Fatigue Crack Initiation and Small-Crack Propagation in Zr-Based Bulk Metallic Glass

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1. Introduction

Because amorphous metals have been produced at a cooling rate higher than \(10^5 \text{K/s}\), very thin ribbons or wires can be obtained. Recently, several families of multicomponent metallic alloy, which shows excellent glass forming ability, have been developed. For the alloys, conventional casting, whose cooling rate is lower than \(10^5 \text{K/s}\), can be applied to produce amorphous alloys. Using the alloy, bulk structural components can be produced, where the alloy is called bulk metallic glass (BMG). The lack of any long-range order and the subsequent absence of microstructures have led to very high strength and very high corrosion resistance. BMG is in a supercooled liquid state at a temperature between the glass-transition temperature, \(T_g\), and the crystallization temperature, \(T_x\), and machine components can be made by precision casting of BMG because of its low shrinkage from the supercooled liquid state to the solid state. Owing to these excellent properties, BMG has received considerable attention as a structural material, especially for small or micromachine components. To ensure the integrity of structures made of BMG, the mechanisms and mechanics of fatigue fracture should be clarified.

The mechanical properties of BMG have been widely studied; however, the fatigue mechanisms and mechanics still remain unclear. In tension tests, BMG specimens break with very little bulk plastic deformation. The lack of plastic deformation may indicate that the fatigue limit of the alloy approaches the ultimate tensile strength. In fact, however, Gilbert et al. reported that the fatigue limit of a Zr-based BMG is as low as 6–8% of the ultimate tensile strength.1) Liaw et al. conducted fatigue tests on notched specimens and reported that the fatigue limit of a Zr-based BMG was much higher than the value reported by Gilbert et al.2–4) Yokoyama et al.5) and Fujita et al.6) reported the same trend. For steels, the ratio is 0.35–0.6, and it decreases with increasing ultimate tensile strength for high-strength steels. Since the ratio for the Zr-based BMG was near the lower limit of steels, they concluded that it is comparable to the ratio for high-strength steels. Although the ratio for steels was obtained from fatigue tests under fully reversed cyclic loading \((R = -1)\), the value for most Zr-based BMGs was obtained from fatigue tests at a stress ratio of 0.1. It is well known that the fatigue limit depends on the stress ratio, \(R\), or mean stress, \(\sigma_m\), and the fatigue limit for sharp-notched specimens is usually higher than that for smooth specimens when notch root stress is employed for the comparison. Thus, in the present study, fatigue tests were conducted under fully reversed cyclic loading using smooth specimens.

2. Material and Experimental Procedures

The material for the present study was bulk metallic glass (BMG), \(\text{Zr}_{55}\text{Cu}_{30}\text{Ni}_{15}\text{Al}_{10}\) (at%). The tensile strength of the BMG was 1560 MPa, Young’s modulus was 87 GPa, and the elongation at fracture was almost 0%. Most of the samples were initially produced in a plate form by high-pressure casting, and some samples were produced by squeezed casting. The plate dimensions were 20 mm wide, 50 mm long, and 2.0 mm thick, and the surface was polished by grinding. The fatigue specimens were wire-electrical-discharge-machined from the plates. The geometry and dimensions of the fatigue specimens are shown in Fig. 1, where the stress concentration factor of the shallow notches is 1.03.7)

The surface of the specimens was polished using emery paper and then electrochemically polished before conducting fatigue tests. A computer-controlled electrodynamic testing machine was employed for the fatigue tests. Fully reversed cyclic bending moment with a frequency of 40 Hz was applied to the specimen.

The surfaces of the fatigued specimens were observed by atomic force microscopy (AFM) to elucidate the fatigue crack initiation mechanism. To observe the fatigue process in

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Fig. 1 Shape and dimensions of specimen (in mm).
detail, replicas of the specimen surface were obtained at the predetermined number of fatigue cycles. The replica films were coated with Au before AFM observation. Although the height of the surface in the replica film was reversed with respect to that of the specimen surface, the height of the replica film in the AFM images was reversed by an image processing technique.

3. Experimental Results

3.1 Fatigue life and fatigue limit

The $S$-$N$ curve of the smooth specimens is shown in Fig. 2. The present material has a clear fatigue limit of 410 MPa, which is 26% of the ultimate tensile strength of the material, 1560 MPa. The number of cycles at the knee point in the $S$-$N$ curve is $6 \times 10^4$ cycles, which is much lower than the value of most crystalline metals.

To represent the effect of mean stress, $\sigma_m$ on the fatigue limit, $\sigma_a$, several relationships were proposed. Among them, the following two equations are well known for crystalline metals:

$$\sigma_a = \sigma_{00} [1 - (\sigma_m/\sigma_B)^2] \quad \text{(Gerber)} \quad (1)$$

$$\sigma_a = \sigma_{00} [1 - \sigma_m/\sigma_B] \quad \text{(Modified Goodman)} \quad (2)$$

where $\sigma_{00}$ is the fatigue limit for $\sigma_m = 0$, and $\sigma_B$ is the ultimate tensile strength. These relations are shown in Fig. 3, together with the experimental results obtained by Wang et al.\textsuperscript{2} and the present study. As in the case of crystalline metals, experimental results fall between modified Goodman and Gerber relationships.

