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Enhanced fatigue lives in AA1050A/AA5005 laminated metal composites produced by accumulative roll bonding

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In this work laminated metal composites with alternating layers of commercial pure aluminum AA1050A and aluminum alloy AA5005 were produced by ARB. In order to vary the layer thickness and the number of interfaces different numbers of ARB cycles (4, 8 and 12) were applied. Subsequently, fatigue tests were performed and the related deformation and damage mechanisms were investigated. At high amplitudes crack growth occurs very straight through the whole sheet and is more or less unaffected by the interfaces of the individual layers in the composite. At lower amplitudes, where pronounced grain coarsening of the AA1050A layers occurs, the cracks deflect in front of the AA5005 layers and propagate in the coarsened AA1050A layers parallel to the bonding plane and the loading direction. Contrary to the composite, in both conventionally ARB processed AA1050A and AA5005 mono-material sheets grains are only locally coarsened around the crack tip and the fatigue cracks propagate very straight through the sheet width. Compared to AA1050A mono-material ARB sheets fatigue life and the cyclic stability are clearly enhanced in all composites.

1. Introduction

It is well known that ultrafine-grained (UFG) materials exhibit significantly enhanced fatigue lives compared to their coarse-grained (CG) counterparts in a Wöhler-SN-plot, e.g. Refs. [1,2]. Concerning laminated metallic composites (LMCs) only limited data on the fatigue behavior is available in literature [3,4]. In that context it has to be mentioned that LMCs can be designated as one of the oldest composite materials made by mankind as they are already mentioned by Homer's Ilias [5]. Furthermore LMCs have been intensively investigated in the last century, see e.g. the review of Lesuer et al. [6]. Although LMCs are a rather “antique” class of materials the combination of the accumulative roll bonding (ARB) process, which leads to an UFG microstructure, together with the well-known laminated structures opens new possibilities for lightweight material designs, e.g. Ref. [7].

As cyclic properties are regarded no data is available in literature for laminated UFG materials up to now. However, several investigations on the cyclic deformation behavior and fatigue life of UFG mono-material materials exist. Due to the small grain size in UFG materials the development of dislocation arrangements/structures, which typically accommodate the strain during fatigue in CG materials, is not possible [8]. Instead Hashimoto et al. [9] and Höppel et al. [10] reported pronounced grain coarsening during cyclic loading of ECAP-processed UFG copper. Moreover, Mughrabi et al. [11] suggested that micro shearbands form either as a consequence of local grain coarsening or due to the strain path change related to ECAP and subsequent cyclic push-pull loading. Very similar observations on grain coarsening were made by Kwan and Wang [12,13] during the fatigue of UFG copper produced by ARB. In all cases grain coarsening leads to pronounced cyclic softening. The extent of grain coarsening and accordingly of the cyclic softening is significantly influenced by the stability of the microstructure [14,15]. With the increasing amount of foreign elements or alloying components, a transition from cyclic softening to cyclic hardening or at least an equilibrium is observed. A small amount of impurities is sufficient to stabilize the microstructure and to suppress grain coarsening. For example UFG copper samples with a purity of 99.99% show cyclic softening [10] whereas cyclic hardening was observed in samples consisting of a purity of 99.8% at the same load range [14]. Similar behavior was reported on the Al–Mg-System by May [16]. It is important to distinguish between low and high cycle fatigue (LCF/HCF) comparing the fatigue lives of UFG and

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CG materials in a total strain fatigue life diagram. In the HCF regime the fatigue lives are mainly influenced by the strength of the material and thus the UFG samples show higher fatigue lives and fatigue limits compared to their CG counterparts [1,2,17–19]. However in the LCF regime a cross-over of the fatigue life curves of the CG and UFG conditions can be found [20]. At high loads UFG samples reveal shorter lives because of a reduced ductility, a lower cyclic stress intensity threshold [21–23] and an easier intergranular crack growth [18,24].

The fatigue life and crack propagation of LMCs was studied by different authors in the last decades on CG samples. Most of the experiments were performed on thick sheets with only a small number of interlayers. Wittenauer and Sherby [3] showed that the fatigue life of ultrahigh carbon steel (UHCS) laminates containing thin Cu layers is significantly higher compared to the UHCS mono-material reference sheets [3]. The Cu layer plastically deforms during fatigue leading to a delamination of the Cu layer and thus crack blunting. Additionally, it was noted, that failure of the first UHCS layer in the laminate occurs at the same number of cycles as failure of the UHCS reference. Consequently, the fatigue life enhancement of the laminated structure is more significant at higher stress amplitudes, at which the fatigue life is dominated by fatigue crack growth [3]. In the high cycle fatigue regime the fatigue life is primarily governed by the number of cycles to initiate a crack and the fatigue life of the laminate was found to be very similar to the behavior of the reference sheet [3]. In contrast to that, UHCS laminates containing Fe–3Si layers presented no delamination during fatigue testing resulting in fatigue lives equally to the UHCS reference sheets at high and low loads [3]. Detailed investigations on the crack propagation in laminated metal composites were performed by Pippan et al. [4]. They found that, if the fatigue crack approaches the interface from the plastically weaker to a plastically stronger material, the crack propagation rate diminishes in the vicinity of the interface and the crack bifurcates. This behavior was attributed to a change of the near crack tip deformation field. Otherwise, as the crack approaches the interface from the plastically stronger material, the crack propagation rate accelerates [4]. The same behavior was observed in interlayer systems as well, only less prominent [4]. Kolednic et al. [25] investigated the crack propagation of LMCs with highly different Young's moduli. Likewise to the laminates exhibiting a high hardness difference at the interface, the crack driving force strongly decreases at the interface leading to crack arrest if the tip is located in the region with a lower Young's modulus and consequently a strong increase of the fracture resistance can be found.

