Synchrotron XRD study of residual stress in a shot peened Al/SiC_p composite

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Abstract

In the present study, residual strain profiles in shot peened specimens of 2124-T4 aluminium alloy matrix composite reinforced with 17vol% particulate silicon carbide (SiC_p) were measured by means of synchrotron-based diffraction using monochromatic, high energy X-ray beams. The stress state was considered in relation with the microstructural and morphological modifications induced in the material by shot peening. Strain-induced changes in the lattice parameters were deduced from diffraction measurements made by two detectors mounted in the horizontal and vertical diffraction planes, providing information on lattice strains in two nearly mutually perpendicular in-plane directions. On the basis of these data, residual strain and stress profiles through the specimen thickness were reconstructed for both phases (silicon carbide and aluminium alloy). Microstructural characterization was performed by means of optical and scanning electron microscopy (SEM), and particle distribution and hardness modification were identified. The effect of shot peening on the reinforcement and matrix stress states was evaluated. The findings are discussed in the context of process optimization for fatigue resistance improvement in aluminium alloy-based MMCs.

Keywords: residual stress, shot peening, MMC, X-rays diffraction

1. Introduction

In the process of controlled shot peening a component is blasted with shots, typically of steel, glass or ceramics. Due to progressive shot impacts, the multiple indentations subject the material to cyclic plastic loading. Outer layers experience an in-plane stretching plastic deformation, while the elastic subsurface try to retain its original shape thus generating compressive residual stress at the surface. Shot peening finds application where an improvement in fatigue behaviour is required, e.g. for gears, valves, crankshafts, springs, turbine blades etc. Benefits are derived both from the surface hardening and from the compressive residual stress near the surface. The process has its greatest effect at long life: for low cycle fatigue, the stresses are high enough to cause yielding, thus washing out residual stress. Beneficial pre-stresses could also be cancelled in the case of exposure to high temperatures.

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(approximately 250°C for steel and 120°C for aluminium) or overstressing [1]. Major drawback for fatigue performances is the increased surface roughness after peening, which could promote crack initiation. Thus, final surface grinding or polishing can be required. The improvement in fatigue life increases with the magnitude and depth of the introduced residual stress up to an optimum peening level. However, over-peening could induce cracks at the surface and worsen the fatigue performance. Shot peening could also be a useful tool for the modification of surface stress state after machining or welding [2].

Metal Matrix Composites are a promising class of structural materials in which a reinforcement phase is dispersed in a continuous metallic matrix aiming to reach better mechanical properties than the parent alloy. In particular, particle reinforced aluminium matrix composite (AMC’s) have been recently considered for automotive and aerospace applications [3]. AMC’s surface treatment processes have the potential to provide a further increase in their fatigue performances, and remain the subject of ongoing research.

The knowledge of residual stresses within engineering components is of crucial importance in predicting their fatigue life. Although predictive process modelling could give an estimation of the stress level, experimental evaluation is necessary for validation. The aim of the present study is to measure the residual stress profiles in a shot peened SiC-particle reinforced aluminium alloy matrix composite by means of high energy, monochromatic X-ray diffraction in relation with the morphological and microstructural modifications induced by the process.

2. Experimental procedure

2.1. Materials

The material considered in the present study was the AMC217xe aluminium matrix composite, comprised of a AA2124 matrix (nominal wt %: Cu – 3.8, Mn – 0.5, Mg – 1.4, Zn < 0.25, Al balance) reinforced with 17vol% of fine (<3μm) SiC particles, manufactured by Aerospace Metal Composites Ltd (Farnborough, UK). In this proprietary production route, fine aluminium and SiC powders were first mixed and de-gassed, then consolidated by hot isostatic pressing. The resulting billets were extruded and plates were finally T4 heat treated. 14mm cubic samples were shot peened on one surface by means of steel shots.

The interaction between matrix and finely dispersed reinforcement results in an improvement in stiffness ($E_{Al} = 73$GPa, $E_{SiC} = 415$GPa, $E_{comp} = 100$GPa) and strength ($\sigma_{Ycomp} = 420$MPa, $UTS_{comp} = 620$MPa) without an appreciable change in density ($\rho_{comp}=2.85$g cm$^{-3}$, $\rho_{Al}=2.78$g cm$^{-3}$), thus resulting in a significant increase in specific properties.

2.2. Synchrotron X-ray diffraction

Lattice parameter measurements were performed at the European Synchrotron Radiation Facility (ESRF, Grenoble, France) by means of a 60 keV X-Ray monochromatic 50x50μm$^2$ beam with sampling steps up to 0.1mm at the surface. The X-ray beam was incident along the x direction in Figure 1. Diffraction patterns (Debye-Scherrer rings) were collected by an area detector, and equivalent line profiles extracted for scattering in the horizontal and vertical diffraction planes, providing information on the lattice strains in the surface normal (z) and transverse (y) mutually perpendicular directions. The X-ray wavelength ($\lambda$) is expressed in terms of the Plank’s constant (h), the speed of light (c) and beam energy (E) as $\lambda=hc/E$. The lattice $d$-spacing is computed from Bragg’s equation, $\lambda=2d_{hkl}sin\theta_{hkl}$, where $d_{hkl}$ is the atomic interplanar distance between the family of planes with the $hkl$ Miller index, and $2\theta_{hkl}$ is the X-ray scattering angle. Comparing the measured values of $d_{hkl}$ with the unstressed (or reference) lattice spacing $d_{hkl}^0$, lattice strains are determined as $\varepsilon_{hkl} = (d_{hkl} - d_{hkl}^0)/d_{hkl}^0$.

