaluated using specimens with the welded region at the centre.

HT-2 condition: welding of plates in annealed condition followed by hardening and tempering

The plate coupons were annealed at 875 °C prior to the welding operation. The welded plate was austenitized at 875, 925 or 1,000 °C and quenched. After this hardening operation, the hardness profile was again taken. Subsequently, the plates were tempered at either 200 or 450 °C, and the tensile properties were determined.

Uni-axial tensile tests were carried out on 7.8 mm thick flat tensile specimens conforming to ASTM A-370 standard. Tensile tests were carried out at a strain rate of 10^{-3}/s with a crosshead speed of 2 mm/min using an INSTRON testing machine.

Fracture toughness evaluation was carried out using compact tension (CT) specimens of 7.5 mm thickness as

well as 12.50 mm thickness in accordance with ASTM E-399 using an INSTRON testing machine.

The procedure for residual strength measurement through SCT test was carried out following ASTM E-740 standard. This involved testing specimens, which had been pre-cracked. Three-point bending was used to pre-crack specimens in fatigue in an INSTRON testing machine. A magnifier of about 20× was used to monitor the fatigue pre-cracking process. The ratio of minimum to maximum cyclic stress 'R' was kept at less than 0.1. The fatigue-cracked specimens were subsequently broken apart by applying uni-axial load.

The steel specimen in the quenched and tempered condition was etched with 2% nital solution and viewed under an optical microscope. Optical metallography using an Olympus PMG-3 metallurgical microscope was carried out to analyse the gross micro-structural features of the steel as well as to estimate its grain size. Fractured surfaces were examined under a Philips scanning electron microscope. For transmission electron microscopy (TEM), thin slices of thickness 0.125 mm were cut from the specimens on a Buehler Isomet low speed saw and they were further thinned by mechanical polishing to 0.03 mm. Discs of 3 mm diameter, punched off from these slices, were electro-polished using 90% acetic acid solution at 30–35 V, at 10 °C. The thin foils were examined in a JEOL CX-200 Transmission Electron Microscope operating at 160 kV.

Results and discussion

Welding of heat treated plates (HT-1 condition)

Typical tensile values obtained from plates welded in heat treated condition are shown in Table 4. The location of fracture on the specimens was always in the weld metal. The tensile properties obtained from plates tempered at two different temperatures fall in almost the same range, indicating that prior heat treatment given to the base metal has practically no effect on the properties of the weld metal. However, the properties are substantially lower than those of the base metal. In fact, the ultimate tensile strength obtained on the weld metal, is comparable to the 0.2% proof stress of the parent metal. The corresponding elongation and fracture toughness (K_{Ic}) of 50–55 MPa√m obtained in this condition was also considerably lower than the parent metal properties. The residual strength values were in the vicinity of 1,020–1,080 MPa even for a comparatively smaller crack size of 3×1 mm^2. These results indicate that welding of this steel in hardened and tempered condition is not the right choice and there is a need to give post weld heat treatment to modify the microstructure to get optimal mechanical properties.

Table 4 Tensile properties of the steel welded after heat treating the plates (HT-1 condition)

Hardening temperature (°C)	Tempering temperature (°C)	UTS (MPa)	0.2%PS (MPa)	%El.
925/O.Q.	200	1,422	1,315	5.60
		1,413	1,285	5.60
		1,393	1,285	7.20
		1,413	1,315	5.20
		1,422	1,315	5.40
925/O.Q.	450	1,452	1,324	5.30
		1,452	1,334	4.90
		1,383	1,275	6.00
		1,393	1,295	5.70

Further study was therefore restricted to steel welded in annealed condition followed by heat treatment (HT-2 condition).

Welding of annealed plates followed by subsequent heat treatment (HT-2 condition)

The plate coupons were annealed at 875 °C before carrying out the welding operation. After welding the hardness profile on the welded plate from the centre of the weld to the parent metal was taken and it is shown in Fig. 2. Subsequently the welded plate was austenitized at 875 or 925 °C and quenched in oil. After this hardening operation, the hardness profile was again taken (Fig. 3). As can be seen, the hardness values did not show any variation from the weld metal to the parent metal in this case. Subsequently, the plates were tempered at two different temperatures of 200 and 450 °C, and the tensile properties were determined (Table 5).

