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Micromechanical aspects of fatigue in a MIG welded aluminium airframe alloy Part 1. Microstructural characterization

F. Lefebvre [a,*], S. Ganguly [b], I. Sinclair [a]

[a] *Materials Research Group, School of Engineering Sciences, University of Southampton, Highfield Campus, Southampton SO17 1BJ, UK*

[b] *Department of Materials Engineering, The Open University, Walton Hall, Milton Keynes MK7 6AA, UK*

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Abstract

Fusion welding of high strength aluminium alloys is being widely considered within the commercial aircraft industry as an alternative to established mechanical fastening methods. The objective of the present article is to characterise the microstructural features of metal inert gas (MIG) welded plate of the conventional damage tolerant alloy, 2024-T351. Micromechanical aspects of the fatigue performance of this welded material are then addressed in a companion article. The general microstructure of the MIG weld was studied using optical microscopy and scanning electron microscopy, in association with electron back-scattered diffraction mapping. Columnar dendritic structures at the edge of the fusion zone are seen, with fairly uniform equiaxed dendritic grain structure dominating in the weld centre. Local microstructural conditions of the different elements of the weld have been assessed via micro-hardness and differential scanning calorimetry, identifying the balance between aging, overaging, re-solutionising and re-precipitation occurring across the weld region. Residual stress measurements by laboratory X-ray and synchrotron X-ray diffraction are also discussed, showing stress relaxation and redistribution occurring under loads representative of smooth specimen fatigue testing. The fusion zone is seen to present a tensile residual stress field, with peak longitudinal stress occurring towards the interface with the fusion zone.

Keywords: Aluminium alloys; Welding; Microstructure; Hardness; Residual stress

1. Introduction

The aircraft industry has historically focused on weight saving, high performance materials: increased economic pressures have however placed increasing emphasis on manufacturing costs and efficiency in recent years. As such, the use of welding as a cost effective alternative to mechanical fastening is being widely considered within the civil aerospace sector, with various reports within the literature concentrating on process development to achieve high quality welds in the high strength 2xxx and 7xxx heat treatable alloys used in aircraft construction [1–3]. The use of welded airframe structures has a number of important implications of course, such as: (i) changes in material microstructure, with dramatic local variations arising in grain structure and precipitate distributions, (ii) formation of new distributions of crack initiating defects not present in conventional wrought airframe alloys, (iii) formation of soft and hard heat-affected zones adjacent to the welds, and (iv) formation of local and global residual stress fields. Whilst different processes of welding exist and continue to be improved, the present work particularly focuses on metal inert gas (MIG) welding of the established damage tolerant airframe alloy, AA2024-T351. In conjunction with the accompanying paper [4], a micromechanical model of fatigue properties of the welds is sought. Principal microstructural features are particularly assessed in this paper via optical and electron microscopy, hardness mapping and

* Corresponding author. Present address: Bat. Saint-Exupery, GEMPPM–INSA Lyon, 20 Av. Albert Einstein, 69621 Villeurbanne Cedex, France. Tel.: +33 472 43 63 81; fax: +33 472 43 85 39.

E-mail address: fabien.lefebvre@insa-lyon.fr (F. Lefebvre).

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differential scanning calorimetry, along with residual stress analysis via X-ray diffraction. Features close to the weld surface are of particular interest given the associated investigation of freely initiating short fatigue cracks [4].

2. Experimental procedure

MIG weld samples were prepared from 13 mm gauge plate of commercial 2024. The material was welded in the T351 condition (solutionised, stretched and naturally aged), parallel to the L-direction (i.e. rolling direction). Filler wire of 2319 alloy was used. Nominal chemical compositions of both alloys are shown in Table 1. The welding was carried out in the downhand (PA) position, in two passes with a travel speed of 450 mm/min. The current and voltage used in the MIG welding process were 268 A and 24.3 V, respectively [5]. All samples were left at room temperature for at least 1 month prior to investigation. Post welding, each plate was skimmed by 2.5 mm with respect to the parent plate surfaces, with no post-weld heat treatments being performed. Here the 'top' of the weld is identified with the surface of the second weld pass.

Micro-hardness measurements were carried out across the weld cross-sections using a standard Vickers micro-hardness machine. The applied load and time were 1 kg and 15 s, respectively. The machine (Matsuzawa MHT-1) was checked before each use on a calibrated test block.

