Elastic and Anelastic Properties of Amorphous Thin Films

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Abstract. The interesting non-linear elastic and anelastic properties recently found in the metal-metal and the metal-metalloid amorphous alloys will be reported. The Young's modulus E_d measured by means of the vibrating reed method with the strain amplitude $\epsilon_t \approx 10^{-6}$ is found to be lower than E_s in the frequency f range below ($10^3 \sim 10^4$) Hz, showing a minimum at around 10^2 Hz, and to increase beyond E_s in the f range above ($10^3 \sim 10^4$) Hz, where E_s denotes the Young's modulus in linear elasticity observed for the static tensile or bending tests. These results suggest that a certain resonant anelastic-process is excited under alternating strain in the f range below ($10^3 \sim 10^4$) Hz. The ϵ_t dependence of E_d measured at $f \approx 10^2$ Hz shows that E_d increases towards E_s with increasing ϵ_t . Further, for the measurements at $f \approx 10^2$ Hz with $\epsilon_t \approx 10^{-6}$, an increase in E_d is also found but under passing an electric direct current (PEC) with 10^7 A/m², suggesting that internal stress is induced under PEC. These results suggest the view that a certain cluster of many atoms undergoes a resonant motion under alternating strain, and also yields internal stress due to the concentration of the electromigration force through a collective motion under PEC. The effective charge number Z^* which measures internal stress induced under PEC is found to be the order of 10^5 for all the amorphous alloys. The number of atoms involved in the above cluster is estimated to lie between 10^2 and 10^5 .

1. INTRODUCTION

The mechanical properties of amorphous alloys have been investigated mostly on thin-tape or thin-film specimens prepared by melt-quenching. For plastic deformation of amorphous alloys (see [1] and references therein), homogeneous deformation is observed above the transition temperature T_c of 0.7 to $0.8T_g$ (T_g , the glass transition temperature), where deformation is viscous. Below T_c , plastic deformation takes place through localized slips, where flow stress is not sensitive to temperature and the strain rate. In the microstrain range below yielding, a localized shear deformation (abbreviated as LSD) plays an important role. The present paper deals with the recent topics on the elastic and anelastic responses listed in Table 1.

It is widely observed that in the tensile tests, amorphous alloys show linear elasticity in the low strain (ϵ_t) range below (0.4~1) % and a non-linear increase in anelastic strain in the high strain range but below yielding, (0.4~1) % $< \epsilon_t <$ about 2 %, ('tensile test' in Table I). For linear elasticity in the low strain range, it is known that both the Young's modulus and the shear modulus of amorphous alloys are lower by about 20 to 40 % than those in the crystalline state but the bulk modulus is similar between the amorphous and crystalline states [2-13]. The pioneer computer simulation study [6] shows that elastic deformation in the amorphous structure is not uniform in an atomic scale, where the LSD in a small region containing several atoms (which will be referred to as the deformable unit, below) is suggested to be responsible for the low Young's modulus and the low shear modulus in the amorphous state. For the non-linear increase in anelastic strain found in the high strain range (e.g., [14]), the computer simulation [1, 15, 16] shows that in the high strain range, the LSD expands in a roughly spherical region of several atomic distances. Meanwhile, in the X-ray study, the bond-orientational anisotropy is found in the amorphous alloys

annealed under stretching [17-19]. Such an anisotropy is expected when the LSD attained during annealing under stretching partly remains after unloading following the annealing. So far the outline of the elastic and anelastic responses of amorphous alloys observed in the tensile tests may be explained by the LSDs and their evolution: The LSD in every deformable unit increases linearly with stress in the low strain range below (0.4~1) %. In the high strain range above (0.4~1) %, the LSD in every deformable unit goes up to a certain saturation value and then, a size of every deformable unit starts to expand, giving rise to the non-linear increase in anelastic strain. Experimentally, however, the knowledge on the LSD in the deformable unit is very limited.

Table 1 The elastic and anelastic responses of amorphous alloys in the pre-yield microstrain range below T_c . For a tensile test, ϵ_t is strain, E_s the static Young's modulus found in linear elasticity for $\epsilon_t < (0.4 \sim 1) \%$, and $(E_d)_{\text{cry}}$ E_s in the crystalline state. For a vibrating reed measurements, ϵ_t is the strain amplitude, f the vibrational frequency, E_d the dynamic Young's modulus deduced from f , and PEC an abbreviation of 'passing electric current'. See Fig.3 or Fig.4 for ϵ_A and ϵ_Y .

