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Amorphous and nanostructured $Al_{85}Ni_5Y_6Co_2Fe_2$ powder prepared by nitrogen gas-atomization

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Abstract

$Al_{85}Ni_5Y_6Co_2Fe_2$ fully amorphous and partially amorphous powders were prepared by nitrogen gas-atomization method. The microstructure and thermal stability as a function of powder size were investigated. A majority of the powders are below 45 μm, and the content of amorphous phase is as high as 97.3% in the <25 μm size range. With the increase of particle size, the phase constituent alters from amorphous to amorphous' + fcc-Al, then fcc-Al + AlNiY + Al_3(Ni, Fe), and finally fcc-Al + Al_9Co_2 + Al_3Y + Al_3(Ni, Fe). All of the powders show three exothermic reactions in the DSC traces, and the <25 μm size range presents a wide supercooled liquid region of 22 K and a primary crystallization temperature of 580 K. The $Al_{85}Ni_5Y_6Co_2Fe_2$ alloy powders exhibit a high microhardness of near 348 Hv, suggesting its good potential for structural application.

Keywords: Aluminum; Amorphous structure; Gas-atomization; Microhardness

1. Introduction

Al-based amorphous alloys with aluminum content over 80 at.%, which exhibited good ductility in addition to tensile strengths exceeding 800 MPa, made its debut in 1988 in Masumoto group [1] and Poon group [2] independently. Since then, extensive researches have been performed on this kind of glass forming system. Recently, a number of Al-based alloys with a mixed structure consisting of α-Al nanoparticles homogeneously embedded in an amorphous matrix have been obtained, exhibiting tensile strengths of over 50% higher than the corresponding amorphous single phase without compromising ductility [3,4]. These data were obtained for melt-spun ribbons, whereas structural applications require the production of bulk components.

Bulk metallic glass (BMG) is regarded as a strong candidate for industrial application in the future, due to its unique and superior performance. Compared with Zr, Pd, etc. -based BMGs, the glass forming ability of Al-rich alloy system is very limited, and the largest dimension is as small as 900 μm [5–10] reported so far. To obtain its bulk amorphous sample, there are two possible processes to try, *i.e.* further optimizing the alloy composition to quench from the liquid state, or consolidation of amorphous powders. The latter is a feasible way for exploiting some marginal metallic glasses industrially. Some studies have been undertaken on the latter recently, and run into success to some extent [11–13].

Consolidation of Al-based amorphous powders is a thermomechanical process, and harmful precipitation of metastable phases or intermetallics may occur. Accordingly, to keep monolithic amorphous or primary nanoscale Al particles plus amorphous phase structure during consolidation is a key issue to prepare Al-based BMG in a powder metallurgy mode. Gas-atomization is an established technology for producing pre-alloyed powders for subsequent consolidation into engineering parts [14]. However, which alloy composition should be chosen for atomization to prepare the amorphous powders remains a challenge. It was reported that an alloy of $Al_{85}Ni_5Y_6Co_2Fe_2$ exhibits a pretty high primary crystallization temperature as well as a wide temperature interval between the primary Al nanocrystal precipitation and the subsequent intermetallics precipitation [15]. The motivation of this study is to characterize the glass forming ability of the gas-atomized $Al_{85}Ni_5Y_6Co_2Fe_2$ alloy. The microstructure, thermal stability,

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Table 1
Size distribution of the gas-atomized $Al_{85}Ni_5Y_6Co_2Fe_2$ powders

Particle size (μm)	Mass fraction (%)
<25	39.6
25–45	46.8
45–71	9.1
>71	4.5

and microhardness were also studied as a function of powder size.

2. Experimental details

A mixture of pure element Al (99.999 wt.%), Ni (99.9 wt.%), Y (99.9 wt.%), Co (99.5 wt.%) and Fe (99.9 wt.%) with the nominal chemical composition of $Al_{85}Ni_5Y_6Co_2Fe_2$ (at.%) was induction-melted three times under a high-purity argon atmosphere. To ensure that the impurity content is as low as possible, the melting chamber was evacuated twice to 10^{-3} Pa, and argon was backfilled in the interval. Atomization was carried out in a close-coupled nozzle atomizing system. The master alloy was induction-heated up to a superheat level of approximately 220 K in an alumina crucible under 10^{-2} Pa. Then the melt was delivered through a guide tube with the diameter of 2.5 mm, and atomized by a jet of nitrogen at 2.7 MPa. The molten metal was sprayed at a rate of about 1.2 kg/min. The $Al_{85}Ni_5Y_6Co_2Fe_2$ powders were allowed to cool down to ambient temperature in inert gas atmosphere of the atomizing chamber. Afterwards, they were collected and sieved in a closed system filled with inert gas.

