

Vacuum Plasma Spraying W-coated Reduced Activation Structural Steels for Fusion Plasma Facing Components

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1. Introduction

Many fusion power plant concepts, such as water-cooled lead-lithium (WCLL), a helium-cooled ceramics/beryllium pebble bed (HCPB), and dual-coolant (DC) blanket systems, joining technologies of dissimilar materials are essentially required [1,2]. Tungsten (W) and its alloys are considered as candidate materials for plasma facing materials of the first wall and divertor components in fusion reactor systems because of high sputtering resistance and low tritium retention in a fusion environment [3]. Therefore, it is considered that the joining between W and reduced activation structural steels, and its evaluation, are critical issues for the development of fusion reactors. However, the joining between these materials is a very challenging process because of significant differences in their physical properties, particularly the mismatch of coefficients of thermal expansion (CTE). For instance, the CTE of pure W is known to be about $4.3 \times 10^{-6} \text{K}^{-1}$; however, that of martensitic steels reaches over three times, about $12\text{-}14 \times 10^{-6} \text{K}^{-1}$ at room temperature even up to 373K. Nevertheless, several joining techniques have been developed for joining between W and structural steels, such as a vapor deposition method, brazing and diffusion bonding [4,5]. Meanwhile, vacuum plasma spraying (VPS) is supposed to be one of the prospective methods to fabricate a sufficient W layer on the steel substrates because of the coating of a large area with a relatively high fabricating rate.

In this study, the VPS method of W powders on reduced activation steels was employed, and its microstructure and hardness distribution were investigated.

2. Methods and Results

2.1 Experimental procedure

The substrate materials used in this study were ferritic/martensitic steel, F82H and ODS ferritic steels. The ODS ferritic steels were fabricated by mechanical alloying and a hot extrusion process. Before the VPS process, contaminated and oxidized surfaces of the substrate materials were mechanically peeled using a sand-blasting process with Al_2O_3 powder. The materials were annealed at 823K, and pure W powder was then thermally sprayed onto the substrate materials by a plasma jet of Ar and H_2 in a vacuum chamber. The

thickness of the joint materials was evaluated to be 3.8 mm for the substrate material, and 0.7 mm for the VPS-W layer. The interfacial microstructure was analyzed using a field-emission type scanning electron microscope (FE-SEM) and electron probe micro analyzer (EPMA). The joints were mechanically wet-grounded and finally polished with colloidal silica to remove the work-hardened surface induced by the mechanical grinding. Neutron irradiation was also conducted in a high flux isotope reactor (HFIR) from Oak Ridge National Laboratory (ORNL) at 773K to 4.5 and 9.6 dpa. After irradiation, the hardness distribution in the vicinity of the cross-sectional joint was measured using a micro Vickers hardness tester installed in a hot cell with 0.98 and 9.87N for 15 seconds at room temperature in regular intervals of 100 μm and 0.5mm, respectively.

2.2 Interfacial microstructures of VPS-W coated joints

Various characteristics of W including physical, thermal and mechanical properties significantly depend on its microstructures with grain and defect distributions. The microstructures can also be varied with its fabrication process, such as melting-solidification, powder sintering process, and VPS. Thus, it is essential to exam the microstructural features of VPS-W and compare them with other tungsten materials. In Fig. 1, backscattered electron images (BEI) of commercial bulk pure W fabricated by conventional melting-solidification process and VPS-W in this study are shown. Bulk W has recrystallized and equiaxed grains with a compound of fine and coarse grains without micro-defects. On the contrary, VPS-W has fine grains in a high aspect ratio and parallel direction of thermal spraying with heterogeneously distributed large particles which were not to be sufficiently melted during the VPS