Fracture of the tensile specimens never occurred in the weld metal, indicating that the weld metal was stronger
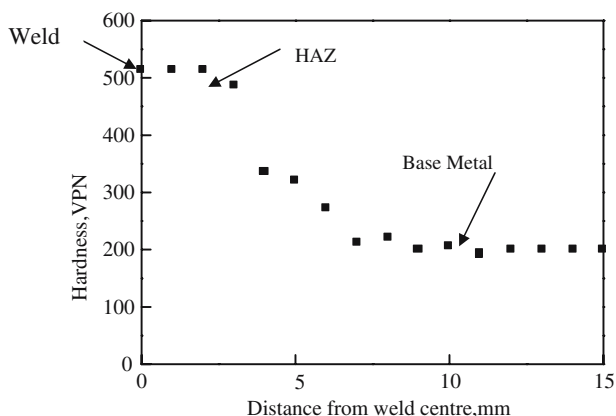
Fig. 2 Variation of hardness from the centre of the weld to the parent metal. Plates welded after annealing
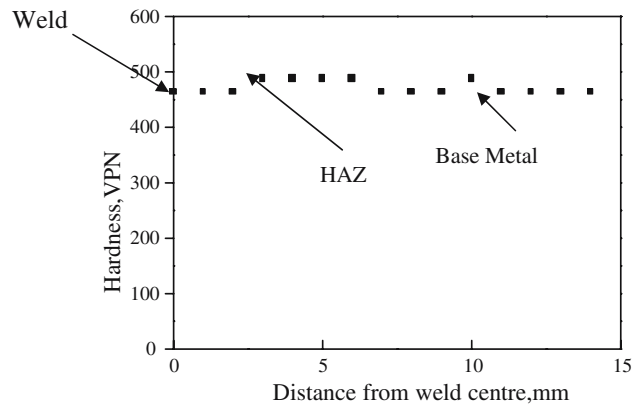
Fig. 3 Variation of hardness from the centre of the weld to the parent metal. Annealed plates were welded and then hardened (HT-2 condition)

than the parent metal. The strength values obtained show trends similar to those observed in the parent metal as expected. For example, the yield strength for samples austenitized at 925 °C and tempered at 450 °C was 1,370 MPa (parent metal 1,410 MPa). The slightly lower elongation measured may not therefore be representative of the material since the elongation would be the sum of the elongations of the weld metal as well as the parent metal, instead of the elongation of weld metal alone.

Fracture toughness

The fracture toughness of the weld metal austenitized at 925 °C and tempered at 200 °C as well as at 450 °C was evaluated and the obtained values are given in Table 6. These values, which are between 76 MPa√m and 80 MPa√m for 7.5 mm thick specimens, are quite high and therefore did not satisfy the thickness validity criterion laid down in ASTM E-399 standard. Hence 12.50 mm thick plates were welded and the fracture toughness values were evaluated for a tempering temperature of 450 °C. These values (Table 6) are found to be valid and hence represent the plane strain fracture toughness, K_{1c}, of the weld material.

It is interesting to note that the above K_{1c} values of the weld metal are as high as 89–94 MPa√m. These values are comparable or even marginally higher than those of the parent metal. This is in spite of the fact that the weld metal has undergone no deformation and it is not ESR processed. The high values of fracture toughness of the weld metal may be attributed to the following reasons: (1) the fused metal has high cleanliness since the parent metal as well as the filler wire are both clean due to prior ESR processing (2) the niobium in the filler wire is expected to have acted effectively as an inoculant in the small weld pool, refining the as-cast structure of the weld metal (3) the multi-pass

Table 5 Tensile properties of welded plates. Annealed plates were welded and then heat treated (HT-2 condition)

Hardening temperature (°C)	Quench medium	Tempering temperature (°C)	UTS (MPa)	0.2%PS (MPa)	%El.
875	Oil	No tempering	1,738	1,295	8.40
			1,727	1,309	7.00
		200	1,641	1,386	8.90
			1,631	1,369	9.00
		450	1,368	1,281	10.10
			1,356	1,267	9.40
		600	1,275	1,197	11.30
			1,268	1,190	12.00
925	Oil	No Tempering	1,770	1,320	8.80
			1,752	1,320	8.90
		200	1,638	1,383	8.40
			1,668	1,354	9.00
			1,668	1,364	9.40
			1,658	1,324	8.80
		450	1,481	1,383	9.30
			1,491	1,403	9.20
			1,427	1,345	10.20
			1,429	1,350	11.80
1,000	Oil	No Tempering	1,757	1,320	9.80
			1,738	1,271	8.60

welding employed, in which each overlaying weld run leads to heat treatment of the layers beneath it, results in a more refined and homogenised structure.

