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The wear resistance of a liquid quenched metallic glass

The mechanical properties of metallic glasses prepared by liquid quenching have been studied extensively in recent years. There have, however, been no reported investigations of the behaviour of these materials under service conditions similar to those that may arise in potential applications. In particular, since the surfaces of suitable metallic alloys can be "glazed" by laser-beam heating [1] it follows that the wear resistance of liquid-quenched metallic glasses should be examined. Furthermore, improvements to the wear resistance may possibly be realized following a controlled heat-treatment of the glass to give a partially crystalline structure. The treatment may involve adjusting the initial cooling rate and/or reheating the as-quenched glass.

This note presents some measurements of the relative wear resistance of amorphous and partially crystalline specimens of a liquid-quenched $Pd_{781}Cu_{55}Si_{164}$ alloy. The crystalline samples were obtained by reheating the glass at a constant rate to temperatures above T_g, the glass transition temperature. The Pd-based alloy and the heat-treatment regime were chosen because the corresponding partially crystalline structures have been examined previously using transmission electron microscopy [2].

Rods of the $Pd_{781}Cu_{55}Si_{164}$ alloy were prepared by sucking the molten alloy into quartz capillaries. The capillaries were then sealed at one end, placed for 3 min in a furnace held at 900 K and quenched into water. The wear rates, at 300 K, of sections of the rods were then measured using a pin-on-disc configuration with metallographic grade, silicon carbide abrasive papers. At the completion of each wear test, the remaining lengths of the amorphous rods were heated in a DTA unit at 20 K min^{-1} to various temperatures above T_g. Following quenching into water, the heat-treated sections were abraded using conditions identical to those employed for the original amorphous rods. Vickers microhardnesses were also determined using a 15 g indenting load.

Fig. 1 shows plots of the rate of specimen weight loss, $\dot{w}$, as a function of the sliding distance, s, for an amorphous $Pd_{781}Cu_{55}Si_{164}$ rod. The rod was not translated during the two separate tests which were conducted at sliding velocities, v, of 70 and 90 mm sec^{-1} using 320-mesh abrasive papers. The wear rates are seen to be non-linear and in the case of crystalline alloys, this behaviour has been attributed to degradation of the abrasive paper [3]. The figure also demonstrates that the initial wear rate, $\dot{w}_i$, increased as the sliding velocity decreased but the origin of this behaviour was not investigated. Systematic variations in $\dot{w}_i$ were also observed on changing the rod diameter, the normal load and the abrasive particle size. However, these effects are predicted by the first order theory [4] of abrasive wear and they are of little immediate interest in the study of the relative wear resistance of amorphous and partially crystalline alloys. Nevertheless, the systematic investigation of the dependence of $\dot{w}_i$ on the wear variables gave the appropriate conditions for making accurate measurements of the relative

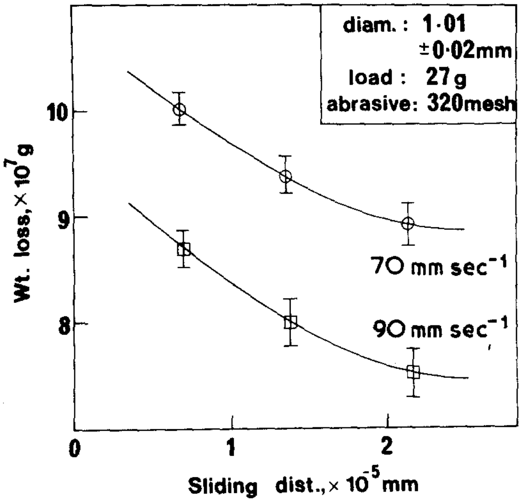

Figure 1 Weight loss curves for an amorphous $Pd_{781}Cu_{55}Si_{164}$ alloy showing non-linear wear behaviour at two different sliding velocities.

wear resistance. These conditions were found to be a normal load of ≃ 25 g for a 1 mm diameter rod abraded against 320-mesh abrasive paper

Since the errors associated with the experimentally determined initial wear rates were large, it was decided to compare the relative wear resistance of the partially crystalline specimen by measuring the ratio, $\bar{w}'/\bar{w}$, of the average wear rates for the partially crystalline and amorphous alloys (the prime denotes a heat-treated sample). Accordingly, each amorphous rod was abraded for a sliding distance of $1.8.10^5$ mm at 75 mm sec^{-1}. The rod was then moved to a fresh track and abraded for a further 1.8×10^5 mm at 85 mm sec^{-1}. The average wear rate of the amorphous rod, $\bar{w}$ was taken to be the total weight loss divided by the total sliding distance. The remaining section of the rod was then heat-treated and its average wear rate, $\bar{w}'$, was measured using the same procedure. Several wear tests were conducted using each of the amorphous and heat-treated rods in order to average out statistical variations.

Fig. 2 gives plots of the wear rate and Vickers hardness ratios, $\bar{w}'/\bar{w}$ and H'/H, for the $Pd_{781}Cu_{55}Si_{164}$ alloy as a function of T_{max}, the maximum temperature to which the amorphous alloy was heated. In the lower half of the figure is shown the DTA thermogram for the alloy; by comparing T_{max} with its position along the thermogram we can identify the types of partially crystalline structures that were present in the wear-test specimens.