type of strain	measurement condition		reference
	elastic and anelastic responses		
static tensile strain (tensile test)	strain range		[1-16]
	$\epsilon_t < (0.4 \sim 1) \%$	$(0.4 \sim 1) \% < \epsilon_t < \text{about } 2 \%$	
	linear elasticity ($E_s < (E_s)_{\text{cry}}$)	non-linear increase in anelastic strain	
alternating strain (vibrating reed)	f -dependence of E_d measured with $\epsilon_t \approx 10^{-6}$		present [20]
	$f < (10^3 \sim 10^4) \text{ Hz}$	$(10^3 \sim 10^4) \text{ Hz} \leq f$	
	$E_d < E_s$	$E_d \geq E_s$	
	ϵ_t -dependence of E_d measured at $f \approx 200 \text{ Hz}$		[20-23]
$\epsilon_t < \epsilon_A (= \text{about } 4 \times 10^{-3})$	$\epsilon_A \leq \epsilon_t < \epsilon_Y (= \text{about } 1.05 \epsilon_A)$		
	E_d increases showing saturation	further steep increase in E_d	
	effect of PEC on E_d measured at $f \approx 200 \text{ Hz}$ with $\epsilon_t \approx 10^{-6}$		[28-30]
	E_d increases with increasing PEC		

Since anelastic strain can vary depending on the relaxation time of the anelastic process and the measurement frequency, a vibrating reed measurement is expected to give an insight into the LSD in the deformable units. The preliminary vibrating reed study for amorphous $\text{Pd}_{80}\text{Si}_{20}$ [20] shows that E_d measured with $\epsilon_t \approx 10^{-6}$ is lower than E_s in the frequency range below $(10^3 \sim 10^4) \text{ Hz}$ and increases beyond E_s for the higher frequencies (' f -dependence of E_d ' for 'vibrating reed' in Table I), where ϵ_t is the strain amplitude, E_d the dynamic Young's modulus determined from the resonant frequency f of a reed, and E_s the static Young's modulus in linear elasticity observed in the low strain range for the tensile tests. This result is surprising, because for anelastic materials as well as amorphous alloys, E_d should be higher than E_s when a resonant anelastic-process is not involved. The vibrating reed study at $f \approx 10^2 \text{ Hz}$ on the ϵ_t dependence of the relative change in E_d for various amorphous alloys [20-23] shows that E_d increases showing saturation in the strain range below about 4×10^{-3} (' ϵ_t -dependence of E_d ' for 'vibrating reed' in Table I). It is noted that in the above strain range, however, linear elasticity is observed in the tensile tests.

These results suggest that a certain resonant anelastic-process takes place under alternating strain in the f range below ($10^3 \sim 10^4$) Hz, and may be responsible for the non-linear anelastic response observed at $f \approx 10^2$ Hz for $\epsilon_t < 4 \times 10^{-3}$. The fact that the above frequency range is so much lower than the Debye frequency presumably suggests that the resonant anelastic-process is associated with a large number of atoms.

Recently, in the electrical resistivity measurements on the structural relaxation and the crystallization processes in amorphous alloys, it is found that these processes are enhanced under passing an electric direct current (abbreviated as PEC) with the current density i_d of 10^7 A/m^2 , suggesting that the concentration of the electromigration force takes place through a probable collective motion of many atoms under PEC ([24-27] and see the section 3). The combination of the above results found under PEC and ' ϵ_t -dependence of E_d ' for 'vibrating reed' in Table I predicts that E_d measured at $f \approx 10^2$ Hz with $\epsilon_t \approx 10^{-6}$ can be modified by PEC. The preliminary study of the effect of PEC on E_d shows that the prediction is the case [28-30] ('effect of PEC on E_d ' for 'vibrating reed' in Table I). The fact that the current density here is so low as 10^7 A/m^2 suggests that a number of atoms are associated with the probable collective motion. That is, both the characteristic anelastic behaviors found under alternating strain and those under PEC suggest that a certain collective motion of many atoms can be excited in amorphous alloys. The following sections are devoted to these topics.

2. EFFECTS OF ALTERNATING STRAIN

Prior to the vibrating reed measurements, we carried out the tensile tests and the three-point bending tests in the strain range near or below 10^{-3} , because such a static test in the low strain range has been very limited before. Figure 1(a) shows an example of the tensile tests for the melt-spun amorphous $\text{Pd}_{80}\text{Si}_{20}$ (a- $\text{Pd}_{80}\text{Si}_{20}$) specimens, where linear elasticity is observed in the present strain ϵ_t range below about 35×10^{-4} .

Figure 1. Examples of (a) the tensile and (b) the three-points bending measurements of a- $\text{Pd}_{80}\text{Si}_{20}$ which were carried out at room temperature (RT). In (b), F , w , t and L denote the load applied to and the width, thickness and gage length of a specimen, respectively (typically, $w=1 \text{ mm}$, $t=20 \mu\text{m}$ and $L=20 \text{ mm}$)

Figure 2. Examples for the f dependence of E_d observed for three a- $\text{Pd}_{80}\text{Si}_{20}$ specimens, where the dynamic Young's modulus E_d is determined from the resonant frequency f of a clamped-free specimen measured with the strain amplitude ϵ_t of 10^{-6} at RT (see [20] for details). The thin arrows denote the static Young's modulus E_s found in Fig.1(b). The bold arrow 1 denotes E_s found in Fig.1(a).