Residual strength

Residual strength of the weld metal for different sizes of surface cracks was determined in accordance with ASTM E-740 standard and the results are given in Tables 7, 8 and 9. The variation of the residual strength and the stress intensity factor with crack size (a/B) for the weld metal in the heat treatment condition of 925 °C/200 °C and 925 °C/450 °C are shown in Figs. 4, 5, respectively.

The residual strength in the presence of crack is lower for the weld metal than for the parent metal for both tempering temperatures. However, the difference is much larger for specimens tempered at 200 °C than at 450 °C. For example, for specimens tempered at 200 °C, the residual strength for the parent metal [8] is 1,630 MPa (crack size 3.51×1.47 mm^2) while that for the weld metal it is only 1,352 MPa (crack size 3.63×1.46 mm^2). The corresponding values for specimens tempered at 450 °C are:

Table 6 Fracture toughness of weld metal of the steel—annealed plates were welded and then heat treated (HT-2 condition)

Specimen thickness (mm)	Heat treatment condition	Fracture toughness (K_Q) of welded plates (MPa√m)	Remarks	Fracture toughness (K_Q) of parent metal (MPa√m)
7.50	925 °C/O.Q./200 °C	76.30	Invalid	100.50
		80.00		100.50
		79.60		99.80
		77.90		98.10
7.50	925 °C/O.Q./450 °C	76.04	Invalid	83.20
		79.80		82.50
		78.28		82.50
		77.67		84.60
		77.77		
12.50	925 °C/O.Q./450 °C	89.10	Valid	81.50
		92.50		82.00
		93.70		80.00

Fracture toughness values of parent metal are also shown for comparison

Table 7 SCT test results for the weld metal for heat treatment condition 925 °C/O.Q./200 °C—annealed plates were welded and then heat treated (HT-2 condition)

Crack size (2c × a) (mm)	Specimen thickness (mm)	a/B	a/φ²	Residual strength (MPa)	K_{Ie} (ASTM) (MPa√m)	K_{Ie} (Newman) (MPa√m)	K_{Ie} (Irwin) (MPa√m)
3.37 × 1.3	7.04	0.18	0.67	1,444	71	69	72
3.63 × 1.46	7.14	0.20	0.72	1,352	69	68	71
4.48 × 1.74	7.20	0.24	0.89	1,336	76	74	77
4.62 × 1.83	7.03	0.26	0.92	1,384	80	78	82
5.88 × 2.16	7.14	0.30	1.15	1,272	84	81	84
6.12 × 2.24	7.15	0.31	1.19	1,207	82	79	81

Table 8 SCT test results for the weld metal for heat treatment condition 925 °C/O.Q./450 °C—annealed plates were welded and then heat treated (HT-2 condition)

Crack size (mm)	Specimen thickness (mm)	a/B	a/φ²	Residual strength (MPa)	K_{Ie} (ASTM) (MPa√m)	K_{Ie} (Newman) (MPa√m)	K_{Ie} (Irwin) (MPa√m)
2.92 × 1.15	4.92	0.23	0.58	1,361	63	61	64
3.06 × 1.18	4.93	0.24	0.60	1,449	68	66	69
3.83 × 1.56	4.90	0.32	0.76	1,374	73	71	74
4.32 × 1.55	4.92	0.32	0.84	1,407	80	77	79
4.86 × 1.62	4.92	0.33	0.93	1,367	83	79	81
4.90 × 1.79	7.30	0.25	0.96	1,392	83	81	84
4.97 × 1.79	7.30	0.25	0.97	1,376	83	80	83
5.16 × 2.02	7.14	0.28	1.02	1,344	83	80	84
5.19 × 1.70	4.94	0.34	0.98	1,391	87	83	85
5.26 × 2.01	7.12	0.28	1.04	1,288	80	78	81

parent metal 1,461 MPa (crack size 3.81 × 1.60 mm²) [8] and weld metal 1,374 MPa (crack size 3.83 × 1.56 mm²). The smaller difference at the higher tempering temperature may be the result of higher ductility in this tempered condition making the material less notch sensitive.

