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Microstructural Evolutions of Friction Stir Welded F82H Steel for Fusion Applications

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1. Introduction

A blanket is the most important component functionalized as plasma confining, tritium breeding, heat exchanging, and irradiation shielding from severe thermo-neutron loads in a fusion reactor. Its structure consists of first walls, side walls, a back board, and coolant channels mainly made of reduced-activation ferritic/martensitic (RAFM) steel, which is the most promising candidate as a structural material for fusion reactors. To fabricate this blanket structure, some welding and joining methods have being carefully applied [1]. However, when fusion welding, such as tungsten inert-gas (TIG) welding, electron beam, and laser welding was performed between F82H and itself, the strength of welds significantly deteriorated due to the development of δ-ferrite and precipitate dissolution. Post welding heat treatment (PWHT) should be followed to restore the initial microstructure. Nevertheless, microstructural discontinuity inevitably occurs between the weld metal, heat affected zone and base metal and this seriously degrades the entire structural stability under pulsed operation at high temperature in test blanket module (TBM) [2]. A phase transformation can also be an issue to be solved, which leads to a difficult replacement of the blanket module. Therefore, a reliable and field-applicable joining technique should be developed not to accompany with PWHT after the joining process. Friction stir welding (FSW) is one of the solid-state processes that does not create a molten zone at the joining area, so the degradation of the featured microstructures may be avoided or minimized.

In this study, FSW was employed to join F82H steels to develop a potential joining technique for RAFM steel. The microstructural features on the joint region were investigated to evaluate the applicability of the FSW.

2. Methods and Results

2.1 Experimental procedure

The material used in this study was RAFM steel, F82H IAE Heat (Fe(bal.)-8Cr-2WVTa in wt%). Butt-FSW was conducted using rectangular plates with dimensions of 220mm in length, 100mm in width and 1.5mm in thickness. The material of the tool used in this study is WC-Co alloy. The tool rotating clock-wise is plunged into the center of two plates and moved along welding direction. The side where the direction of the rotation is the same as that of the tool traveling is called the 'advancing side' (AS), and the other is the 'retreating side' (RS). F82H steel was welded at a tool rotating speed of 100-400rpm and tool traveling speed of 100mm/min. The rotating tool was tilted 3° from the plate normal welding direction with 3 tons of load.

The specimens for microstructural observation were extracted across the weld joints by electro-discharged machining (EDM). The joints were mechanically wet ground and finally polished with colloidal silica. The grain morphology was observed by FE-SEM. To investigate the mechanical property differences between the base material (BM) and stirred zone (SZ), a hardness test was carried out. The hardness profile was measured on a transverse cross section, using a Vickers micro-hardness tester with a 9.8N load for 15 s at a regular interval of 1 mm.

Cross-sectional TEM thin films were carefully sampled by a focused ion beam (FIB) technique. TEM observation was conducted with a JEOL-2100FS electron microscope with an acceleration voltage of 200kV.

2.2 Microstructure of friction stir welded F82H steel

A F82H thin plate of 1.5mm thickness was successfully friction stir welded at the tool rotating speed from 100 to 400 rpm with a traveling speed of 100mm/min. An optical micrograph of friction stir welded F82H steel on transverse cross section is shown in Fig. 1. Two distinct zones can be clearly identified as the base metal (BM) and stirred zone (SZ), and their boundary could be called as heat affected zone (HAZ). The hardness variation usually reflects microstructural changes and a hardness line profile of friction stir welded F82H steel on the transverse cross-section was shown in Fig. 2. SZ of the tool rotating speed from 200 to 400rpm showed very high hardness above 400Hv, while the hardness of BM is around 220Hv. Interestingly, a similar hardness distribution was exhibited in the SZ of 100 rpm. These indicate that friction stir welding of F82H steel with 100rpm and 100mm/min was achieved under a phase transformation temperature of F82H steel, A_{C1}, which generates α/γ-

