Wear 304 (2013) 162–172

Contents lists available at SciVerse ScienceDirect

Wear

journal homepage: www.elsevier.com/locate/wear

Minimizing tribolayer damage by strength–ductility matching in dual-scale structured Al–Sn alloys: A mechanism for improving wear performance

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ARTICLE INFO

Article history:
Received 22 February 2013
Received in revised form
28 April 2013
Accepted 3 May 2013
Available online 22 May 2013

Keywords:
Sliding wear
Contact mechanics
Tribochemistry
Bearings

ABSTRACT

A mechanism for significant improvement of wear properties has been investigated. This operates by inducing a dual-scale structure consisting of coarse-grains (CG) and ultrafine-grains (UFG) in Al–12 wt% Sn alloys, in comparison with uniform UFG or CG structured alloys. It has been found that a dynamic steady tribolayer consisting of fine crystalline oxides plays a dominant role in improving the wear properties of both the UFG and dual-scale alloys. For the CG alloys, poor wear properties, caused by delamination wear of the tribolayer, could not be maintained on the worn surface. The tribolayer is formed on the worn surface by compacting and tribo-sintering of fine loose particles produced by sliding wear. However, the damage of the dynamic steady tribolayer is governed by the matching between ductility and strength of the matrix. The low ductility of the UFG alloy substrate causes the tribolayer to suffer crack damage rather easily, which limits any further improvement of its wear resistance. In contrast, for the CG alloys, the tribolayer is broken up and extruded into the matrix as a result of its low strength and a stable tribolayer could not be formed. The dual-scale structured alloy has excellent ductility–strength matching, and therefore a dynamic stable tribolayer can easily be maintained on the worn surface, leading to excellent wear performance.

1. Introduction

Grain refining is a method widely used to promote the wear properties of bearing alloys because the ploughing and micro-cutting in the wearing are reduced by the hardening introduced by grain refining [1,2]. In addition, the wear resistance of alloys is proportional to their hardness, and can be described by the empirical Archard equation [3] over a certain range. Thus, different methods, such as mechanical alloying (MA) [4], rapid solidification [5], physical vapor deposition [6] and severe plastic deformation [7], have been used to fabricate nanocrystalline (NC) bearing alloys.

However, many reports [8–13] have revealed that the wear properties of NC alloys are not improved, or even worsened, in comparison with their coarse grained (CG) equivalents, even though their hardness and strength have both increased significantly. This is mainly attributed to the mechanical properties of NC alloys and the complex wear mechanisms operating in these samples [14]. These can be summarized by the following three factors. (1) The ductility and fracture toughness is poor for NC alloys. As demonstrated experimentally, the low plasticity greatly facilitates the removal of material via brittle fracture during the sliding wear of a NC AISI52100 steel [8]. (2) Reduced strain hardening capability and the presence of non-equilibrium grain boundaries (GB) are expected to reduce the wear resistance of NC materials. A decrease in the wear resistance after processing by equal-channel angular (ECA) and accumulative roll-bonding (ARB) for Al-1050 [9], Al-6061 and Al sheet [10,11], respectively, were attributed to the lack of significant strain hardening and non-equilibrium high energy and unstable GB character in alloys subjected to such severe plastic deformation processing. (3) The oxide tribolayer plays an important role in determining the wear resistance of the alloys in both the NC and CG states. Purcek et al. [12,13] demonstrated that an exceptional increase in strength and hardness after ECA process did not lead to an improvement of the wear resistance. They found that an oxide layer induced by the tribo-chemical reaction played a dominant role and even counteracted the strengthening effect on the wear resistance.

Therefore, to enhance the friction and wear properties of NC alloys, efforts should be made to improve certain inferior mechanical properties, including limited plasticity and weak work-hardening tendencies, non-equilibrium high energy GB and worn surface structural evolution [14]. Actually, the poor ductility and fracture toughness are currently the main obstacles preventing the use of NC alloys in industrial applications, such as for structural

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0043-1648/$ - see front matter
http://dx.doi.org/10.1016/j.wear.2013.05.001

components [15]. Considerable effort has been made to toughen high strength NC alloys, and an important finding is that significant toughening is achieved by creating a bimodal microstructure consisting of CG in an NC matrix [16].

These achievements also provide hints for enhancing the wear properties of NC tribological components. Recently, we demonstrated that a dual-scale structure of the soft CG part in an ultrafine grain (UFG) matrix, which was created by mechanical alloying (MA), leads to a remarkable improvement of wear performance of Al–Sn alloys [17]. However, the contribution of the wear mechanism and worn surface structural evolution to the great improvement in wear properties of this unique structure remain unknown. Actually, the discrepant structure and unique characteristics of worn surface structure evolution are intimately connected to the excellent wear performance [18,19].

The objective of the current work is to determine the worn surface structural evolution associated with the wear properties of a dual-scale Al–12 wt%Sn alloy in comparison with the corresponding monolithic UFG and CG samples. These structural changes and the correlation between these changes and the CG +UFG dual-scale structure provide a novel approach to reveal the tribological behavior of bearing alloys.