From Table 9 it can be seen that the stress intensity factor (SIF) initially increases with '$a/φ^2$' value, which is a good measure of the effect of crack size, and then remains relatively constant as was observed in the case of the parent material. '$φ$' is called the shape factor for elliptical cracks. The constant value of SIF may be taken as the apparent K_{1c} value. These apparent K_{1c} values are slightly lower than those obtained from compact tension (CT) specimens. For samples tempered at 450 °C, the apparent K_{1c} value of the
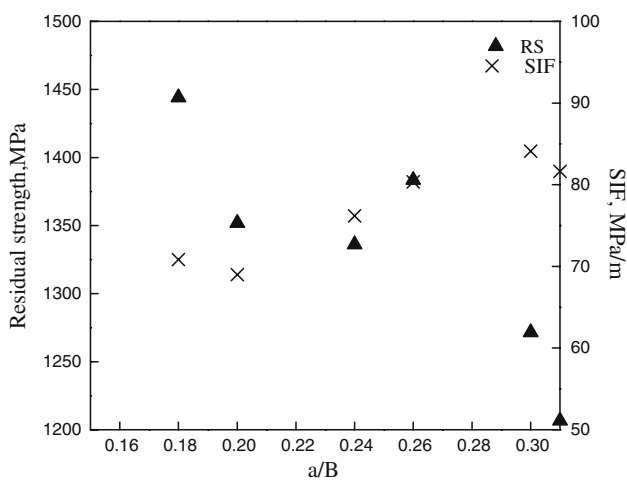

Fig. 4 Variation of residual strength (RS) and stress intensity factor (SIF) of the weld metal with a/B for the heat treatment condition 925 °C/O.Q./200 °C—annealed plates were welded and then heat treated (HT-2 condition)
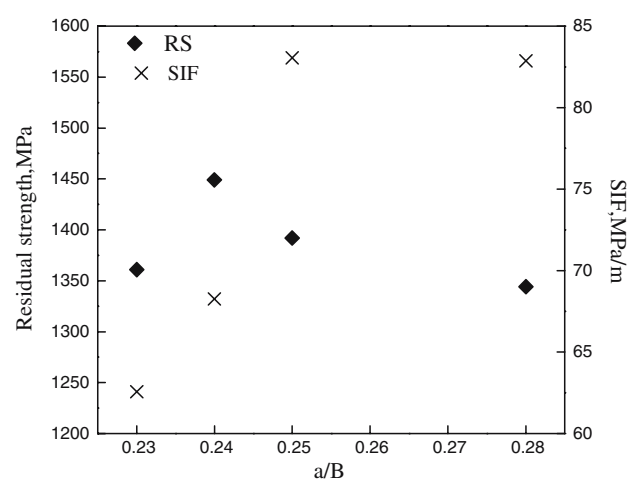
Fig. 5 Variation of residual strength (RS) and stress intensity factor (SIF) of the weld metal with a/B, for the heat treatment condition 925 °C/O.Q./450 °C—annealed plates were welded and then heat treated (HT-2 condition)

Table 9 Residual strength of the weld metal (HT-2 condition)

Heat treatment condition	Crack size (mm)	Specimen thickness (mm)	a/B	a/ϕ^2	Residual strength (MPa)	SIF (MPa$\sqrt{m}$)	$\frac{NSFS}{RS}$	Crack Area [A] (mm^2)	$\frac{A}{BW}$
925 °C/O.Q./200 °C	3.37 × 1.30	7.04	0.18	0.67	1,444	71	1.0142	3.44	0.0140
	3.63 × 1.46	7.14	0.20	0.72	1,352	69	1.0169	4.16	0.0166
	4.48 × 1.74	7.20	0.24	0.89	1,336	76	1.0249	6.12	0.0243
	4.62 × 1.83	7.03	0.26	0.92	1,384	80	1.0278	6.63	0.0270
	5.88 × 2.16	7.14	0.30	1.15	1,271	84	1.0416	9.97	0.0399
	6.12 × 2.24	7.15	0.31	1.19	1,207	82	1.0449	10.76	0.0430
925 °C/O.Q./450 °C	2.92 × 1.15	4.92	0.23	0.58	1,361	63	1.0155	2.64	0.0153
	3.06 × 1.18	4.93	0.24	0.60	1,449	68	1.0167	2.83	0.0164
	3.83 × 1.56	4.90	0.32	0.76	1,374	73	1.0281	4.69	0.0273
	4.32 × 1.55	4.92	0.32	0.84	1,407	80	1.0315	5.25	0.0305
	4.86 × 1.62	4.92	0.33	0.93	1,367	83	1.0372	6.18	0.0359
	4.90 × 1.79	7.30	0.25	0.96	1,392	83	1.0277	6.89	0.0269
	4.97 × 1.79	7.30	0.25	0.97	1,376	83	1.0281	6.98	0.0273
	5.16 × 2.02	7.14	0.28	1.02	1,344	83	1.0339	8.18	0.0327
	5.19 × 1.70	4.94	0.34	0.98	1,391	87	1.0417	6.93	0.0401
	5.26 × 2.01	7.12	0.28	1.04	1,288	80	1.0345	8.29	0.0333

SIF—stress intensity factor, NSFS—net section failure stress, B—specimen thickness, W—specimen width, O.Q.—oil quench, A—cross sectional area

weld varies between 80 MPa$\sqrt{m}$ and 87 MPa$\sqrt{m}$, whereas the values obtained from the CT specimens vary between 82 MPa$\sqrt{m}$ and 94 MPa$\sqrt{m}$.

