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A study on influence of heat input variation on microstructure of reduced activation ferritic martensitic steel weld metal produced by GTAW process

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ABSTRACT

Reduced activation ferritic martensitic (RAFM) steel is a major structural material for test blanket module (TBM) to be incorporated in International Thermonuclear Experimental Reactor (ITER) programme to study the breeding of tritium in fusion reactors. This material has been mainly developed to achieve significant reduction in the induced radioactivity from the structural material used. Fabrication of TBM involves extensive welding, and gas tungsten arc welding (GTAW) process is one of the welding processes being considered for this purpose. In the present work, the effect of heat input on microstructure of indigenously developed RAFM steel weld metal produced by GTAW process has been studied. Autogenous bead-on-plate welding, autogenous butt-welding, butt-welding with filler wire addition, and pulsed welding on RAFMS have been carried out using GTAW process respectively. The weld metal is found to contain δ-ferrite and its volume fraction increased with increase in heat input. This fact suggests that δ-ferrite content in the weld metal is influenced by the cooling rate during welding. It was also observed that the hardness of the weld metal decreased with increase in δ-ferrite content. This paper highlights the effect of heat input and PWHT duration on microstructure and hardness of welds.

1. Introduction

Reduced activation ferritic martensitic steel (RAFMS) otherwise known as LAFMS is a major structural material for test blank module (TBM) in fusion reactor [1,2]. This steel has been developed based on the composition of commercial ferritic/martensitic steels by replacing Mo and Nb of high-chromium heat resistant martensitic steels such as modified 9Cr–1Mo (P91) with W and Ta respectively. These steels possess desirable properties such as good mechanical properties, adequate creep resistance up to 550 °C and limited radiological activation [3,4]. Apart from conventional welding processes, advanced welding processes such as electron beam welding, laser beam welding and hybrid welding processes are also available for joining of RAFM steel but their use is restricted to heavy section [5]. Manufacturing of TBM component involves extensive welding, currently GTAW process is being considered for joining of headers of first wall, sidewall and hall of test blank module. In order to ensure better mechanical properties fully martensitic microstructure is preferred in weld metal in the as-welded condition. Duplex structure such as martensite along with δ ferrite in high Cr steels reduces toughness of base metal as well as welds. This has been due to δ ferrite formation along with carbides at the ferrite/martensite interface [6,7]. Anderko et al. observed that above 2% volume fraction of δ ferrite, cleavage fracture is favoured due to the presence of brittle $M_{23}C_6$ type carbides between ferrite and martensite matrix. It has been also observed that in tungsten bearing high Cr steels, toughness was found to be low due to the presence of δ ferrite [8]. There was a steep increase in DBTT with increase in δ ferrite content in the matrix. Hence delta (δ) ferrite is not preferable in weld as it reduces notch toughness, promotes sigma-phase precipitation, and reduces creep ductility at high temperatures [9]. The ferrite formation depends on composition as well as welding process parameters.

In the present work, GTAW process studies using continuous mode, and pulsed mode were carried out on RAFMS steel. Post weld heat treatment (PWHT) is mandatory before putting the weld in service due to untempered, brittle martensite structure in weld [10]. The present study highlights the effect of heat input and PWHT duration on microstructure and hardness of welds.

2. Experimental work

The base material was supplied in the normalized (980 °C – 0.5 h) and tempered (760 °C – 1 h) condition by MIDHANI, Hyderabad. Its nominal chemical composition evaluated using optical emission spectroscopy (wt.%) is as follows: Cr 9.03, C 0.13, Mn 0.53, V 0.25, W 1.05, Ta 0.07, N 0.04, P 0.002, S 0.001, Ti 0.005, Ni 0.02 and Si 0.03. The mechanical properties at room temperature in

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Table 1
Heat input and hardness of welds.

Weld	Heat input (kJ/mm)	As weld hardness (VHN)	Hardness after PWHT at 750 °C (VHN)		
			1 h	2 h	3 h
Autogenous bead-on-plate	0.81	340	275	271	256
	0.66	377	277	273	261
	0.52	388	280	275	265
Autogenous butt joint	0.81	352	286	261	215
	0.66	360	289	273	225
	0.52	371	271	277	224
Butt joint with filler	0.63	391	338	257	233
	0.57	400	337	250	238
	0.45	410	365	236	198
Base metal			215	209	204

the as-received condition are as follows: yield strength: 500 MPa; ultimate tensile strength: 646 MPa; elongation: 24% and reduction of area: 76%. Rectangular pieces of size 80 mm × 30 mm × 2.5 mm were used for the welding trials. Bead-on-plate, autogenous butt welding, single V-groove (included angle 65°; root face 0.5 mm) butt-welding using GTAW process with filler wires and fabricated form of the same base metal of 1.2 mm diameter were used in the present investigation. Pulsed GTAW was also carried out on autogenous mode to study the influence of pulsing on microstructural constituents. The continuous welding parameter used in the present is listed in Table 1 and pulsed GTAW (heat input 0.696 kJ/mm) parameters are as follows: peak current 60 A (duration 5 s), peak voltage 12 Vand background current 20 A (duration 2 s).