2. Experimental

The starting Al and Sn powders, ~200 mesh in size and 99.5% purity, were supplied by Guanghua Chemical Factory Co., Guangzhou, China, and Sinopharm Chemical Reagent Co., Shanghai, China, respectively. The Al–12 wt%Sn CG powder was simply obtained by mixing the Al and Sn powders. NC powder of the same composition was prepared by MA as follows. The powders were sealed in a stainless steel vial together with hardened steel balls in a ratio of 1:15 and milled for 40 h at a rotation speed of 250 rpm. Prior to milling, 1 wt% absolute ethanol was added as a process control agent to inhibit the cold-welding process. The powder handling and milling process were performed under a pure argon atmosphere. Subsequently, the NC powders were mixed with the CG powders at weight ratios of 9:1, 7:3, 5:5, 3:7 and 1:9, denoted as CG-10, CG-30, CG-50, CG-70 and CG-90, respectively. The powder mixtures were consolidated using a uniaxial press at a stress of 660 MPa and sintered in a vacuum oven at 823 K for 1 h. The final bulk alloys consisted of UFG together with uniform discrete CG regions as given in Fig. 1(b) and (c). For comparison, Al–12 wt%Sn alloys made of either monolithic NC powders or monolithic CG powders were also prepared under the same conditions. These are denoted as CG-0 and CG-100, which are shown in Fig. 1(a) and (d), respectively. Most of the Sn dispersoids are homogeneously distributed in the UFG Al matrix, but there is still some netlike distribution of Sn inevitably due to divorced eutectic reaction between Al and Sn as shown in Fig. 1(a). The grain sizes of the Al and Sn in the UFG region range from about 200 to 500 nm and 100 to 200 nm, respectively. While the grain sizes of CG Al and Sn remain within the microscale size of about 2∼3 μm and 1∼3 μm, respectively. More detailed information has been published in previous work [17].

A Philips X'pert MPD X-ray diffractometer (XRD) using Cu-K (λ=0.1541 nm) radiation, a Nano 430 scanning electron microscopy (SEM) equipped with an energy dispersive spectroscopy facility (EDS) were used to characterize the microstructure of the alloys. Micro-hardness was measured by using a HVS-1000 digital hardness tester, supplied by Guangshi Instrument Co., Guangzhou, China, with a load of 4.9 N. Each hardness data point was an average value of not less than seven measurements. The average hardness of the CG-0 sample is about 78.8 HV and there is a decrease with reducing mass fraction of the CG region as shown in Fig. 2. A CSM NHTX nanoindenter was also used to measure the hardness of tribolayers with a Berkovich indenter and a maximum load of 10 mN maintained for 5 s.

Wear properties were measured with an M-2000 wear tester, supplied by Beilun Balancing Machinery Co., Xuanhua, China, in a block-on-ring mode under dry conditions at ambient temperature and pressure. The samples, with sizes of $10\times10\times3$ mm^3, were placed in contact with a steel ring with a rotation speed of 214 rpm under a defined load for 1 h. The ring is 47 mm in diameter and 10 mm in width and made of a hardened steel containing 1.5 wt%Cr and 1 wt%C, with hardness of 58∼60 HRC. The samples were polished and then cleaned in acetone in an

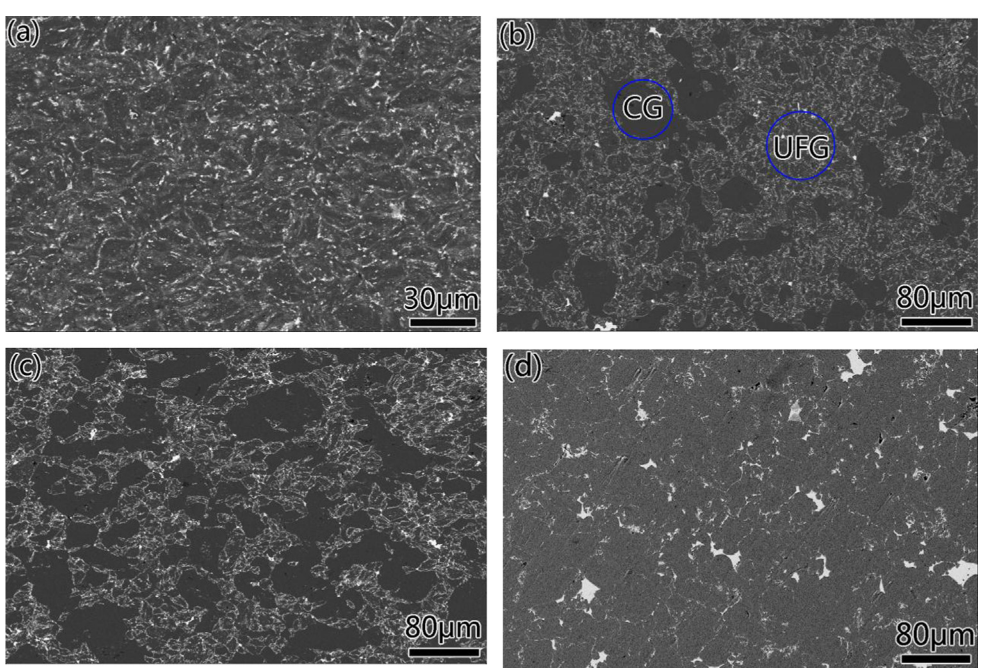

Fig. 1. SEM images of (a) monolithic UFG, (b) CG-30, (c) CG-50 and (d) monolithic CG alloys after sintering at 823 K for 1 h.