Hall [9] had found that the calculated fracture toughness from surface crack tension (SCT) test results would be reasonably constant if both the crack depth 'a' and the uncracked ligament '$(B – a)$' were greater than about 0.5 $(K_{1c}/\sigma_{ys})^2$. However, this is not borne out by the present results. Thus, although in almost all the cases Hall's criterion is satisfied, the stress intensity factor (SIF) shows a variation at the lower crack sizes investigated. On the other hand the SIF value seems to become relatively constant at and above 'a/ϕ^2' values of about 0.90 for both the heat treatment conditions studied. It may be noted that even in the case of the parent metal [8] this criterion seems to hold true. This can be a subject for further investigation.

Micro-structural examination

Optical microscopy

The microstructure of the weld metal in the as welded condition showed fine dendritic structure (Fig. 6). Within the grains, fine laths could be seen though some areas are hazy (Fig. 7). Presumably the formation of martensite in the weld metal was facilitated by the cooling of the weld by the thick (7.8 mm) base plates resembling a quenching operation. The heat affected zone has a microstructure consisting mainly of lath indicating the presence of martensite or
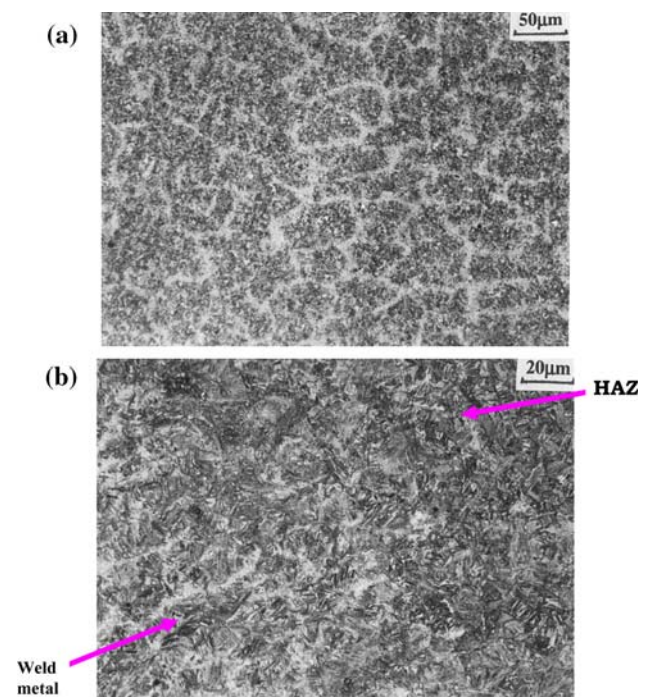

Fig. 6 Microstructure of (**a**) weld metal of steel in as-welded condition, 200× (**b**) region of weld metal and heat affected zone at higher magnification, 500×

tempered martensite. After austenitizing and quenching, the structure became fully martensitic as shown in Fig. 8.

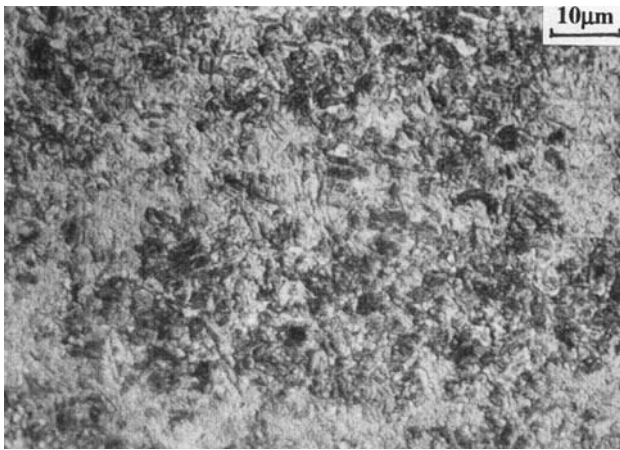The weld metal begins to solidify epitaxially on preexisting grains of the base metal. Due to the high tem-

Fig. 7 Microstructure of weld metal of steel in as-welded condition showing presence of fine martensite lath, 960×

Fig. 8 Microstructure of weld metal of steel in as-quenched condition, 960×

perature gradients, the structure becomes predominantly columnar thereafter. The shape and size of the austenite grains formed are extremely important in the evolution of the final microstructure, since the austenite grain boundary is the nucleation site for subsequent phase transformations [1]. Solidified metal is also inhomogeneous due to segregation, varying welding parameters and varying dilution from the parent metal. The final microstructure therefore can be quite non-homogeneous and may consist of ferrite, pearlite, martensite and retained austenite in pro-eutectoid steels. The weld metal can undergo further transformations during subsequent passes.

