Ultra-High Cycle Fatigue Properties in $\text{Zr}_{55}\text{Al}_{10}\text{Ni}_5\text{Cu}_{30}$ Metallic Glassy Alloy Using Ultrasonic Fatigue Testing Machine[†]

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by

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In this work, the ultra-high cycle fatigue behavior of $\text{Zr}_{55}\text{Al}_{10}\text{Ni}_5\text{Cu}_{30}$ metallic glass was investigated by using an ultrasonic fatigue testing method. Test alloy rods with a diameter of 8mm were prepared by tilt-casting with a copper mold. Test specimens were machined to hourglass shape (the minimum diameter ; 2.0mm), and then manually polished with polishing paste. They were tested at a frequency of 20kHz under a stress ratio of -1 . The fatigue limit (σ_w) (half a total stress amplitude) and fatigue ratio (σ_w/σ_B) of the alloy were 893MPa and 0.53, respectively. Furthermore, no fish-eye like crack-growth morphology was seen on the fracture surfaces. The initiation region of fatigue cracking was always observed on the side surface of the specimens. A viscous-flow fracture region with large dimples between the fatigue crack-growth region and the final unstable fracture region was observed. This peculiar fracture morphology may be due to adiabatic heating of the metallic glass specimens during the fracturing, resulting in transition to supercooled liquid, under extremely high frequency of fatigue stress using ultrasonic loading.

Key words : *Amorphous, Giga cycle fatigue, Fractography, Ultrasonic fatigue, Metallic glass, Supercooled liquid*

1 Introduction

Metallic glass is in a sub-group of amorphous alloys. Amorphous alloys have random atomic arrangement without defects such as dislocations and grain boundaries, leading to a perfectly equiaxed homogeneous solid. It is well-known that amorphous alloys show significantly different properties than crystalline alloys. For example, amorphous alloys have no certain crystallographic slip systems, with the result that they are deformed by sliding along the maximum shear stress plane. Amorphous alloys do not show plastic deformation with dislocation movement, and so, they possess very high mechanical strength in general.¹⁾ Furthermore, amorphous alloys are known for their excellent corrosion resistance. This is because amorphous alloys do not have any grains or grain boundaries, leading to the formation of the homogeneous corrosion-resistant passive film on their surface.²⁾

Conventional amorphous alloys can be produced by freezing the molten liquid structure by rapid quenching of molten alloys at a cooling rate of around 10^6K/s . Thus, the shape of the products is strictly limited to thinner/smaller ones such as melt-spun ribbons, powders and wires.³⁾

From the end of the 1980s to the beginning of the 1990s, a series of compositions of amorphous alloys showing distinct glass transition and a wide supercooled liquid region was reported.⁴⁾ These significant amorphous alloys

were dubbed “metallic glass” to differentiate them from conventional amorphous alloys having poor glass transition. Molten alloys do not need to be cooled at an extremely high cooling rate of around 10^6K/s to obtain metallic glasses. Thus, it is possible to produce relatively larger samples of metallic glasses than in the cases of conventional amorphous alloys.

Moreover, hot plastic working of amorphous alloys is difficult due to their poor supercooled liquid region, leading to rapid and severe embrittlement at around the crystallization temperature, T_x . On the contrary, metallic glasses show a wide supercooled liquid temperature region just below the T_x . Thus, hot plastic working of metallic glasses is possible in a wide supercooled liquid region due to a significant decrease in viscosity, showing Newtonian flow, which can achieve a superplasticity of a few tens of thousands %.

As mentioned above, metallic glass can be recognized as a structural material because of its good hot-workability, high mechanical strength and excellent corrosion resistance. So, fatigue behavior of metallic glass, in particular, the S-N curves and the fatigue crack growth in the Pd-based and Zr-based metallic glasses, have been widely studied by many research groups.^{5)~10)} The characteristics of fatigue behavior of metallic glasses previously reported are summarized below.

1. In some cases, metallic glass shows extremely lower

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fatigue strength σ_w/σ_B (σ_w : fatigue limit, σ_B : tensile strength, if $R \geq 0$ then $2\sigma_w/\sigma_B$) than crystalline alloys although the tensile strength of metallic glass is very high. For example, it has been reported that the fatigue strengths of the Zr-based metallic glasses with a single glassy phase and with nano-crystalline particle dispersion were about 0.08 ($R = 0.1$)⁵ and 0.26 ($R = 0.1$)⁶, respectively, in the early works. Recently, it has been reported that the fatigue strengths of the Ti-based and Zr-based metallic glasses were about 0.8 ($R = 0.1$)^{7, 10} and 0.46 ($R = 0.1$)⁸, respectively.

