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A critical assessment of the microstructure and mechanical properties of friction stir welded reduced activation ferritic–martensitic steel

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ABSTRACT

Bead-on-plate friction stir welding was conducted on 6 mm thick plate of Reduced Activation Ferritic–Martensitic Steel employing polycrystalline cubic boron nitride tool with rotational speeds of 200, 300, 500 and 700 rpm and traverse speed of 30 mm/min. The interface temperature between shoulder bottom and top surface of the plate was monitored by non-contact in-line thermography which served to identify the peak temperature attained in the stir zone (SZ). This temperature for 200, 300 and 500, and 700 rpm was respectively below Ac_1, between Ac_1 and Ac_3, and above Ac_3. In the base metal (BM), the prior austenite grain and martensite lath boundaries were decorated with chromium and tungsten rich $M_{23}C_6$ precipitates while intra-lath regions revealed Ta and V rich MX type carbides. Rotational speeds greater than 300 rpm led to martensite formation and simultaneous recovery, recrystallization and grain growth in SZs with wide distribution in grain size whereas SZ of 200 rpm and BM possessed similar distribution. The grain boundary $M_{23}C_6$ dissolved and very fine needles of Fe_3C precipitated in all SZs. The hardness of all SZs was unacceptably higher compared to the BM. The 200 rpm weld exhibited higher impact toughness in the absence of martensite in SZ.

1. Introduction

Modified 9Cr–1Mo Ferritic–Martensitic steel (Mod.9Cr–1Mo) had been often used for many high temperature components in steam generators of nuclear power plants, fossil-fired power plants and chemical process industries. The extensive usage of this alloy was based on an exceptional combination of the high temperature yield strength, tensile strength, ductility, excellent forming and welding characteristics and superior creep resistance at elevated temperatures [1,2]. In spite of its excellent mechanical properties, Mod.9Cr–1Mo was deficient as a structural material for construction of first wall and test blanket modules of the fusion reactors due to the high dose of residual radioactivity, originating from the long lived transmutation nuclides of Mo, Nb and nitrogen [3–5]. In order to facilitate easy handling of the core components in fusion reactors at the end of service, several reduced activation ferritic–martensitic (RAFM) steels for first wall and blanket modules are under consideration. The rapid decay of induced radioactivity after irradiation in a fusion reactor has been planned to achieve by substituting W and Ta respectively for Mo and Nb present in Mod.9Cr–1Mo steel. Very extensive research work worldwide led to the development of Eurofer-97 (Europe), F82H (Japan), CLAM (China), and 9Cr2WVTa (ORNL, USA) [6–12]. India is one of the countries associated with the development and testing of test blanket modules (TBMs) in International Thermonuclear Experimental Reactor (ITER). India's participation in ITER programme necessitated the development of India-specific RAFM steel for TBM. Based on extensive mechanical tests including impact, tensile, creep and fatigue on several heats with tungsten in the range 1–2 wt.% and tantalum in the range 0.06 to 0.14 wt.%, the RAFM steel having 1.4 wt.% tungsten with 0.06 wt.% tantalum was found to possess better combination of strength and toughness needed for TBM [13–16]. This steel is designated as Indian-specific RAFM (INRAFM) steel and currently being used for TBM fabrication.

All RAFM steels derive their high temperature strength mainly from the complex microstructures comprising of a high dislocation density, grain, lath and sub-grain boundaries decorated with $M_{23}C_6$ type carbides, and very fine MX type of carbides/carbonitrides in the intragranular regions. In general MX type of carbides are made up of Ta, V and Carbon, whereas $M_{23}C_6$ type carbides are chromium rich with considerable solubility for W in precipitate depending on the

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Table 1
Chemical composition of reduced activation ferritic–martensitic steel (INRAFM).

Elements	C	W	Mn	Cr	Ni	S	P	Mo	Ti	Co	V
wt.%	0.1	1.34	0.54	8.97	0.008	0.002	0.004	<0.002	<0.005	0.007	0.23
Elements	Ta	B	A1	Cu	0_2	N_2	Nb	As + Sn + Sb + Zr	Fe	Si	
wt.%	0.066	<0.0001	0.008	<0.002	0.008	0.031	<0.001	<0.03	Bal	0.05	

Table 2
Transformation temperatures of 9Cr–1.4W RAFM steel. [37].

Melting temp	Recrystallization temp (0.4 Tm)	A_{CI} (austenite start)	A_{C3} (austenite finish)	Martensite start M_s	Martensite finish M_f
1791 ± 5 K	716 ± 5 K	1091 ± 5 K	1130 ± 5 K	640 ± 5 K	583 ± 5 K
1518 ± 5 °C	443 ± 5 °C	818 ± 5 °C	857 ± 5 °C	367 ± 5 °C	310 ± 5 °C

alloy content and the tempering conditions. It has been found that W exerts a greater influence in the diffusion of substitutional elements both during precipitation and dissolution [14,16,17].

