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Influence of particle and short-fibre reinforcement on the very high cycle fatigue behaviour of aluminium matrix composites

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ABSTRACT

The fatigue behaviour of aluminium matrix composites (Al-MMCs) reinforced by alumina particles (15 vol% Al_2O_3) or short-fibres (20 vol% Saffil), respectively, was investigated in the very high cycle fatigue (VHCF) regime at stress amplitudes $\leq$140 MPa to reach fatigue lives of up to 10^{10} cycles using an ultrasonic fatigue testing device. Fatigue tests were complemented by *in situ* thermographic measurements to record the temperature of the whole specimen and to find "hot spots" indicating changes in microstructure and, therefore, the initiation or growth of cracks. Results obtained in the VHCF-region fit well with results obtained in LCF/HCF-region. A significant influence of test frequency was not found. Surprisingly, the reinforced Al-MMCs are capable to bear up to 10^9 cycles, although damage starts early during cyclic loading. In addition, although in the LCF/HCF regime the fatigue life and fatigue strength of the short-fibre reinforced Al-MMCs are significantly smaller compared to particle reinforced and unreinforced material, no significant differences to the particle-reinforced and unreinforced Al-MMC were observed at stresses below 140 MPa. Moreover, at stresses below 80 MPa not fatigue failure occurred at all up to 10^{10} cycles.

1. Introduction

Since many years, lightweight metal matrix composites (MMCs) are in the focus of scientific and industrial interest. Ceramic reinforcements in terms of particles, short-fibres or long fibres embedded in lightweight metal matrix (aluminium, magnesium) yield improvement of numerous properties, such as wear resistance, high temperature strength or Young's modulus, which are limited for the unreinforced lightweight metals [1–5]. Due to their good stiffness-density and strength-density ratios, respectively, lightweight MMCs are used in the automotive industry for components, which require excellent damping and stiffness properties such as pistons, break discs or power trains. Since these components are undertaken frequently strong cyclic and sometimes even high-frequent cyclic loading conditions, a good knowledge of the material behaviour during cyclic loading is essential for further improvement of material properties. Besides the physical and mechanical properties of lightweight MMCs, investigations on the damage behaviour of these materials under cyclic loading conditions are of great interest.

Today, there are numerous results concerning the fatigue behaviour of unreinforced aluminium alloys in the low-cycle fatigue (LCF), high-cycle fatigue (HCF) and very high cycle (VHCF) region [6–10]. At higher stress levels, the crack initiation starts at the surface of the specimens whereas in the VHCF crack initiation is observed at pores and inclusions [11–13].

The existing results for aluminium matrix composites reinforced with ceramic particles or short-fibres are not such manifold. Following remarks refers only to the LCF and HCF area. However, the strength properties of these materials are significantly improved due to the different mechanisms known from literature [14]: (i) interaction of dislocations of matrix material with the reinforcements, (ii) thermally-induced internal stresses in the two-phase material due to the manufacturing process and (iii) increased dislocation density resulting in obstruction of each other. Moreover, a grain refinement of the matrix is observed compared to the unreinforced alloys. The largest influence on the mechanical properties have the reinforcement/matrix interface and the microstructure of the composite [14]. Therefore, failure is initiated by (i) debonding of the matrix/reinforcement, (ii) fracture of the reinforcement or (iii) matrix failure.

Furthermore, Liu and Bathias [15] investigated the correlation between the level of stress amplitude and preferred damage mechanisms for aluminium matrix composites reinforced with 20 vol% of short-fibres and observed at: (i) $\sigma_a < 0.45\ R_m$ debonding of the matrix/fibres interface and low local plastic deformation, (ii) $\sigma_a < 0.6\ R_m$ low plastic deformation and individual broken fibres, and (iii) $\sigma_a > 0.6\ R_m$ high amount of broken fibres. Liu and Shang [16] mentioned sometimes also

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a "pull-out" effect in the case of poor bonding of fibre and matrix caused by the infiltration process [5]. It is also known from literature that crack initiation starts at about 10% of the lifetime due to broken fibres or particles while the crack initiation due to the debonding of fibre/matrix starts at 70–90% of the fatigue life [17,18]. This is also confirmed by Hartmann et al. for LCF and HCF [19,20].

Other factors influencing the fatigue life are (i) the size effect, (ii) the formation of clusters and (iii) the kind of reinforcement. Thus, the probability of damage increases with the reinforcement size [21]. Uematsu et al. [22] reported that the fatigue life is even higher if the SiC-particles show a diameter of 5 µm compared to 20 µm. Related observations were shown for a magnesium alloy [21]. A high amount of equally distributed particles increases the fatigue life [19,23,24]. Allison and Jones [25], Llorca et al. [26] as well as Hartmann [19,20] showed a preferred crack initiation at particle clusters due to higher stress concentrations in these zones in comparison to the surrounding matrix. A direct comparison of different reinforcements with the same matrix material was carried out only by Hartmann [19,20]. Hartmann found for strain controlled fatigue tests that aluminium reinforced with short-fibres shows lower lifetimes than the particle reinforced material. However, it was also shown that the fatigue live data of unreinforced reference alloy and particle-reinforced and short-fiber reinforced Al-MMCs (AA6061, peak aged condition) fall into a narrow scatter band if the maximum achieved stress amplitudes are plotted versus the number of cycles to failure. Although the fatigue behavior beyond 10^7 was not in the focus, a tendency became visible that the maximum achieved stress amplitudes converge in the regime of $N_f > 10^7$ for the three investigated materials suggesting a comparable fatigue strength of these materials in the VHCF regime. Ding et al. [27–29] developed models to predict the fatigue lives of these materials.