2. Since metallic glass shows mechanical strength as high as high-strength steels, relaxation of stress concentration around the crack tip and the micro-defects is difficult. Micro-defects can thus lead to a severe decrease in fatigue strength and lifetime.⁶
3. Crack closure is difficult in metallic glasses, in particular, at small stress intensity factor range (ΔK) because of less plasticity-, oxide- and roughness-induced crack closures in glassy alloys.⁹
4. In general, glassy alloys possess very high mechanical strength. Thus, if they contain no micro-defects such as voids and precipitates, it is presumed that their fatigue strength σ_w can achieve tensile fracture strength, σ_B .¹⁰

Recently, an accelerated aging test method by ultrasonic loading has been established in order to investigate the fatigue behavior of materials in the giga-cycle range, much important knowledge having been accumulated and reported so far.^{11)~16)} For example, the specimens fractured under the range of the number of cycles to failure, N_f , of 10^6 – 10^7 cycles, show an ordinary fracture surface in which crack is initiated around the side surface of the specimens. On the contrary, it has also been reported that the S-N curve starts to decrease again under the range of the number of cycles to failure, N_f , of more than 10^7 cycles, showing a modal shift of the fracture mechanism of crack initiation from around the side surface to inertial micro-defects, when fatigue tests are continued over 10^8 – 10^9 cycles in the case of high-strength alloys.^{15), 16)}

In many cases, fatigue behavior of metallic glasses has been studied by using a conventional fatigue testing machine (e.g., servo-hydraulic type) at frequencies of 10–100 Hz up to about 10^7 cycles. To the authors' knowledge, there have been few reports on fatigue behavior of metallic glasses studied at extremely high frequencies over the range of very high cycles.

Therefore, in this work, fatigue behavior of the typical glassy alloy $Zr_{55}Al_{10}Ni_5Cu_{30}$ ¹⁷⁾ was studied by using an ultrasonic fatigue testing machine in order to obtain the S-N curve and to observe the fatigue fracture surfaces.

2 Experimental

Ingots of the $Zr_{55}Al_{10}Ni_5Cu_{30}$ quaternary alloy were prepared by arc-melting high-purity raw materials of Al (99.99% and 99.999%), Ni (99.999%), Cu (99.999%) and Zr (99.9%, surface cleaned by acid) in a dilute Ar atmosphere. The raw materials were purchased from High Purity Materials Kojundo Chemical Laboratory Co., Ltd. (Al, Ni and Cu) and Toyo Success Co., Ltd. (Zr). The rod-shaped glassy alloy samples were produced by tilt-casting in a dilute Ar atmosphere. The glassy rods were 8.5 mm in diameter and about 60 mm in length.

One-mm-thick disc-shaped pieces were cut from both ends of the rods and those discs were used for X-ray diffraction (hereafter denoted as XRD) analysis to identify a glassy single phase with Cu-K α radiation at 40 kV and 40 mA and at a scanning rate of 2 min/ $^\circ$. Moreover, thermal analysis was conducted for those pieces to investigate the thermal parameters such as the glass transition temperature, T_g , and the crystallization temperature, T_x .

Fatigue tests were conducted with an ultrasonic fatigue testing machine (Shimadzu, USF-2000) at a stress ratio ($R = \sigma_{min}/\sigma_{max}$) of -1 and a frequency of 20 kHz.

Fig. 1 shows the geometry of the hourglass-type specimen prepared in this study, the resonance frequency of which was determined to be 20 kHz by using the following equations (1)–(3).¹⁸⁾

$$L = \frac{c}{\omega} \cdot \tan^{-1} \left[\frac{c}{\omega} (\beta \cdot \coth \beta g - b \cdot \tanh bg) \right] \quad (1)$$