The grain size and impurities in the weld metal play a very decisive role in controlling the weld metal toughness. Fine grains can deform better to accommodate contraction stresses during solidification [10]. Due to the large grain boundary area, the concentration of harmful segregates at the grain boundary is also reduced. Oxide and sulphide inclusions can bring down the toughness drastically and need therefore to be kept to a minimum [11].

Transmission electron microscopy

The TEM photographs of thin foils of the weld metal of the steel under various heat treatment conditions as viewed at 160 keV under a JEOL 200CX TEM are given in Figs. 9–12.

The structure of the steel in the as-welded condition consisted of martensite laths. Within the laths as well as along their boundaries precipitates were seen which were identified as cementite platelets. This is similar to a structure obtained after quenching and tempering at 450 °C or higher. The cementite platelets were found to share Bagaryatski orientation relationship with the martensite as shown in Fig. 9d.

The specimens austenitized and quenched after welding had a fully martensitic structure with high dislocation density, as in Fig. 10. A few twins could also be observed in samples quenched from 875 °C. These twins were more frequent in the weld metal than in the parent metal. This is probably due to segregation of carbon during solidification. However no retained austenite could be observed in the weld metal.

After tempering at 200 °C, the dislocation density decreased and fine platelets of cementite began to precipitate out (Fig. 11). On increasing the tempering temperature to 450 °C, the cementite platelets grew in number and precipitated along the lath boundaries as well as within the laths (Fig. 12). The presence of globular vanadium carbide particles of about 100–200 Å size could be seen in addition to the cementite inside the lath. The vanadium carbide particles followed the Kurdjumov-Sachs orientation relationship with the martensite as given below:

$$(1\,0\,0)_{VC} \sim 5°(1\,1\,0)_M$$
$$(0\,0\,1)_{VC} \sim 5°(1\,-1\,0)_M$$
$$(-1\,0\,1)_{VC} \sim 5°(0\,1\,0)_M.$$

The microstructure evolution during heat treatment in the weld metal is similar to that observed in parent metal of the steel. The details of microstructure study are already published elsewhere [12].

Conclusions

1. 7.8 mm as well as 12.5 mm thick hot rolled plate of 0.3C–CrMoV(ESR) steel was successfully welded utilizing the GTAW process without any preheat. It

Fig. 9 TEM of weld metal in as-welded condition (a) bright field image showing cementite precipitates along martensite lath boundaries and within lath (b) dark field image of the same (c) SAD pattern of precipitate
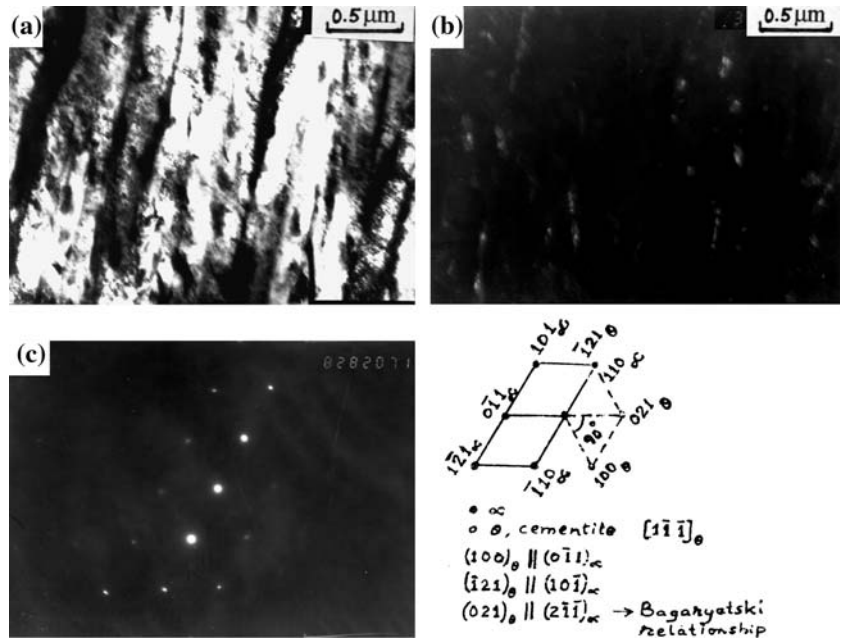

