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Modelling of friction stir welding of 7xxx aluminium alloys

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Abstract

A combined modelling approach was applied to predict the behaviour of high strength 7000 aluminium alloys. Thermal modelling, microstructure modelling and strength modelling were performed in succession to give some insight into the complex precipitation mechanism occurring during friction stir welding (FSW). A quantitative assessment of a recent numerical model to predict the evolution of the precipitate distribution is performed for a high strength 7449 aluminium alloy subjected to a FSW process. An optimised model calibration procedure is also presented for the 7449 alloy. The robustness of this calibration is subsequently tested by applying the model to a different 7000 series alloy, 7150 in peak-aged condition, after FSW. Microstructure predictions are found to be highly dependent on the peak temperature reached during the weld thermal cycle as well as heating and cooling rates. A range of precipitation sequences involving metastable to equilibrium phase transformation and dissolution/coarsening of precipitates are predicted. The predicted microstructures are found to be in good quantitative agreement with the characterised experimental microstructures. Predicted precipitate distributions are used to estimate the strength of the material. These predictions generally agree well with measured hardness values.

Keywords: Friction stir welding; Numerical modelling; 7xxx AA

1. Introduction

Friction stir welding, a solid state joining method, has attracted significant interest from aircraft manufacturers for joining high strength aluminium alloy components that are presently riveted. This process, in which a rotating tool is used to mix material across the join line, subjecting it to both heat and deformation considerably improve the weldability of these alloys that exhibit poor performance after fusion welding processes. FSW of high strength age hardenable 7xxx aluminium alloys significantly modifies the precipitate distribution across the weld [1] compared to the starting condition. The size, distribution, and nature of the precipitates after welding depend on the local thermal cycle and extent of deformation which are both functions of the alloy, process conditions and position relative to the tool. In view of the complex interactions between these parameters it is clear that modelling the strengthening precipitate distribution evolution during FSW presents significant advantages in terms of welding process optimisation and mechanical property improvements whilst providing an understanding of the underlying precipitation mechanisms.

Semi-empirical process models to predict hardness profiles after FSW have been applied by Frigaard et al. [2] and Russell [3] to 6082 alloys, Hyoe et al. [4] and Sullivan et al. [5] to 7000 series alloys, and Shercliff et al. [6] to 2000 alloys.

However, whilst these models yielded reasonable hardness predictions for the FSW alloys, these models assume the change in hardness can be directly related to the dissolution of the strengthening precipitate and as such, are only applicable for alloys where the precipitate fraction is at a maximum prior to welding (T6 or T7 conditions). Also, since the models do not predict microstructure explicitly, it is not possible to use them to investigate how the particle distribution is modified, which is critical in determining more complex properties such as fracture toughness and corrosion resistance. More sophisticated models based on the Kampmann and Wagner numerical (KWN) framework [7] which models the evolution of precipitate distribution have been applied for welding of 6000 alloys [8,9] and 7000 alloys [10]. Recently, the authors have developed a model to predict precipitate evolution during FSW of commercial 7000 aluminium alloys [11]. As well as describing the precipitate evolution during FSW, two key development were made in this model [11]: (i) the metastable η' phase and equilibrium pre-

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Table 1
7449 and 7150 alloys nominal composition in wt%

	Zn	Mg	Cu	Zr	Fe	Si	Al
7449	7.5–8.7	1.8–2.7	1.4–2.1	<0.2	<0.15	<0.15	Balance
7150	5.7–6.7	1.8–2.7	1.5–2.3	<0.2	<0.15	<0.15	Balance

cipitate η phase populations competing for the available solute are considered as well as the $\eta' \rightarrow \eta$ transition, which allows a greater range of applicability of the model, (ii) grain boundary precipitate distribution and precipitate free zones (PFZ) sizes are explicitly calculated. Whilst the model does not consider the influence of the deformation imparted by the tool on precipitate evolution, it was shown in [11] that the model predicts qualitatively the expected precipitate evolution. The aim of the present work is to compare quantitatively the model predictions to the detailed microstructural characterisation of a friction stir welded 7449 alloy. Also a rigorous calibration procedure involving isothermal treatments and ramp heating is proposed. The range of applicability for modelling FSW of any similar 7000 alloy of the calibration procedure performed for a 7449 alloy is subsequently demonstrated by comparing the model prediction to the FSW of 7150 alloy. The precipitate distributions predicted by the model are also used to predict the precipitation strengthening following the Deschamps and Brechet approach [12] which is integrated in a yield strength model.