$$b = \frac{1}{g} \cdot \cosh^{-1} \left(\frac{R}{H} \right) \quad (2)$$

$$\beta = \sqrt{b^2 - \left(\frac{\omega}{c} \right)^2}, \quad (3)$$

where c and ω are longitudinal wave velocity ($= \sqrt{E/\rho}$, E : Young's modulus and ρ : density) and angular velocity ($= 2\pi f$, f : resonant frequency, 20 kHz), respectively. The specimen geometry was determined with the ordinary values of Young's modulus, $E = 90$ GPa and the density, $\rho = 6.83 \times 10^3$ kg·m $^{-3}$ of $Zr_{55}Al_{10}Ni_5Cu_{30}$ metallic glass. As a result of calculation, the following parameters were obtained: $H = 1$ mm, $R = 4$ mm, $g = 7$ mm, $L = 13.3$ mm. However, the practical resonance frequency of this specimen may have been different from 20 kHz because of the difference of Young's modulus and/or density of the specimens produced in practice. So, first, specimens

Fig. 1 Shape and dimensions of test piece.

with $L = 11.3, 12.3, 13.3$ and 14.3mm were prepared and then used to study the relationship between the experimentally obtained resonance frequency and the L value. After that, the L value was fixed at 12.0mm . The stress concentration factor of this specimen was 1.044 , so, it was not necessary to correct the practical stress with the stress concentration factor in this study. The stress was controlled by the software during the tests.

The stress amplitude, σ_a of the fatigue test was determined by the following equation¹⁸⁾:

$$\sigma_a = aE\beta \cdot \cos\left(\frac{\omega L}{c}\right) \cdot \cosh b\gamma \cdot \left(\frac{1}{\sinh \beta g}\right), \quad (4)$$

where a is displacement amplitude at the free end of the specimen. Prior to the fatigue tests, the value was measured by an eddy current sensor to calibrate the output level of the ultrasonic actuator controller.

The ultrasonic fatigue testing system has an air compressor to cool the ultrasonic actuator. A part of the cooling air was used to spray the middle of a specimen with a minimum diameter to decrease its temperature. Preliminarily, the temperature in the middle of the specimen subjected to cooling air was monitored and found to be around 310K . So, the ultrasonic fatigue tests were conducted continuously with cooling air and without intermittent loading/arrest intervals.

The specimens were machined with cooling water/oil so as not to increase the temperature. They were then polished manually with #2000 emery paper in the longitudinal direction and finished with polishing paste.

Furthermore, fracture surfaces of specimens were observed using a scanning electron microscope (SEM).

3 Results and Discussion

3.1 Specimen Preparation

Fig. 2 shows an example of a glassy rod, 8mm in diameter, produced by tilt-casting. The length of the rod is about 60mm . A disc of about 1mm in thickness was cut from both ends of the rod, polished with #2000 emery paper and then used for XRD observation. In the case that the XRD patterns obtained from both ends of the rod

Fig. 2 Examples of XRD patterns obtained from as-cast rod.

specimen showed a broad halo peak without any distinct peaks coming from crystalline phase, the rod was recognized as being a glassy one possessing only a single glassy phase. Amorphicity of all the specimens was confirmed before use for the fatigue tests.

The glass transition temperature, T_g , and the crystallization temperature, T_x , of the $Zr_{55}Al_{10}Ni_5Cu_{30}$ glassy alloy were 686K and 772K , respectively. The glass transition region, $\Delta T_x (= T_x - T_g)$, was 86K . These values corresponded fairly well with those reported in a previous paper,¹⁷⁾ indicating that this alloy is a typical glassy alloy with a wide glass transition region.

3.2 S-N Curve

Fig. 3 shows the S-N curve obtained in this study. The vertical axis indicates stress amplitude, σ_a , and the horizontal axis indicates cycles to failure, N_f . In this study, the test was terminated if the specimen did not show any fracture even after 10^9 – 10^{10} cycles. It has been reported that the tensile strength, σ_B , of the alloy is around 1600 – 1800MPa .¹⁹⁾ In this study, the σ_B was fixed to 1700MPa . The fatigue limit, σ_w , was determined based on a medium value between the destructive minimum stress and the non-destructive maximum stress after 10^9 cycles. The fatigue limit, σ_w , and fatigue ratio, σ_w/σ_B obtained in this study were 893MPa and 0.53 , respectively. A two-step decrease in fatigue strength in the giga-cycle region of the S-N curve,^{15), 16)} as reported previously for Fe-based crystalline alloys and steels, was not obtained in this study. Furthermore, initial crack started from the specimen side surface in all the specimens, with the result that the fish-eye like morphology,¹²⁾ reported in many previous papers, did not appear in this study, as mentioned in the next section. Therefore, it can be concluded that the S-N curve obtained in this study is an ordinary type S-N curve with one fatigue limit in the giga-cycle region. However, it should be noted that some specimens fractured after 10^7 cycles.