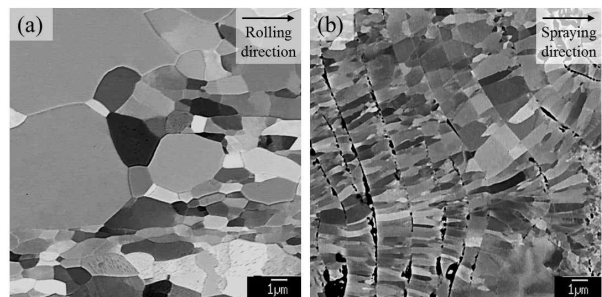


Fig. 1 SEM images of (a) commercial bulk pure W and (b) VPS-W

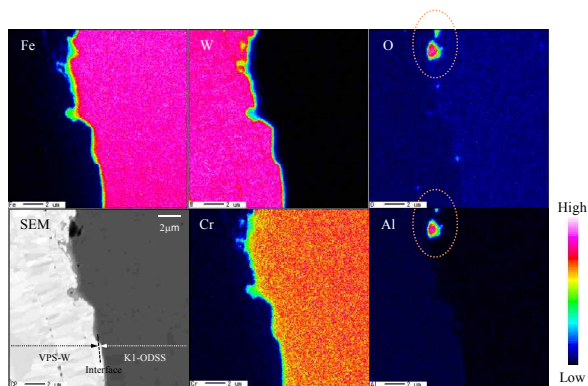


Fig. 3 Elemental distributions of VPS-W coated K1-ODS steel

process. Some interfacial voids and crack-type defects between lamellar layers were also found in VPS-W, owing to repeated coatings in each spraying-solidification cyclic process. These defects are known to be one cause of lower thermal properties than those of bulk W. Nagasaka et al. were reported that the thermo-mechanical property degradation of VPS-W occurred owing to crack-type voids and impurities [6]. Elemental distribution maps of VPS-W coated K1-ODS steel are represented in Fig. 2. The coating interface between W and the substrate is vertically located in the center of the maps. Al oxides were observed at the interface, which is considered to be resided media, an Al_2O_3 powder from sand-blasting process before VPS. This no doubt degrades the thermo-mechanical properties, and therefore alternative media or a removal treatment should be further examined. However, there is no mass diffusion area or intermediate layer of constituent elements, such as W, Fe, and Cr. For joining between dissimilar materials, an intermetallic compound layer by mass diffusion mainly contributes to a reduction in the thermal and mechanical property degradations of the joints [5]. In this respect, the VPS-W process is one of the potential ways to join between W and structural steels with a favorable interfacial property.

2.3 Hardness distributions of VPS-W coated joints

Vickers hardness distributions for VPS-W coated specimens are shown in Fig. 3. VPS-W showed about 310Hv in all substrates. It has a relatively lower hardness, about 60%, than that of commercial pure bulk W fabricated via melting-solidification, which is about 500Hv. Microstructural features of VPS-W, such as many interfacial voids and cracks, result in this low hardness, while it has a more homogeneous distribution of fine grains. Because an intermediate diffused layer does not exist at the coating interface between the W and substrates, hardness changes reflect the substrate properties with their different chemical compositions. ODS steel has a high hardness of about 320-400Hv owing to complex oxides with very fine diameter (<10nm) and fine grains (<3 μm). F82H steel is

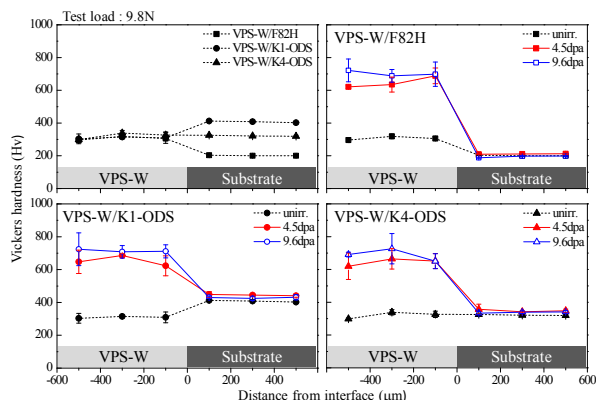


Fig. 2 Vickers hardness distributions for VPS-W coated tempered martensitic steel with coarse M_{23}C_6 and MX type precipitates, and thus presents the lowest hardness, at about 200Hv. The three substrate materials consequently showed uniform hardness distributions in an adjacent interface.

After neutron irradiation to 4.5dpa at 773K, the hardness distributions in VPS-W are comparatively uniform, and irradiation hardening significantly occurred as presented in Fig. 3. For more irradiation to 9.6dpa, hardness distribution has the same tendency with those of 4.5dpa. However, hardening is not very high, at about $\Delta 50\text{Hv}$, in spite of the double dose rate. This is because the dose rate, which influences the hardening of pure W, may be already or continuously saturated at a dose rate of 4.5 to 9.6 dpa, and it was reported that hardening is experimentally saturated between 1 and 10 dpa in an HFIR [7].

The relationship between irradiation hardening and dose rate is shown in Fig. 4. While a bulk W fabricated powder-sintering method shows significant hardening of about $\Delta 530\text{Hv}$ [7], VPS-W was not so high, at about $\Delta 360\text{Hv}$. VPS-W has more defects than bulk W, which seems to be associated by hydrogen and oxygen impurities. These possible defects may play an important role as a sink site during neutron irradiation. The substrate materials underwent nearly no irradiation hardening with very stable and uniform hardness distributions. It is well known that the irradiation defects of steels are diffused easily at high temperatures, as

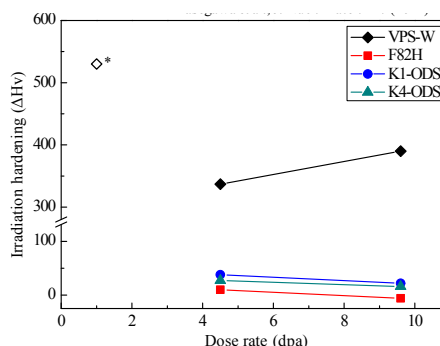


Fig. 4 Irradiation hardening of VPS-W and substrate materials as a function of dose rates

irradiation defect annealing occurs more effectively than hardening due to defect development on irradiation at around 773K.

3. Conclusions

ODS ferritic steels and F82H steel were coated by VPS-W, and the microstructure and hardness distribution were investigated. A microstructure analysis revealed that pure W was successfully coated on steel substrates by the VPS process without an intermediate layer, in spite of a mismatch of the CTE between dissimilar materials. After neutron irradiation, irradiation hardening significantly occurred in the VPS-W. However, the hardening of VPS-W was lesser than that of bulk W irradiated HFIR at 773K. Substrate materials, ODS ferritic steels, and F82H steel, did not show irradiation hardening because of the defect annealing effect at 773K.

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