2. Experimental

2.1. Calibration experiments

Several experiments were performed combined with data taken from the literature to calibrate the different unknown physical parameters contained in the microstructural model. The choice of these experiments was made in order to calibrate the fitted parameters individually, i.e. reversion of 7xxx in T7 condition involves only the dissolution of η precipitates and since the solvus of η can be determined from thermodynamic calculations, the only unknown parameter to be calibrated is the interfacial energy of η.

SAXS experiments were carried out at the 2.1 beamline at the CCLRC Daresbury laboratory on 7449 alloy solutionised for 1 h at 480 °C, water quenched and subsequently aged at 120 °C for 10 min, 1 h, 4 h, 12 h and 48 h. All samples were kept below 0 °C between the heat treatment and the SAXS experiment to minimise natural ageing.

Experimental results from Nicolas [13] were also used to calibrate the model. Nicolas [13] performed some SAXS experiments on the D2AM beamline at the ESRF consisting of in-situ reversion treatment on 7449 alloy in T7 overaged condition. Reversion treatments were conducted at 200, 225, 250, 275 and 300 °C.

Differential scanning calorimetry (DSC) results from Kamp et al. [14] were also used for the calibration procedure. DSC disc samples of 5 mm diameter, 0.5 mm thickness were studied in a Shimadzu DSC-50 with a pure aluminium reference disc. A heating ramp of 10 K/min from room temperature to 500 °C was used.

Jominy end quench (JEQ) tests were carried out on cylindrical specimens, 100 mm in length and 25 mm in diameter, according to ASTM standard 255 using a Buehler Metaserv end quench unit to produce cooling rate varying from 4 to 150 K/s from the solutionising temperature. Cooling rates were measured using thermocouples situated 3, 10, 15, 20, 35, 50 and 70 mm from the end quenched. Specimens were prepared by grinding and polishing, and grain boundary precipitates and the PFZ were subsequently examined using a Philips XL30 FEG-SEM operated at 20 kV.

2.2. Instrumented welds

Friction stir welding was performed on two 7449 alloy 40 mm thick plates and a 7150 alloy 6.5 mm plate by Airbus UK, nominal composition for these alloys are given in Table 1. The 7449 welds have a 20 mm penetration thickness. Welding parameters are summarised in Table 2. The 7449 plate was initially in an underaged temper corresponding to the first step ageing treatment of a two-stage peak ageing treatment and is referred to as TAF (for age-forming temper). The 7150 alloy was initially in T6 condition. One 7449 weld was supplied as welded and naturally aged for several months and the other one was post weld heat treated (PWHT) to a T7 condition, i.e. the material was overaged after welding. In the following, the first 7449 weld is referred to as 7449TAF-FSW and the second as 7449TAF-PWHT. The temperature evolution for 7449TAF-FSW was recorded by eight thermocouples located at different positions (thermocouples positions are different for 7150FSW and 7449FSW).

Hardness measurements were performed across the 7150 weld near the surface and the bottom of the welded plates using a 20 kg loaded Vickers indenter. Similarly for the 7449TAF-FSW Vickers hardness measurements were performed across the weld at 2 and 10 mm depth with a 5 kg loaded Vickers indenter. Vickers micro-hardness measurements were performed across the weld at 0.5 mm, 5 mm and 10 mm depth for the 7449TAF-PWHT weld and at top and bottom of the weld for 7150T6-FSW using an Instron Wilson Tukon with a 1 kg load.

Grain boundary precipitates populations in the different weld zones were characterized in the 7449 welds using FEG-SEM. The fine precipitates in the same welds were characterised using a Philips CM200 TEM, operated at 200 kV. TEM samples were

Table 2
Welding parameters

	Tool/pin diameter/shoulder diameter (mm)	Advance speed (mm/min)	Rotation speed (rpm)	Weld thickness (mm)
7449 FSW	Triflat/17/34	95	215	20
7150 FSW	Threaded/8/25	95	215	6.5