Wang et al. reported that the fatigue ratio of a similar Zr-based glassy alloy was 0.455 ($\sigma_a = 865\text{MPa}$) obtained

Fig. 3 S-N curve for the ultrasonic fatigue of $Zr_{55}Al_{10}Ni_5Cu_{30}$ metallic glass.

at a frequency of 10Hz.⁸⁾ Because this value corresponds fairly with the fatigue ratio obtained in this study, it can be concluded that the frequency dependence of fatigue properties and fracture mechanism is small in the Zr-based glassy alloys. Metallic glass can have extremely high fatigue strength because of its extraordinary high tensile/compressive strength. On the contrary, however, its fatigue strength may be significantly sensitive to inclusions because of its very high hardness.⁶⁾ In many previous reports, very low fatigue strength, σ_w , was observed due to the difficulty of sample preparation without microdefects such as inclusions, which can readily play a role of crack initiation sites. Fatigue strength higher than the values reported in many previous papers^{5), 6)} was achieved in this study since clean alloy ingots were successfully synthesized with fewer microdefects such as inclusions and precipitates by using 5N-class high-purity raw materials. However, metallic glass is a non-equilibrium material produced by a rapid cooling process. Thus, complete avoidance of probable precipitation of fine crystalline phases and microdefects is impossible. Therefore, a broad scatter of the data points of fatigue lifetime was observed in the S-N curve even at the same stress amplitude level. Some specimens in this study did not fracture even after 10^9 – 10^{10} cycles, probably because of fewer internal microdefects. Further study of ordinary fatigue behavior (e.g., at 10Hz) of the same $\text{Zr}_{55}\text{Al}_{10}\text{Ni}_5\text{Cu}_{30}$ glassy alloy and comparison of the result with that obtained in this study from the viewpoint of the microdefect creation probability, fatigue fracture mechanism and frequency effect of fatigue strength is necessary.

3.3 Investigation of Fracture Mechanism

Figs. 4–6 show fractographs obtained from three fractured specimens: Sample-1 ($\sigma_a/\sigma_B = 0.5$, $N_f = 2.12 \times 10^8$ cycles), Sample-2 ($\sigma_a/\sigma_B = 0.6$, $N_f = 2.00 \times 10^8$ cycles) and Sample-3 ($\sigma_a/\sigma_B = 0.7$, $N_f = 2.49 \times 10^4$ cycles). They are represented in Fig. 3 as solid circles (●). An arrow in each figure indicates crack growth and propagation direction.

Fig. 4 shows the fractograph of Sample-1. As seen in the figure, the crack was initiated at the side surface of

Fig. 4 Overall fatigue fractograph of Sample-1.

Fig. 5 Fractographs of Sample-2. Overall fractograph (a), microscopic images of crack-growth region with striation-like patterns (b) and unstable fracture region with vein patterns (c).

Fig. 6 Overall fractograph of Sample-3.

the specimen (indicated in the open circle in Fig. 4), accompanied by beach marks, and the final fracture region showed vein-like patterns. It has a typical fracture morphology.

Fig. 5 (a) shows the fractograph of Sample-2. Figs. 5 (b) and (c) are magnified images of the fracture surface. In Fig. 5 (a), the crack initiation can be seen at the side surface of the specimen in an open circle, this fatigue crack proceeding from right to left. Typical striation-like patterns are visible in Region A in Fig. 5 (b) and typical vein patterns characteristic of the tensile fracture surface of metallic glass can be seen in Region B in Fig. 5 (c). The appearance of vein patterns indicates that the temperature of shear bands significantly increased during sliding, leading to a supercooled liquid state around the fracture surface. Therefore, Region B is an unstable fracture region with a fast sliding fracture. Furthermore, a characteristic fracture region with large dimples, which may be formed in a supercooled liquid state (hereafter denoted as viscous flow fracture region), was observed between Region A and Region B in Fig. 5 (a).