Fabrication of TBM for ITER under construction necessitates the usage of fusion welding processes. The various welding methods that are considered for TBM fabrication include tungsten inert gas welding (TIG), Narrow Gap-TIG, Laser, and Electron Beam Welding (EBW) [18–19]. The high heat input Shielded Metal Arc (SMA) and TIG welding processes, during the weld thermal cycle, promote a wider heat affected zone (HAZ) and generate an inhomogeneous microstructure in the HAZ of ferritic–martensitic steels, resulting in a marked variation in mechanical properties across the weld joint. Therefore, the full performance of 9–12% ferritic–martensitic steels has often not been realised in service due to the premature failures at the parent metal/HAZ interface of weld joints. These failures are commonly referred to as type IV cracking and have been reported to occur due to the pronounced localization of creep deformation coupled with preferential creep cavitation in the soft intercritical region of HAZ [20–22]. Strength reduction in the intercritical HAZ was associated with the combined effects of coarsening of precipitates and dislocation substructure [20–22]. The weld joints of RAFM steels fabricated by SMA and TIG welding processes would also be expected to undergo type IV cracking failures.

Type IV cracking could be minimised to a certain extent by using low heat input and high speed welding processes such as EBW and laser welding that produce a smaller HAZ. However, it is a very difficult task to eliminate the occurrence of δ- ferrite that nucleates during the solidification of liquid metal generated by conventional and advanced fusion welding processes; the δ-ferrite deteriorates the mechanical properties at elevated temperatures [23–25]. Since there is no liquid metal formation and subsequent solidification in the solid state friction stir welding (FSW) process, no δ-ferrite could be expected in the weld. FSW is also beneficial in narrowing down the HAZ of RAFM steel.

In FSW process a rotating tool containing the shoulder and pin is plunged into the joint between the two flat plates which generate heat due to friction and plastic deformation [26–28]. The work piece is softened around the tool and material is transported from advancing side to retreating side. The major parameters that control the heat input in FSW are tool rotational speed, tool traverse speed, axial force, shape of tool and tool material. Out of these, the variation in tool rotational speed is the major parameter that determines the peak temperature and cooling rate achieved during FSW. The peak temperature experienced during FSW is much lower than the melting temperature of the materials being welded, hence no melting and solidification related defects would be expected in the weld metal. Since no fusion welding is involved, much of the thermal contraction associated with liquid metal solidification and cooling that promotes distortion could be significantly reduced. The wide usage of FSW in the fabrication of steels was hindered by non-availability of a suitable tool material. In recent years, remarkable success has been achieved in FSW of low alloy steel grade DH36 used for ship building [29,30], maraging steel for aerospace applications [31] and AISI 430 ferritic stainless steel [32].

There have been few investigations exploring the possibility of using FSW in case of conventional ferritic–martensitic and RAFM steels [33–36]. In general, these investigations have revealed that the

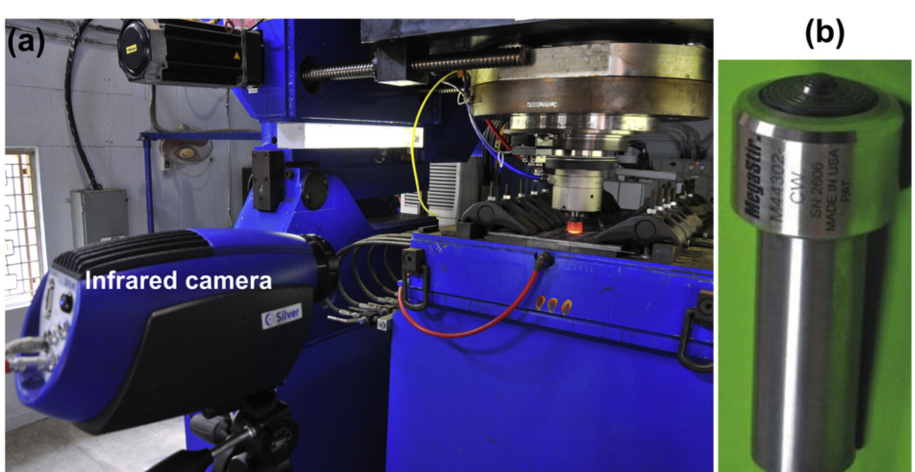

Fig. 1. (a) FSW experimental set-up with infrared camera arrangement for monitoring interface temperature. (b) Polycrystalline cubic boron nitride tool depicting shoulder and pin.