The results in the VHCF region carried out by Papakyriacou et al. [24] can be summarized as follows: (i) The Al-alloy (AA6061) reinforced with 12–21 vol% Al_2O_3 particles in T6 conditions shows 10 times lower fatigue life than the unreinforced aluminium alloy at $N < 5 \times 10^8$ cycles. (ii) The maximum amplitude for final failure drops from 145 MPa (AA6061) to 115 MPa (AA6061-15p), and (iii) the lower fatigue life is ascribed to the early initiation of cracks near to large broken particles. In particular, clusters of broken particles lead to an early crack initiation (iv), and (v) the failure at the particle/matrix interface is only rarely observed.

The main goal of the present paper is the investigation of the fatigue behaviour of particle and short-fibre reinforced aluminium (AA6061) MMCs in the VHCF region in comparison to the unreinforced material. The influence of the reinforcement geometry and matrix strength is discussed. The results of the VHCF region were correlated with the results of Hartmann [19,20] and compared to the unreinforced material state. Furthermore, first findings on the underlying damage mechanisms were carried out. For such experiments in the VHCF-region, a special testing equipment is necessary. The ultrasonic fatigue testing device was developed by BOKU Vienna. It is possible to investigate fully-reversed tension-compression loading conditions ($R = -1$). Besides the ultrasonic fatigue testing equipment, which allowed to record the resonance frequency and nonlinearity parameter as indicators for damage initiation, thermography was used for a better understanding of the underlying microstructural processes. In addition, a new method of full-surface view thermography was implemented for these experiments.

2. Experimental details

2.1. Material

The mechanical behaviour under cyclic loading conditions in the VHCF regime was investigated on the unreinforced aluminium alloy AA6061 and on reinforced AA6061 with either 15 vol% of alumina particles or 20 vol% of alumina short-fibres (Saffil). The particle reinforced material was manufactured by Duralcan™ process and had a rod diameter of 11 mm. Typical size of alumina particles used for the Duralcan™ process was about 8–12 µm. The short-fibre reinforced material was produced by gas pressure infiltration using ceramic preforms with dimensions of $148 \times 76 \times 14\,mm^3$ (Leichtmetall-Kompetenzzentrum Ranshofen, Austria). 3–5% of silica were used as binder for the short-fibre preform. The short-fibres had a random-planar arrangement in the preform. The chemical compositions of the investigated material conditions are given in Table 1.

Table 1
Chemical composition of the Al-alloy and the matrix of the Al-MMCs (wt.%).

State	Al	Mg	Si	Fe	Ti	Cr	Cu	Mn
AA6061	bal.	0.76	0.54	0.73	0.09	0.13	0.29	0.17
AA6061-15p	bal.	1.16	0.57	0.18	0.04	0	0.19	0.05
AA6061-20s	bal.	1.42	0.99	0.63	0.03	0	0.18	0.16

Figs. 1 and 2 show the microstructure of the particle reinforced Al-MMC and the short-fibre reinforced Al-MMC in the initial state. Besides optical micrographs perpendicular and parallel to the extrusion and infiltration direction, respectively, detailed SEM images on the particular reinforcements are shown.

The alumina particles had a slightly preferred orientation along the rod axis due to the extrusion process (Fig. 1a) and a mean particle diameter of about 11 µm (Fig. 1b). The mean grain diameter of the aluminium matrix was estimated at 14 µm (Fig. 1c). Fig. 1d and e show an individual alumina particle extracted by deep etching. Alongside with the spattered shape of the alumina particles, it becomes obvious that small crystallites are attached on the particle surface, which were identified as spinel ($MgAl_2O_4$), which led to a better bonding between particles and metal matrix [30].

The short-fibres had a mean diameter of 3 µm and a typical length of 200 µm (Fig. 2a). Fig. 2b shows the non-infiltrated preform of the material. The aluminium matrix is coarse-grained with an average grain diameter of about 258 µm (Fig. 2c). In addition, a secondary iron- and silicon-rich phase was observed quite frequently around the alumina fibres (Fig. 2d, bright phase marked by black arrows) as well as non-infiltrated regions between individual fibres are visible (Fig. 2e).

All three materials were tested in peak-aged (T6) condition. The parameters for the T6 heat treatment were chosen according to Hartmann et al. [19,20], which consisted of an annealing process at 560 °C for 0.5 h and water-quenching followed by an aging procedure at 165 °C and air-cooling. The aging time was different for the materials tested: (i) 24 h for the unreinforced state, (ii) 16 h for the particle reinforced material, and (iii) 8 h for the short-fibre reinforced material to reach the peak-aged condition. Furthermore, the short-fibre reinforced material were also tested in an overaged state, which means same annealing procedure and aging temperature, but the aging time was extended to 96 h.

The resulting specific material properties for the T6 conditions are summarized in Table 2. Thus, the Young's modulus (determined via ultrasonic measurements) and the tensile strength were increased significantly by reinforcements, whereas the ductility was significantly reduced for the short-fibre reinforced Al-MMCs. The mechanical properties determined for the investigated Al-MMCs are in good agreement with data reported in literature [31]. The slightly lower tensile strength of the short-fibre Al-MMC in comparison to the particle Al-MMC was caused by the significantly lower ductility and was, therefore, attributed to the lower strain hardening capability.

No significant differences of the mechanical properties were expected in different directions since the arrangement of the short-fibres was random planar. However, the tensile tests revealed differences at room temperature and moderate strain rates. Table 3 shows the mechanical properties of the material sampled either parallel (L) or transverse (T) to the infiltration direction (compare also Fig. 2a). The