Fig. 6 shows the fractograph of Sample-3. The crack initiation site is in the open circle and the fatigue crack grew from left to right in the figure. The viscous flow fracture region with large dimples, seen in the center of the fractograph in Sample-3, is wider than that of Sample-2 in Fig. 5 (a).

As mentioned above, it was found that the crack was initiated at the side surface of the specimens in all cases and that inertial crack initiation and fish-eye type fracture morphology¹²⁾ were not observed in this work.

As for the fractographs observed in this work, a crack growth region (Region A) and an unstable (fast) fracture region (Region B) show ordinary morphological features usually obtained in fatigue study of metallic glass. However, to the authors' knowledge, there have been few reports on fatigue behavior of metallic glass studied by using an ultrasonic testing machine, in which the appearance of a viscous flow fracture region with large dimples is mentioned. The appearance of this viscous flow fracture region is considered by the authors to be a characteristic feature of ultrasonic fatigue fracture of metallic glass. Thus, the formation mechanism of such a characteristic fracture morphology will be discussed below.

Fig. 7 shows schematic illustrations indicating the fracture process of Samples-2 and 3. First, fatigue crack was initiated at the side surface of the specimen and grew with striation-like patterns, as shown in Fig. 7 (a). When the fatigue crack grew further under tensile stress conditions, the stress intensity factor K increased, leading to an increase in the fatigue crack growth rate. At that time, the temperature around the front of the crack tip increased with repetitive plastic deformation. As can be seen, the yield stress decreased with temperature, resulting in larger repeating plastic deformation at the crack tip. This led to a rapid temperature increase more and more at the crack tip.²⁰⁾ In general, metallic glass possesses lower thermal conductivity than crystalline alloys. Thus, if it is deformed at a high strain rate and its temperature is raised in an adiabatic process, it is presumed that rapid heat evolution of local temperature increase is difficult. As a result, the temperature around the crack tip increased up to the glass transition temperature, T_g , and the viscosity around the crack tip significantly decreased, leading to the capability of a large deformation. At that time, cavities were created, then they became larger and were combined with each other at around the crack tip front as shown in Fig. 7 (b) and the crack rapidly proceeded as can be seen in Fig. 7 (c). However, since the test frequency was extremely high, the load started to decrease during the crack propagation. After a half period ($25\mu\text{s}$), load direction changed

Fig. 7 Schematic illustration of the peculiar fracture mechanism of the metallic glass under the ultrasonic fatigue condition. The crack-growth region with striations (a), significant temperature increase and formation of cavities in front of the crack tip region (b), significant fast crack propagation through coalescence of cavities (c), crack closure under compressive stress (d) and formation of extremely large dimples under tensile stress at the supercooled liquid region (e).

from tension to compression, and then crack closure and subsequent reshaping occurred at the crack tip. After that, it is presumed that the crack propagation stopped as shown in Fig. 7 (d). The front border of the viscous flow fracture region with large dimples/cavities is straight, as can be seen in Fig. 5 (a) and Fig. 6. This observation may be a proof of our supposition. The load direction then changed again from compression to tension after a further $25\mu\text{s}$. The crack slightly propagated under $K_{\text{max}} > K_{\text{fc}}$ (fatigue fracture toughness), and the fracture mode moved to the unstable fast fracture as can be seen in the conventional fatigue tests shown in Fig. 7 (e). Thus, elastic strain energy under tension was suddenly evolved at once, the temperature around the crack tip further increased again,²¹⁾ and then the specimen was broken with vein-like patterns on the fracture surface, as shown in Region B in Fig. 5 (c) and Fig. 6. Meanwhile, a characteristic morphology with large dimples was formed in the region where the crack closed in a supercooled liquid state. This is because cavities were created, grown and combined with each other under three-dimensional stress conditions and were subsequently torn off in the final fracture process. This characteristic morphology will be further described in Figs. 8 and 9. The glass transition temperature, T_g , and the crystallization temperature, T_x , of the $\text{Zr}_{55}\text{Al}_{10}\text{Ni}_{15}\text{Cu}_{30}$ glassy alloy are $T_g = 686\text{K}$, and $T_x = 772\text{K}$, respectively.

