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Variant selection in stationary shoulder friction stir welded Ti-6Al-4V alloy

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ABSTRACT

Stationary shoulder friction stir welding of Ti-6Al-4V of 7 mm thickness was conducted with varying welding speeds and rotation speeds. Variant selection analysis was carried out based on the inherited α phase texture and the reconstructed β phase texture. The weld surfaces became significantly smoother with increasing welding speed and decreasing rotation speed. Heat input decreased greatly with increased welding speed and it decreased slightly with decreased rotation speed. The orientation relationship between the prior β grains was measured based on the reconstructed electron backscattered diffraction (EBSD) data. Weak Variant selection has occurred in all the welds because most of the prior β grains did not share {110} poles. Strong links between crystal orientation of the prior β grains and hardness have been found.

1. Introduction

Friction stir welding (FSW) has been widely applied in the joining of aluminum alloys since it was invented in 1991 by The Welding Institute (TWI) [1]. As a solid state welding process, FSW uses a rotation tool plunged into the specimen and softened material is passed from the front of the tool to the back, where it consolidates to form a solid state bond. No melting is involved in this process, thus, many problems are eliminated or reduced, such as porosity, shrinkage and distortion, which are often found in conventional friction stir welding [2,3].

Stationary shoulder friction stir welding (SSFSW) was developed by TWI in 2004, compared to the conventional FSW, SSFSW uses a rotating probe located in a separate non rotating shoulder component, which slides over the surface of the specimen during welding. Thus, the shoulder does not directly contribute to the heat generated during welding. This produces a more uniform heat input through the weld thickness, eliminating the problem of surface overheating and it improves the process stability [4,5].

The weld zones of friction stir welded titanium alloys are usually divided into the base material (BM), the heat affected zone (HAZ), the thermo-mechanically affected zone (TMAZ) and the stir zone (SZ). A region adjacent to the BM is the HAZ which is influenced by the temperature but without deformation. Outside of the HAZ is a region called the TMAZ consisting of materials influenced by both deformation and temperature during FSW. The TMAZ is surrounded by the SZ and the high temperature β texture of the SZ has been reported to follow the simple shear texture of bcc materials [6–9]. To our knowledge, Davies et al. [7] and Pilchak et al. [8] are the first two researchers to have used a β reconstruction technique.

The transformation of the high-temperature parent β phase to the α_s phase of titanium alloy follows the Burgers orientation relationship, namely, (0001)α//{110}β and <11$\bar{2}$0>α//<111>β, which reflects the inheritance of the texture of the titanium alloy. According to the Burgers orientation relationship, the high-temperature parent β phase will produce 12 inherited α variants at room temperature. However, all variants do not often develop with the same probability due to material transformation conditions and variants are observed when the β phase is strongly deformed in the high temperature β phase field. This phenomenon is called variant selection due to deformation imposed on the material and severe deformation of the material in the β phase field [10,11].

Four possible mechanisms of variant selection in titanium alloys include:

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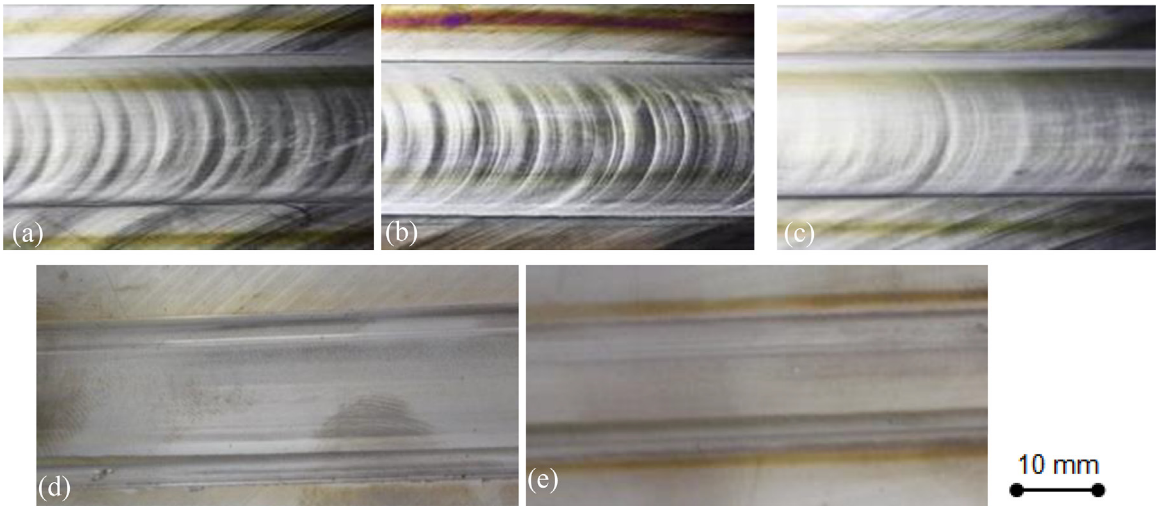

Fig. 1. Photographs of the weld surface of (a) Weld A, (b) Weld B, (c) Weld C, (d) Weld D and (e) Weld E.