The phase constituent of the $Al_{85}Ni_5Y_6Co_2Fe_2$ particles was investigated by X-ray diffraction (XRD) using a Rigaku D/max 2400 diffractometer with monochromated Cu Kα radiation ($\lambda = 0.1542$ nm). The thermal stability of amorphous sample against crystallization was examined with differential scanning calorimetry (DSC) in a Perkin-Elmer DSC 7 at a heating rate of 40 K/min under a dynamic argon atmosphere. The morphologies and microstructures of the as-atomized powders were characterized using scanning electron microscopy (SEM). To assess the mechanical properties of the powders and the capacity to be consolidated, Vickers microhardness was measured on the polished cross-sectional particles by applying loads of 50 g for 5 s.

3. Results and discussion

The size distribution of the $Al_{85}Ni_5Y_6Co_2Fe_2$ powder prepared by gas-atomization with N_2 is listed in Table 1. The dependence of cumulative mass distribution on the particle size is shown in Fig. 1. It is indicated that most of powders are below 45 μm in diameter (about 80%) and the finest particles (−25 μm) approximately account for 40% mass fraction.

SEM observations of the powders are given in Fig. 2. The amount of coarse particles is small in the general landscape (Fig. 2a), consistent with its mass fraction. Most of the powder particles are spherical in shape. Besides, there are some non-spherical particles which are formed by large particles being capped with small ones (Fig. 2b). The solidification rates are different for particles with different diameters. There exists a lower solidification rate for larger particles due to a slow heat transfer rate, whereas a higher solidification rate for smaller ones. During atomization, particles with various sizes would collide with each other in the gas turbulence. As a result of various solidification rates, particles of different size might take on different solidification state in a certain flying trajectory length. In other words, the larger droplets might be still in liquid or semi-solid state while the smaller ones have already solidified entirely in the jet gas turbulence. It can be reckoned that these particles in different states would adhere together readily on collision. Thus, it may occur that some particles are composed of larger ones with small caps in the cooled powders. This case is helpful to improve the powder packing efficiency. From Fig. 2c, it is also noted that the surface of the particles below 25 μm is pretty smooth, while a ragged surface in the larger particles, associated with the occurrence of a non-planar solidification front and the subsequent solidification shrinkage.

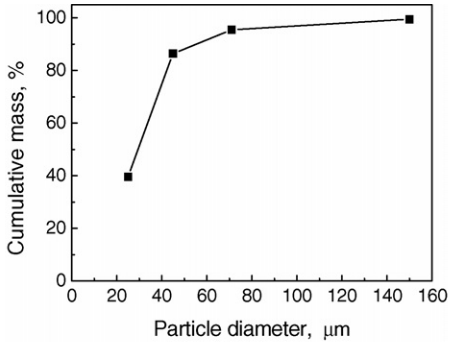

Fig. 1. Cumulative mass distribution dependence on the particle size.

XRD patterns of the as-atomized $Al_{85}Ni_5Y_6Co_2Fe_2$ powder as well as the as-cast ribbon are shown in Fig. 3. There exhibits a broad halo and some weak crystalline peaks in the particles of below 25 μm, indicating the coexistence of an amorphous phase and nano-Al phase. In comparison, the XRD pattern of the as-cast ribbon sample shows only a broad halo of fully amorphous phase. This reflects that the cooling rate of gas-atomization is apparently lower than that of melt-spinning. Further, for the 25–45 μm range particles, it shows a weak halo and a set of peaks of fcc-Al, and the peaks due to AlNiY and Al_3(Ni, Fe) are also detected. The powders in the 45–71 μm size range present a pattern similar to that of 25–45 μm, except for the narrowing of the halo and the strengthening of the crystalline phase peaks. The largest size range, 71–150 μm, is almost crystalline, with a very narrow halo, fcc-Al, Al_9Co_2, Al_3Y and Al_3(Ni, Fe) peaks on the background. It is noted that AlNiY exists in size range of both 25–45 and 45–71 μm, but does not show up in the 71–150 μm range. Therefore, it is concluded that the cooling rate of 71–150 μm range is the lowest, and AlNiY, as a metastable phase in the $Al_{85}Ni_5Y_6Co_2Fe_2$ alloy, appears only at a certain cooling rate.

From the diffraction patterns, one can also notice that the fcc-Al peaks shift to smaller angle region regarding to that of the equilibrium phase. This shift is due to the effect of rapid solidification on the supersaturated solid solution. The most remarkable change occurs in the finest powders owing to its largest cooling rate, whereas little change in the coarsest powders. The lattice parameter of the fcc-Al phase in these powders varies from 0.4081 nm of the below 25 μm range powders to 0.4061 nm of the 71–150 μm range powders. All these values are higher than that of the as-cast sample (0.4055 nm) [16], reflecting that the higher the solidification rate is, the larger the saturation degree of Al solid solution with solute atoms. The Goldschmidt atomic radii of aluminum, nickel, yttrium, cobalt and iron are 0.143, 0.125, 0.178, 0.125 and 0.124 nm, respectively [17]. The radius