So, it is presumed that the surface temperature of the viscous flow fracture region increased up to around 686K in the fracture process. Furthermore, the band width of the viscous flow fracture region observed in the middle of the fracture surface of Sample-3 in Fig. 6 is larger than that of Sample-2 in Fig. 5. This is related to the area in which temperature increases up to around T_g around the crack tip under large stress and large evolution of elastic strain energy stored around the crack tip. Liu et al. reported that a significant temperature rise occurred on the fracture surface, as shown by conventional fatigue test results at 10Hz, and that a droplet-like morphology was observed on the final fast fracture surface, which seemed to be formed in a supercooled liquid state due to the temperature rise on the fracture surface.²¹⁾ However, to the authors' knowledge, there has been no report of distinct large dimples in the viscous flow fracture region on fracture surfaces similar to those shown in Figs. 5 and 6 in this study. The present authors think that this fracture surface with distinct large dimples (about 100 μ m) is characteristic of the fracture surface of metallic glass obtained by an extremely high loading frequency under ultrasonic fatigue tests.

Fig. 8 Fracture surface with extremely large dimples of Sample-2. (a) and (b) show fracture surface morphology in the viscous-flow fracture region. The arrow in (a) indicates the crack-growth direction.

Fig. 9 Fracture surface with extremely large dimples of Sample-3. (a) and (b) show fracture surface morphology in the viscous-flow fracture region. (c) is a side-view of the fracture surface indicating extremely large dimple height.

Fig. 8 shows magnified fractographs of the viscous flow fracture region in Sample-2. Significantly large dimples were observed in the region. An arrow in Fig. 8 (a) indicates the direction of the fatigue crack propagation. No inclusions were observed in the bottom area of the dimples. Thus, it can be concluded that the formation of these large dimples is due to the formation of large bubbles originating from the creation, growth and combination of small cavities in a supercooled liquid under tensile condition.

Figs. 9 (a) and (b) are magnified fractographs of the viscous flow fracture region with large dimples in Sample-3. Fig. 9 (c) is a side view of the fracture surface. The height of the dimples is about 100 μ m, having a very high aspect ratio. This observation clearly shows that those dimples were formed by viscous flow in a supercooled liquid state at a temperature higher than T_g .

Fig. 10 shows fractographs near the fatigue crack initiation site in Samples 1-3. As seen in Fig. 10 (a), a precipitate about 8 μ m in diameter was observed in the fatigue crack initiation region in Sample-1. On the contrary, precipitates were not clearly seen in the fatigue crack initiation region in Samples-2 and 3. The lifetime of Sample-1 in Fig. 10 (a) was $N_f = 2.12 \times 10^8$ cycles ($\sigma_a/\sigma_B = 0.5$) and that of Sample-2 in Fig. 10 (b) was $N_f = 2.00 \times 10^8$ cycles ($\sigma_a/\sigma_B = 0.6$). Sample-2 has no distinct defects such as a precipitate phase around the fatigue crack initiation site, resulting in a long lifetime of more than 108 cycles, even at the fatigue ratio σ_a/σ_B of 0.6. Although Sample-3 in Fig. 10 (c) also has no distinct defects around the crack initiation site, the stress ratio was so high ($\sigma_a/\sigma_B = 0.7$) that it fractured at $N_f = 2.49 \times 10^4$ cycles.

4 Conclusions

In this work, fatigue behavior of the $Zr_{55}Al_{10}Ni_5Cu_{30}$ metallic glass was investigated by using an ultrasonic fatigue testing machine at 20kHz at a stress ratio of $R = -1$. The results obtained are summarized below.

Fig. 10 Specimen surface morphology near the crack initiation region in (a) Sample-1, (b) Sample-2 and (c) Sample-3.

(1) The S-N curve obtained in this work did not decrease again after 10^9 cycles, unlike previous reports for steels. The fatigue limit and the fatigue rate were $\sigma_w = 893\text{MPa}$ and $\sigma_w/\sigma_B = 0.53$, respectively.

(2) Crack was initiated at the side surface in all the specimens. Fish-eye type inertial crack initiation, reported in many previous papers on the giga cycle fatigue test, was not obtained in this work. Moreover, when stress higher than $\sigma_a/\sigma_B = 0.6$ (1020MPa) was loaded, a viscous flow fracture region with large dimples appeared between the crack growth region with striation-like patterns and the final unstable fracture region. Those large dimples were formed under repetitive tensile/compression stress conditions in a supercooled liquid state at very high frequency. The appearance of a viscous flow fracture region with large dimples in the fracture surface seems to be a characteristic feature of metallic glass under fatigue testing at very high frequency.

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