(I) Variant selection is caused by the residual stress, and the residual stress in the high-temperature phase will result in the distorted anisotropic lattice spacing, due to volume changes by phase transformation strain on heating and cooling [12,13]. Davies et al. [7] and Daymond et al. [14] also pointed out that variant selection was driven by thermal stresses which leads to the preferential growth of certain variants by cold work, as well as the phase transformation strain [8,14].

(II) Variant selection is caused by dislocation in the β phase field [15,16], in this case, the α variants nucleate on the dislocations within the coarse planar slip bands in the β matrix [15] during the nucleation stage of α variants [17].

(III) The β-α phase transformation strain contributes to the formation of specific variants [15,18]; Variant selection was caused by the favorite orientated α laths by self-accommodation of stresses [18].

(IV) The presence of specific grain boundaries between β grains that share a common $\{110\}\beta$ plane in the high-temperature β phase will lead to the nucleation of α variants [19–26]. The fourth mechanism is the best possible reason accounting for the strong variant selection during phase transformations for both titanium and zirconium alloys [26].

Texture and variant selection are used to understand and predict the mechanical properties of FSW joints. There have been a number of studies investigating texture evolution during FSW, but few of these studies have examined variant selection in the weld zones. Thus, in this work, variant selection was studied for Ti-6Al-4V welds produced by SSFSW using electron backscattered diffraction (EBSD) technique.

2. Experimental procedures

The as received material was a rolled Ti-6Al-4V plate with a chemical composition of Ti 88.6%, Al 6.44%, V 3.98% O 0.19%, Fe 0.14%. Stationary shoulder FSW Ti-6Al-4V plates with a thickness of 7 mm were produced by TWI Technology Center (Yorkshire) Ltd, UK. The tool shoulder was made of Si_3N_4 with a diameter of 17.8 mm. The tool probe was a parallel tool made of a tungsten-rhenium alloy (W-25% Re) 9.8 mm in diameter, 6.1 mm in length and 2 mm in radius at the end.

Five welds were produced under force control mode and argon atmosphere was supplied to prevent the oxidation of the workpiece by the oxygen and hydrogen from the ambient atmosphere. The three welds Weld A, Weld B and Weld C were performed with a rotation speed of 900 rpm and traverse speeds of 50, 100 and 150 mm/min, respectively; another three welds were made with 150 mm/min welding speed and rotation speeds of 800 rpm (Weld D), and 600 rpm (Weld E), respectively.

All the five welds were sectioned into the cross sectional specimens with a dimension of 30 mm × 7 mm × 5 mm. Specimens were mounted, ground and polished using standard procedures. EBSD investigations were carried out on an Oxford-HKL EBSD system attached within a Sirion FEGSEM microscope at 20 kV accelerating voltage and spot size of 5. EBSD data were represented by the orientation imaging maps (OIM) consisting of inverse pole figure coloring maps, band contrast maps and pole figures. Pole figures are the representation of texture.

3. Results

3.1. Surface profile and heat input

Fig. 1 shows photographs of the weld surfaces for all the five welds. All the welds had relatively flat surfaces, but, as shown in Fig. 1, there are rings associated with the rotation of the tool which are more defined at the slowest traverse speed. Surface profile measurements are shown in Fig. 3. From visual inspection it was quite clear that all the three welds with constant rotation speed had a flat surface with obvious rings which must have been associated with the rotation of the tool. It was also obvious that the surfaces became significantly smoother with increasing welding speed. Surface profile measurements confirmed this observation. This would suggest that the process stability increases with increasing welding speed. The data, however, also confirm a bigger variation in height of the weld surface from the retreating side (RS) to the advancing side (AS) for Weld C compared to Welds A and B. This possibly suggests that with increasing welding speed a greater down force is required than used in the current work to get ideal surface flatness. The width of the weld is 17.75 mm which is the shoulder diameter. Detailed analysis of the spacing of the surface rings showed that at the lowest welding speed their spacing was irregular whilst at the highest speed the spacing became more uniform.

The calculation of heat input (H.I.) is given in Eq. (1), and the result is shown in Fig. 2.

$$\text{H.I.} = \omega \cdot T/V = 2\pi \cdot r \cdot T/(60 * V)(\text{kJ/mm}) \quad (1)$$