Differential scanning calorimetry (DSC) measurements were carried out on a power-compensated calorimeter (Perkin-Elmer Pyris 1) at a heating rate of 10 °C/min in nitrogen shielding gas. Based on the hardness results, key locations from the weld centre to the parent plate were identified for study. Small slices of 0.5 mm thickness and ∼3 mm diameter in the L–S plane were extracted, i.e. sampling small, notionally uniform volumes of material from within the weld/HAZ cross-section.

Microstructural observations were carried by optical microscopy and by secondary electron (SEI) and back-scattered (BEI) microscopy using a Jeol JSM-8500 FEG-SEM operating at 15 kV. Electron backscatter diffraction (EBSD) was performed using an HKL Channel 5 system attached to the FEG-SEM. Electron back scattered diffraction (EBSD) specimen were first mechanically polished and then electropolished at 30 V for ∼5 s in a 10% nitric acid in methanol solution cooled to −30 °C. Automatic image analysis software (Carl Zeiss KS300) was used to characterise weld defect populations. Parameters such as area, centre of gravity, Feret diameters (largest and smallest diameters), circularity and orientations of pores were derived from binarised images.

Table 1
Nominal chemical compositions of 2024 and 2319 alloys, in wt.% [17]

Alloy	Cu	Mg	Mn	Si	Fe	V	Al
2024	3.7–4.5	1.2–1.5	0.15–0.80	<0.15	<0.20	–	Bal.
2319	5.8–6.8	<0.02	0.20–0.40	<0.2	<0.3	0.1	Bal.

Residual stress measurements were carried out by laboratory X-ray and synchrotron X-ray diffraction techniques on electro-polished fatigue test specimens before loading and after applying a surface stress of 270 MPa, as described in the accompanying paper [4]. Two experimental arrangements were used for the laboratory measurements: (i) (3 3 1) plane reflections, using a Cu X-ray tube with a penetration depth of ∼40 μm, and (ii) (3 1 1) plane reflections, using a Cr tube, offering a penetration depth of ∼12 μm. With the (3 3 1) reflection measurements, the diffraction angle corresponds to ∼78°, leading to high errors in the strain calculation. With the Cr Kα facility a higher diffraction angle is achieved using (3 1 1) or (2 2 2) reflections, with the (3 1 1) plane representing the better choice in FCC polycrystals (least anisotropy and best representation of the average macroscopic strain [6]). Synchrotron diffraction measurements were also carried out at the ESRF (European Synchrotron Research Facility) in Grenoble by The Open University. The effective gauge volume of these measurements covered a depth of ∼500 μm, centred at the depth of ∼650 μm from the weld surface.

3. Results and discussions

3.1. Micro-hardness and DSC assessment

Fig. 1a shows a two dimensional hardness map across the MIG weld, along with a single hardness trace from the top surface of the skimmed specimens (i.e. ∼2.5 mm below the parent surface) (see Fig. 1b). The central region of the weld exhibits a wasted/hourglass shape consistent with the double pass nature of the MIG weld. It is clear that this region (i.e. the fusion zone) presents a minimum in hardness, corresponding to approximately 70% of the parent plate. Two main hardness peaks in the HAZ may be noticed: one is close to the fusion line, with a hardness level somewhat lower than the parent plate, whilst the second is remote from the fusion line, with a hardness level greater than the parent metal (see arrows in Fig. 1b).

Fig. 2 highlights the hardness trace for the depth at which DSC samples were extracted (key positions are highlighted) with the corresponding DSC results being shown in Fig. 3. Three main effects may be identified in these thermograms: from 150 to 240 °C an endothermic peak may be attributed to the zone and/or cluster dissolution, between 240 and 330 °C an exothermic peak, B, attributable to the formation of metastable and/or equilibrium precipitates, particularly S'/S (Al_2CuMg). A broad endothermic band, C, above 350 °C, is identified as a progressive dissolution of previous precipitates [7,8]. Above 500 °C a sharp endothermic peak may also be identified and attributable to incipient melting [7,8]. A more detailed description of the results is as follows:

- Between position 6 and 5, the hardness increases (after an initial shallow 'dip' in hardness). The DSC thermo-