Fig. 10 TEM of weld metal quenched from 875 °C (a) bright field image showing dislocated martensite laths (b) bright field image showing twins and dislocations (c) dark field image of twins (d) SAD pattern of martensite
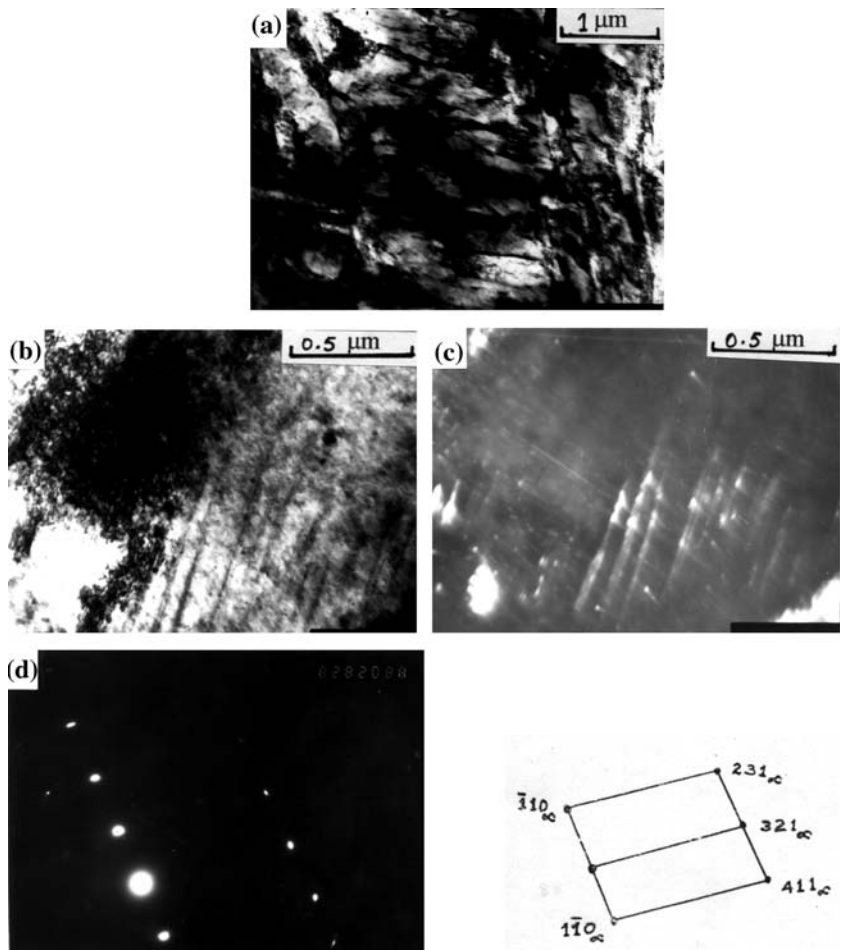
Fig. 11 TEM of weld metal quenched from 875 °C and tempered at 200 °C (**a**) bright field image of martensite lath with cementite precipitates (**b**) same at higher magnification (**c**) SAD pattern of precipitates
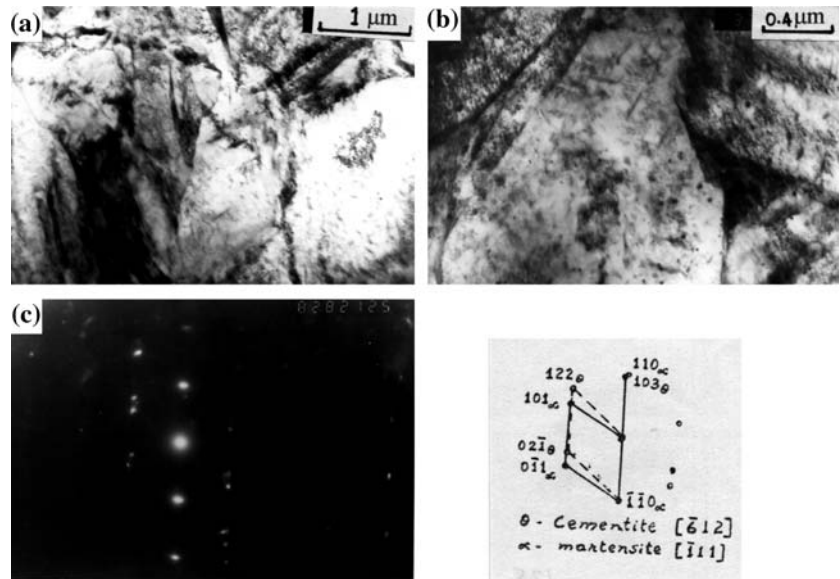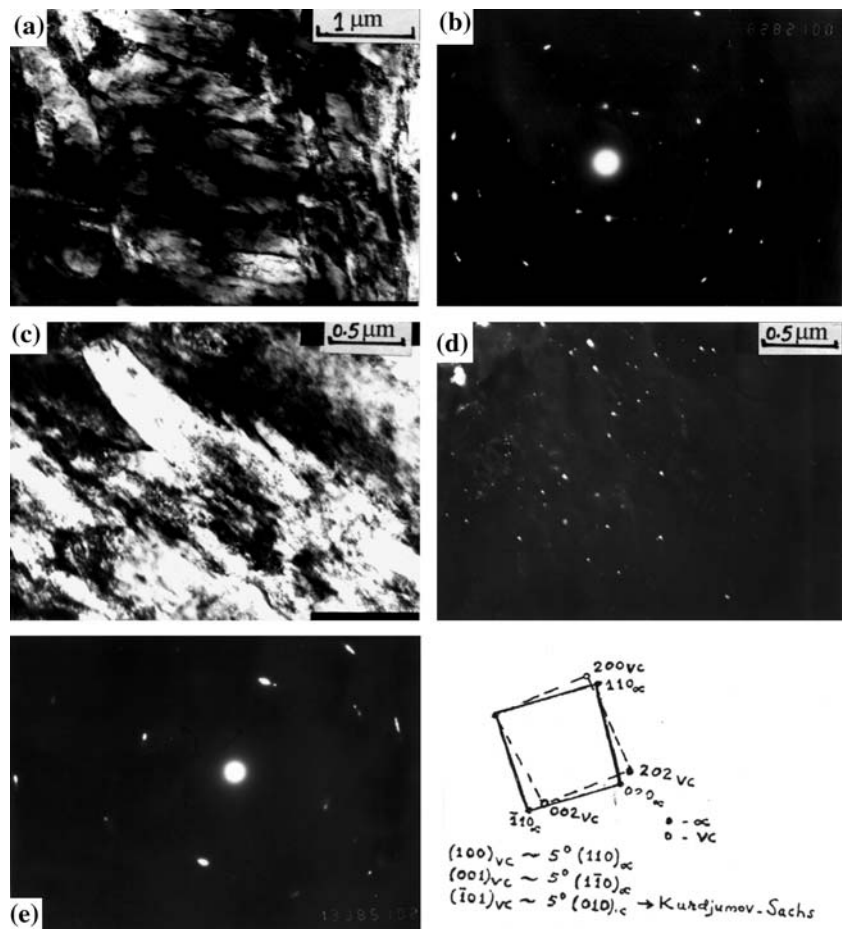