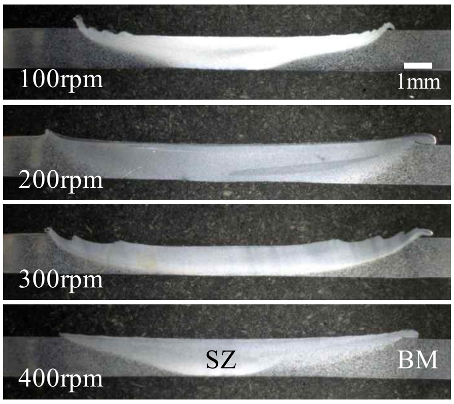

Fig. 1 Optical micrographs of friction stir welded F82H steels induced by different tool rotating speeds

phase transformation during the cooling process. The transformation temperature of this steel is known to be over 850^{o}C. Although the peak temperature was not measured during FSW in this study, this was actually measured by Y. D. Chung et al. in a dissimilar FSW between F82H and stainless steel 304, which was maximally reached at 839^{o}C [3].

BM has a typical tempered martensitic structure consisting of a martensite lath and carbide precipitation along the grain boundaries, as shown in Fig. 3. SZ induced by 100 rpm exhibits very fine and equiaxed ferritic grains under 1 μm, while severely deformed martensitic grains are entirely distributed in the SZ fabricated by a high rotating speed of 400 rpm. Meanwhile, deformed martensite and coarse ferrite partially co-existed in 200rpm stirred zone, as shown in Fig. 3(c).

To investigate the evolutions of precipitates during FSW, the precipitate distributions on each welding condition were observed by TEM with a carbon replica method. Fe and Cr- rich $M_{23}C_6$ precipitates are mainly laid on the prior austenite grain boundaries. Comparably fine precipitates including $M_{23}C_6$ and Ta, V- rich MX precipitates are also located at the lath boundaries. In a SZ of 100 and 400 rpm, on the other hand, uniformly distributed precipitates are observed, and it is assumed that the microstructures were simultaneously stirred during the FSW process by a tool pin. There is no

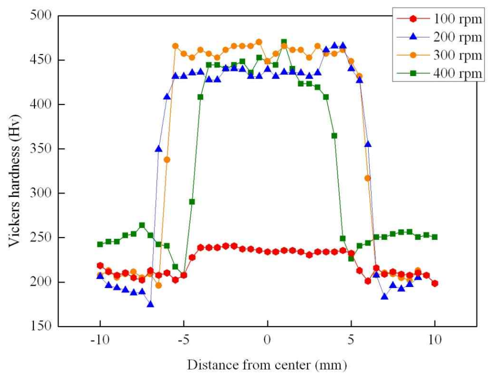

Fig. 2 Vickers hardness profiles of friction stir welded F82H steels

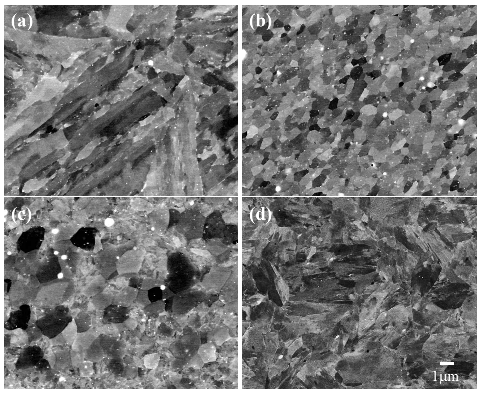

Fig. 3 SEM images on grain distribution of (a) BM, (b) 100, (c) 200 and (d) 400 rpm

evidence that the tool rotating speed affects the distribution and mean diameter of precipitates in F82H steel.

3. Conclusions

Microstructural evolution of friction stir welded F82H steel was investigated to develop a potential joining technique for RAFM steel. In the welding condition, 100rpm and 100mm/min, a stirred zone represented a comparable hardness distribution with the base metal. The base metal showed typical tempered martensite with precipitates on the PAG and lath boundary, whereas the stirred zone induced by 100rpm reserved uniformly distributed precipitates and very fine ferritic grains. Therefore, friction stir welding is considered to be a potential welding method to preserve the precipitates of F82H steel without PWHT.

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