3.2 Crack initiation

Wang et al. reported that fatigue cracks are initiated from casting defects, shear bands, or shear-off steps, and formation and propagation of shear bands are important for the initial stage of the fatigue damage process.\textsuperscript{3} To understand the fatigue damage in more detail, AFM was employed for the observation of the fatigue crack initiation process. Figure 4 shows AFM images before and after crack initiation, where the stress amplitude $\sigma_a$ was 510 MPa, and the fatigue life at this stress amplitude was $N_f = 7.6 \times 10^4$ cycles. In this case, the specimen surface was observed after every 500 cycles, and it was found that a crack was initiated at $4.2 \times 10^4$ cycles ($N_i/N_f = 0.55$). Up to $4.15 \times 10^4$ cycles, however, there was no significant change in the specimen surface at the crack initiation site.

Surface topologies around the crack were examined from these AFM images. The shapes of cross sections at the positions from A to E, which are designated in Fig. 5, are shown in Fig. 6. At positions A and B, the crack opens and a large step is formed across the crack, whereas only a step of the surface without any crack opening exists at positions C.
and D, and the surface is almost flat at position E. The step may have been produced by shear banding. These observations indicate that the crack was initiated at the bottom of the shear step just after its formation. After the initiation, the crack grew along the shear step, while the surface length of the step did not increase afterwards.

An AFM image of another crack initiation site at the same specimen is shown in Fig. 7. In this case also, a crack was initiated from a shear step, and it grew along the step. When it exceeded the shear step, the crack propagated without producing a step, and it grew in the Stage II manner, which is controlled by the number of cycles of normal stress. The maximum height of the step formed by shear banding in this case was about 200 nm, and shear steps with heights greater than 200 nm were also not observed for those formed near the final unstable fracture surface.

Since metallic glasses do not have slip systems, the plane of shear banding should be coincident with the maximum resolved shear stress plane that is given by the continuum mechanics, where the normal of the plane rotates $45^\circ$ from the loading axis under a uniaxial stress condition such as plane bending. In this case, the angle of the shear trace at the specimen surface relative to the loading axis should be between $45^\circ$ and $90^\circ$ depending on the shear direction, which can be estimated from the angle of the surface shear trace.8–10) For the shear trace angle of $90^\circ$ (Fig. 4), the angle between the shear banding plane and the specimen surface should be $45^\circ$, and it should be $90^\circ$ for the shear trace angle of $45^\circ$ (Fig. 7).

### 3.3 Fractography

The fracture surfaces of the specimen tested at a stress amplitude of 500 MPa are shown in Fig. 8, where (a)–(c) correspond, respectively, to the crack initiation, fatigue crack propagation, and final unstable fracture regions. At the crack
initiation site (a), the fracture surface is flat and no defects are observed. In the crack propagation area (b), striation-like morphologies can be observed, but they are most likely not striations because their spacing is much larger than the crack propagation rate. The final unstable fracture surface (c) shows a vein structure.\textsuperscript{11}

3.4 Small-crack propagation

Figure 9 shows the crack propagation rate as a function of stress intensity range, $\Delta K$, where the value of $K$ was calculated using equations derived by Raju and Newman.\textsuperscript{12} In the equations, the shape of the crack was assumed to be a semielliptical and the aspect ratio of the crack was considered to have the same value as that reported by Nakai and Tanaka for low-carbon steel.\textsuperscript{13} The crack propagation rates are plotted using open and solid marks for Stages I and II, respectively.

The growth rate of a long fatigue crack in the present material at $R = 0.1$ is also shown in the figure.\textsuperscript{14,15} It was found that the crack propagation rate, $da/dN$, in air under cyclic stress for this material was independent of loading frequency (cycle dependent) and stress ratio, $R$, and the threshold stress intensity range $\Delta K_{th}$ was $1.8 \text{ MPam}^{1/2}$. The crack propagation under cyclic stress in 3.5% NaCl solution, however, was time-dependent, and the crack propagation rate $da/dt$ was about $4.0 \times 10^{-6} \text{ m/s}$, independent of $\Delta K$,\textsuperscript{16}

In Stage I where the crack propagated along the shear step, the crack propagation rate increases with crack extension, while it has an almost constant value independent of $\Delta K$ in Stage II, where the decrease in crack growth rate at $\Delta K = 3.2 \text{ MPam}^{1/2}$ is attributed to coalescence with another crack. The transition of crack propagation from Stage I to Stage II took place when $\Delta K$ exceeded the threshold stress intensity range of long cracks.

The growth rate of small cracks is much higher than that of long cracks in air, which is indicated by a dash-dotted line in Fig. 9. In Stage II, the rate is almost equal to that in 3.5% NaCl solution, which is indicated by a dotted line. Electrochemical polishing may be responsible for this acceleration. For smooth specimens, electrochemical polishing was carried out to remove residual stress near the surface, which was possibly induced by machining, while only mechanical polishing was conducted for long crack propagation test specimens. It is known that hydrogen can easily enter Zr-based metallic glass during electrochemical polishing. Therefore, cracks may propagate with the assistance of hydrogen in the smooth specimen.

4. Conclusions

In the present study, fatigue tests on Zr-based bulk metallic glass (BMG), $Zr_{55}Cu_{30}Ni_{5}Al_{10}$, were conducted, and the following results were obtained.

1) In contrast to most brittle materials, Zr-based BMG showed fatigue behavior. The fatigue strength of the metallic glass was 410 MPa, which is 26% of the ultimate tensile strength of the material, 1560 MPa. The number of cycles at the knee point in the $S$-$N$ curve was $6 \times 10^6$ cycles, which was much lower than the value for crystalline metals.

2) From the surface observations of fatigued specimens, neither slip bands nor shear steps were observed at the crack initiation site before crack initiation. Cracks were initiated at the bottom of shear steps just after the formation of shear bands.

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