Fig. 12 TEM of weld metal quenched from 875 °C and tempered at 450 °C (**a**) bright field image showing martensite lath (**b**) SAD pattern (**c**) bright field image at higher magnification (**d**) dark field image of the same (**e**) SAD pattern

was found that a post-weld heat treatment of hardening from 925 °C followed by tempering at 200 or 450 °C was necessary to improve the toughness of the weld.

2. After heat treatment the tensile strength values obtained from welded plates in HT-2 condition were comparable to those of the parent metal. For the heat

treatment cycle of 925 °C/450 °C, the welded plates yielded average UTS of 1,457 MPa, 0.2% proof strength of 1,370 MPa and %Elongation of 10.10 where as the corresponding values for the parent metal are 1,530 MPa, 1,408 MPa and 11.20, respectively.

3. The plane strain fracture toughness (K_{Ic}) of the weld metal is quite high and the values obtained are comparable or even higher (80.10–93.70 MPa√m) than those of the parent metal (80–82 MPa√m). The residual strength values of the weld metal are however lower than those of the parent metal for both tempering conditions. For specimens tempered at 200 °C, the residual strength for the parent metal is 1,630 MPa (for crack size 3.51×1.47 mm^2) while that for the weld metal it is only 1,352 MPa (for crack size 3.63×1.46 mm^2). The corresponding values for specimens tempered at 450 °C are: parent metal 1,461 MPa (for crack size 3.81×1.60 mm^2) and weld metal 1,374 MPa (for crack size 3.83×1.56 mm^2).

4. Based on this study, it is found that this new steel can produce high quality welds with improved toughness and tensile properties, and that GTAW process can be employed successfully for fabricating thin-walled pressure vessels for space applications using this steel.

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