Commercially pure argon (99.99%) was used a shielding gas at a flow rate of 10 l/min during welding. Both preheat and interpass temperatures were maintained at 250 °C and monitored using thermal crayons during welding. An arc gap of 1 mm was maintained during welding trails. The heat input during welding for each process is listed in Table 1.Subsequently the joints were subjected to PWHT at 750 °C for various durations (1 h, 2 h and 3 h) in a muffle furnace followed by cooling in still air. Optical microscopy was done on samples etched with Vilella's reagent. The δ ferrite content was quantified using optical microscope interfaced with an image analysis software. For each sample nearly 10 fields of observation were carried out and the reported value is the average of the same. Vickers macrohardness was carried out on both as-weld and post weld heat-treated samples at a load of 5 kg with duration of 30 s as per ASTM E-92.

3. Results and discussion

3.1. Microstructure

3.1.1. Base metal and autogenous weld

The base metal in the as-received condition shows tempered martensite (Fig. 1), free from δ ferrite. The average prior austenite grain size of the base metal is ASTM No. 9. The microstructure consists of predominately carbides at prior austenite grain boundaries with a few small MX type precipitates in the matrix as reported elsewhere [2].

Fig. 2a shows the optical micrograph of autogenously produced as-weld microstructure. Clearly δ ferrite was observed in the untempered martensite microstructure. More amount of δ ferrite was observed at grain boundaries, whereas the interior of grain shows less δ ferrite phase. This is due to more free energy available at the grain boundary than at the grain interior, hence the variation in quantity of δ ferrite. There was a wide variation in morphology and size of δ ferrite phase as observed in Fig. 2a.

The weld produced using heat input of 0.66 kJ/mm has the ferrite content of 6%. There was a considerable amount of decrease in δ ferrite at heat input 0.66 kJ/mm as compared with heat input of 0.81 kJ/mm (ferrite content 10%). Fig. 2b shows the as-weld microstructure at a heat input of 0.52 kJ/mm. A decrease in δ ferrite content with a decrease in heat input hence increase in cooling rate trend is observed. Fig. 3 shows the SEM image of as-weld microstructure. At a higher magnification, it was clearly observed that δ ferrite was free from carbide.

From the results presented above, it is clear that δ ferrite content in the weld metal is a function of heat input during welding. In order to substantiate the above facts, without preheat also a weld joint was prepared at a heat input of 0.66 kJ/mm to observe the effect of cooling rate (85 °C/s) on ferrite content. Fig. 4 shows the micrograph of weld done without preheat during welding. There is nearly complete absence of δ ferrite (Fig. 4) compared with weld completed using preheat at a heat input of 0.66 kJ/mm. This also confirms the observation that there is a critical cooling rate for weld above in which the ferrite formation is more pronounced. In an unpublished report [11], there was an absence of δ ferrite in electron beam welded (EBW) RAFM steel. It is evident that the heat input to the EBW weldment is significantly lower than that input to GTAW weldment and thus the EBW specimen cools far more rapidly. Also, in contrast to the GTAW welding process, the EBW technique generates a high energy density and therefore results in a low heat input and a rapid cooling rate. Fig. 5 shows the microstructure of pulsed GTA weld of RAFM steel. There is a complete absence of ferrite in comparison with continuous mode microstructure of

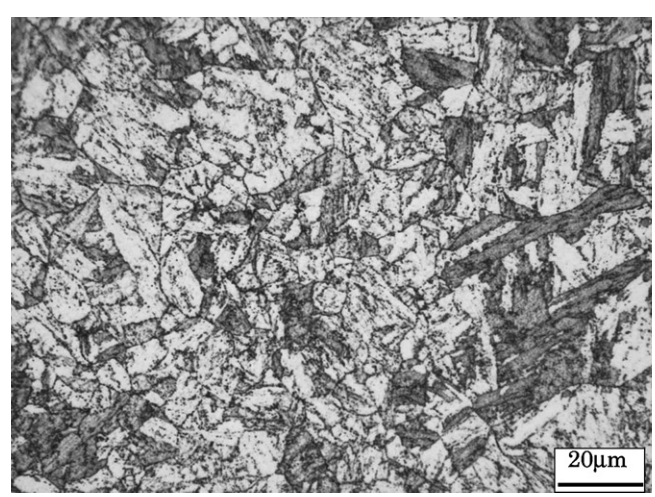

Fig. 1. Normalized and tempered microstructure.