![Fig. 1.(a) Synchrotron scanning line and reference system.](image)
Lattice spacing measurements were made for three sets of hkl planes for Al (111, 200, 311) and one for SiC (111), by assuming Silicon Carbide to have isotropic elastic behavior. In the absence of reference strain-free lattice sample and in order to isolate the effects of peening, unstressed lattice parameters were calculated by averaging $d_{hkl}$ values in the bulk of the specimens. The thickness of the specimens doesn’t allow one to assume a plane stress state, so equi-biaxial strain state was assumed. In the planes parallel to the peened surface, plastic strains after peening were thought to be isotropic in the (xy) planes [4]. Indeed, when a large number of shots impinge normally to the surface on a homogeneous isotropic material, the resulting plastic strain tensor can be assumed free from deviatoric components, so that $\varepsilon_{xx}^p = \varepsilon_{yy}^p = -\frac{\varepsilon_{zz}^p}{2}$.

Plastic deformation experienced by the outer layer causes its length $L_0$ to increase to $L_p = L_0 + \int \varepsilon_{xx}^p dx = L_0 + \delta L$. Since the bulk of the sample didn’t undergo any plastic strain, it needs to deform elastically to match the changed dimensions of the near surface layers. Hence, tension develops in the bulk while compression arises in the outer layers, thus giving rise to residual stresses within the sample. Since elastic residual strains and stresses arise in order to match the plastic deformation of the outer layers, $\varepsilon_{yy}$, surface residual strain, is assumed to be equal to $\varepsilon_{xx}$ due to the symmetry in all (xy) planes, thus allowing one to calculate the stresses near the surface by means of y- and z-direction measurements alone. According to this hypothesis, residual stresses in the composite were computed by means of Hooke’s law for a tri-axial stress state separately for the Al matrix and SiC reinforcement phases as:

$$\sigma_{xx} = \frac{E}{(1 + v)(1 - 2v)} \left[ (1 - v)\varepsilon_{xx} + v(\varepsilon_{yy} + \varepsilon_{zz}) \right]$$

(1)

Similar expressions apply for $\sigma_{yy}$ and $\sigma_{zz}$. Finally, macrostresses were computed by means of the rule-of-mixture:

$$\sigma_{\text{Macro}} = f\sigma_{\text{tot}} + (1 - f)\sigma_{\text{tot}}^{Al} = 0.17\sigma_{\text{tot}}^{SiC} + 0.83\sigma_{\text{tot}}^{Al}$$

(2)

2.3. Microstructural analysis

The peened specimens were studied by means of scanning electron microscopy (SEM), while polished cross sections where observed through optical microscopy (OM) in order to investigate the presence of cracks at or near the surface and particle distribution after peening. The magnitude and depth of peening process on hardness was evaluated through Vickers hardness profiling along the z-axis.

3. Results and discussion

Fig. 2 shows the residual stress profiles computed using the equations given above for the y (transverse) and z (surface normal) directions for the Al and SiC phases, together with the residual macrostress in the composite, as a function of the distance from the peened sample surface (z).

![Fig. 2](image-url) (a) y-residual stress and (b) z-residual stress profiles as a function of the distance from the peened sample surface (z).
Maximum compressive longitudinal residual stresses are reached in the Al phase at about 0.3 mm from the surface, while tensile stress of low magnitude arises at approximately 0.6 mm depth. Interestingly, strongly tensile residual stresses are induced in the SiC phase both for the y- and z-directions, turning into compressive from about 1.5 mm depth. SiC particles probably retained tensile stress component state as they didn’t undergo any plastic deformation while the outer layers where stretched in the xy plane during peening. Discrepancy in the stress state between the matrix and reinforcement particles in presence of strong plastic flows has been already observed in this kind of materials [5]. Further investigation would be needed to validate this hypothesis.

Hardness profiling showed hardness increase from 218 HV for the base material (average) up to 260 HV at 0.3 mm from the surface (Fig. 3a), coinciding with the maximum residual stress level in the Al phase. SEM micrograph showed a strongly deformed (Fig. 3b) outer layer with the presence of cracks, as seen in the cross section of Fig. 3c. It is here also noticeable in these micrographs that particle-free zones in the base material resulting from the extrusion process are washed out approaching the peened outer layers. This increase in the particle density, together with the presence of matrix compressive residual stress, contributed to the measured hardness increase.

4. Results and discussion

In the present study, residual stress analysis of shot peened SiCp reinforced Al matrix matrix composites was carried out by means of high energy synchrotron X-rays diffraction and discussed in relation with the microstructural modification induced in the material. Compressive residual stress were detected in the Al phase while tensile residual stress were found in the SiC reinforcement phase, as a consequence of the deformation mechanism during peening. Peak hardness is reached where Al compressive residual stress and particle density are highest. The presence of a slightly tensile macrostress and cracks at the surface could be avoided by optimization of the peening parameters, thus demonstrating how the process of controlled shot peening can be a valid tool to improve MMC performance.

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