Earlier research [2] has established that the first DTA exotherm, labelled MS-I in Fig. 2, involves the formation of numerous small and closely spaced crystals of an fcc phase. The subsequent MS-II exotherm on continuous heating, involves the nucleation and growth of two-phase "spherulites" or "cells" that were designated MS-II(2). The other MS-II crystallization product, namely the single-phase silicide grains designated MS-II(1), was only formed in appreciable amounts during the isothermal heat-treatment, at relatively high temperatures, of ≃ 10 μm thick filamentary-cast specimens [2]. Optical microscopy of the heat-treated rods confirmed this general picture in that equi-axed MS-II(2) spherulites (and no MS-II(1) grains) nucleated and grew during the second DTA exotherm. It was also established that the

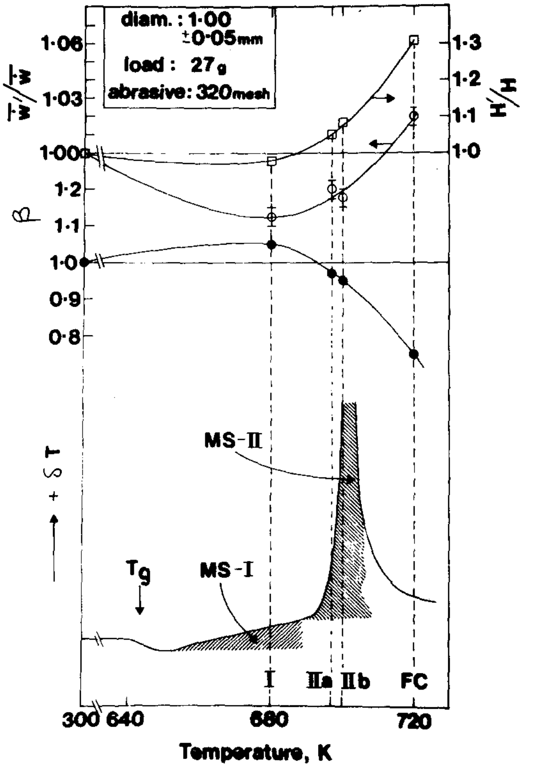

Figure 2 Upper curves show observed average wear rate ratio, hardness ratio and relative wear coefficient following heating to various T_{max} temperatures. Lower curve is the DTA thermogram obtained on heating the amorphous $Pd_{781}Cu_{55}Si_{164}$ alloy at 20 K min^{-1}.

four heat-treated alloys (designated I, IIa, IIb and FC) used in the wear tests, contained 0, 5 ± 1, 30 ± 5 and 100% by volume of MS-II(2).

Fig. 2 illustrates that both the wear rate and hardness decreased slightly during the MS-I crystallization stage and increased dramatically during the subsequent MS-II stage. These results are not predicted by the first order theory of abrasive wear [4] for which $\dot{w} \propto H^{-1}$, i.e. the wear rate increases as the hardness decreases. This relationship between $\dot{w}$ and H can be used to define a relative wear coefficient $\beta = \bar{w}H/\bar{w}'H'$, where the small density change observed on crystallization has been ignored. The experimentally determined coefficients are plotted in Fig. 2 in order to demonstrate that β increased modestly during the MS-I stage and then decreased rapidly during the MS-II stage.

The concepts used in modelling the wear behaviour of crystalline two-phase materials can be used to rationalize the observed dependence of the wear resistance on the microstructure of the heat-treated metallic glass. The rate of abrasive wear is thought to be determined not only by the resistance to indentation (as in the first-order theory) but also by the nucleation and propagation of the cracks that are required to remove material [5]. Accordingly, the observed decrease in $\bar{w}$ as H decreased during the MS-I stage, and the increase in $\bar{w}$ as H increased during the MS-II stage, imply that the wear resistance of the partially crystalline $Pd_{781}Cu_{55}Si_{164}$ alloy is largely controlled by crack nucleation and growth and not by the specimen hardness. A critical examination of this proposal would require some knowledge of the fracture toughnesses of the heat-treated glasses but unfortunately these measurements are not available. However, fracture stresses, σ_f, of partially crystalline metallic alloys have been measured [6–8]: in all cases it would appear that samples having a small volume fraction of the crystallization products display fracture stresses which are larger than those for the totally amorphous alloys. In the case of an isothermally heat-treated $Pd_{80}Si_{20}$ glass there was found [6] to be an increase in σ_f during the initial stages of the MS-I crystallization reaction followed by a gradual decrease at longer crystallization times. The onset of the MS-II reaction then gave an accelerated rate of decrease of σ_f. It is, therefore, not unreasonable to argue that MS-I microcrystals reduce the rate of crack propagation by providing obstacles to plastic flow [7]. It is this reduction which brings about the increase in the wear resistance. The ease of crack nucleation is presumably not affected by the presence of the MS-I (fcc) crystals because these crysals are "coherent" with the matrix (i.e. there is no free volume located at the crystal–amorphous phase interface [9]) and they are sufficiently small (<20 nm) that the energy condition for crack nucleation is not satisfied [10]. The rapid decrease in the wear resistance during the MS-II stage can be attributed to easy crack nucleation within the brittle silicide constituent of the MS-II(2) cells; and to the propagation of cracks between the cells and within the silicide constituent.

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