LMCs produced by ARB offer the opportunity to further enhance the fatigue properties with their ultrafine-grained microstructure and the possibility to optimize the layer architecture. In this paper the influence of the layer thickness and the numbers of interfaces on the fatigue properties of the composites was investigated by applying different numbers of ARB cycles (4, 8 and 12). The aim of this paper is to understand in detail how the interfaces, which can be designed in ARB processed LMCs together with the UFG-structure, affect the cyclic deformation behavior and the fatigue life.

2. Experimental

2.1. Material processing and initial layer properties

Laminated metal composites consisting of AA1050A (Al 99.5) and AA5005 (AlMg1) were produced with up to 12 ARB cycles. A detailed description of the process is given in Ref. [26]. In order to prevent necking and rupture of the harder AA5005 layer at higher numbers of ARB cycles the N8 and N12 sheets were recrystallized for 10 min at 500 °C and then cooled in air after 4 ARB cycles (N8) or, respectively, after 4 and 8 ARB cycles (N12). The number of interfaces and the layer thickness of the processed composites are given in Table 1. As it is described in a previous paper, see Refs. [26], the AA1050A and AA5005 layers can be clearly identified visually in the N4 and N8 composites by a much finer grain size and mechanically by a higher hardness in the AA5005 layer. The grain size and the mechanical properties of both materials in the composite only slightly increase from ARB cycle 4 to 8. In contrast to that, in the N12 composite the material contrast at the layer interface disappears. The N12 composite exists of a more or less uniform grain structure and hardness across the whole sheet thickness, which are both similar to the ones of the AA5005 layers in the N4 composite, cf. [26]. This can be explained by additional grain subdivision due to the high number of interfaces as well as by stronger pronounced diffusion of solute atoms during the intermediate heat treatments at higher numbers of ARB cycles. During one heat treatment step the calculated diffusion path of solute Mg atoms is around 10 μm. As the layer thickness reaches 8 μm after 8 ARB cycles and 0.5 μm for N12, the Mg distribution is supposed to become more and more homogenous across the layers with increasing number of ARB cycles. Consequently, the material contrast at the interface diminishes increasingly and grain refinement in the AA1050A layer becomes more pronounced during the ARB cycles 9–12. Thus, for N12 composite a similar grain size and hardness in both layers were observed. For reference, conventional ARB processed AA1050A and AA5005 mono-material sheets with 4 ARB cycles were produced. Comparing the AA1050 and AA5005 mono-material sheets with the N4 and N8 composites, the grains of both materials are slightly smaller in the composite due to the additional shear at the layer interfaces [26]. Nevertheless the hardness of the layers in the composites is very similar to those of the mono-material sheets.

2.2. Experimental procedure

Flat test specimens with a gauge length of 10 mm, a width of 4 mm and a thickness of 1 mm were machined out of the sheets parallel to the rolling direction. In the gauge section the specimens were mechanically ground down to a grit size of 6 μm. Fatigue experiments were performed on the servo-hydraulic testing system (MTS810, MTS System Corporation, Eden Prairie, USA) at room temperature. Most of the experiments were performed under total strain control with a constant strain rate of $2 \cdot 10^{-3}$ s^{-1}. The total strain amplitude $\Delta\varepsilon_{tot}/2$ ranged between $0.75 \cdot 10^{-3}$ and $2.00 \cdot 10^{-3}$ by a mean strain ε_m between $1.25 \cdot 10^{-3}$ and $2.5 \cdot 10^{-3}$ resulting in R-values $R = \varepsilon_{min}/\varepsilon_{max}$ of 0.1–0.25. Since buckling of the specimens occurred at higher strain amplitudes additional stress controlled fatigue experiments were performed. The stress amplitude of these experiments was set according to the stress amplitude at the beginning of the strain controlled tests on the N4 composites. Due to the susceptibility to buckling of the AA1050A sheets only stress controlled fatigue experiments were performed for this material. The minimum stress was kept at 5 MPa resulting in R-values $\sigma_{min}/\sigma_{max}$ of 0.03–0.04.

After fatigue exposure the crack path and the grain structure were analyzed by scanning electron microscopy (Crossbeam 1540 EsB, Zeiss, Oberkochen, Germany) in secondary electron and backscattered electron contrast techniques, respectively. All sheets

Table 1
Number of interfaces and the calculated layer thickness of the ARB processes LMCs.

ARB cycle no.	N	4	8	12
No. of interfaces	$(2^N/2)$	8	128	2048
Calc. layer thickness	$(1\ mm \cdot 2/2^N)$	125.0 μm	8.0 μm	0.5 μm