For the five specimens measured, the values of the static Young's modulus E_s lay between 8.9 and 10.3×10^{10} Pa, where the mean value is $9.6 \pm 0.7 \times 10^{10}$ Pa. The present value of E_s is slightly higher than the values of E_s reported for a-Pd₈₀Si₂₀; 6.7 [5], 8.0 [12] and 8.8 [13] $\times 10^{10}$ Pa. Figure 1(b) shows an example of the three-points bending tests of a-Pd₈₀Si₂₀ in the strain range below about 4×10^{-4} . In Fig. 1(b), although the strain range is limited, linear elasticity is again observed for the bending tests, where the values of E_s observed for the three specimens lay between 7.6 and 8.8×10^{10} Pa (see the thin arrows in Fig. 2). The values of E_s found in the bending tests are slightly lower than the values of E_s found in the tensile tests, presumably reflecting that a bending test is more sensitive to the surface roughness of a specimen than a tensile test. After taking into account the difference in the test conditions, one can say that the elastic and anelastic responses under static bending strain are the same to those under static tensile strain.

Figure 2 shows the preliminary results for the measurement frequency f dependence of the dynamic Young's modulus E_d observed for the three a-Pd₈₀Si₂₀ specimens, which were measured by adjusting the gage length of a vibrating reed, where the strain amplitude ϵ_t during the measurements was 10^{-6} . In Fig. 2, E_d reported for a-Pd₈₂Si₁₈ [9] is also plotted, which lies nearby the present data. The f vs. E_d data are slightly different among the three specimens, presumably reflecting that the surface roughness is not equal among the three specimens. Apart from the minor difference, in general, E_d is lower than E_s in the frequency range below ($10^3 \sim 10^4$) Hz, showing a minimum at around 10^2 Hz, and then, increases beyond E_s with increasing f . Since the elastic and anelastic responses under static bending strain are the same to those under static tensile strain as mentioned for Fig. 1, the characteristic f -dependence of E_d found in Fig. 2 is not due to bending strain in the vibrating reed measurements but due to alternating strain. The fact that E_d is lower than E_s for f below ($10^3 \sim 10^4$) Hz suggests that a certain resonant anelastic process is excited under alternating strain in the f range.

Meanwhile, the ϵ_t dependence of the relative changes in E_d at around $f \approx 10^2$ Hz has been compiled for various amorphous alloys including a-Pd₈₀Si₂₀ [20-23], where for a clamped-free composite-reed of a

Figure 3. (a) an example of the ϵ_t dependence of f observed at RT for a-Pd₈₀Si₂₀, where the gage length L of a clamped-free composite-reed is kept at a given value, about 6 mm. (b) a redrawing of (a) but here the normalized stress $(f/f_e)^2 \epsilon_t$ is plotted. See text for details.

Figure 4. Similar to Fig. 3(a) and 3(b) but here the results observed for a-Cu₅₀Zr₅₀ are shown.

given gage length, the ϵ_t dependence of f is measured in the strain range below about 40×10^{-4} . Figure 3(a) shows an example of the ϵ_t dependence of f observed for a-Pd₈₀Si₂₀, where an increase and saturation of f for ϵ_t below ϵ_A ($\epsilon_A \approx 39 \times 10^{-4}$, here) and an abrupt increase in f for ϵ_t between ϵ_A and ϵ_Y ($\epsilon_Y \approx 1.05\epsilon_A$) can be seen. For ϵ_t beyond ϵ_Y , f changes over to decrease and the prolonged vibrations causes the fatigue failure of specimens, suggesting that micro-yielding starts (not shown here). On the other hand, when ϵ_t stays below ϵ_Y , the ϵ_t dependence of f is reproducible. The internal friction remains almost constant for ϵ_t below ϵ_A , and shows a steep increase for ϵ_t above ϵ_A (see Fig. 1(b) in [23] but for a-Cu₅₀Zr₅₀). These anelastic responses observed for ϵ_t below ϵ_Y remain almost unchanged for the wide temperature range between 80 and 400 K (not shown here).

From the saturation behavior of f found in the strain range below ϵ_A , we estimated the ultimate value f_e and then, deduced the normalized stress $(f/f_e)^2 \epsilon_t$ vs. ϵ_t relation shown in Figure 3(b) from the data shown in Fig. 3(a). In Fig. 3(b), the observed non-linear elasticity is assumed to consist of the constituent linear elasticity with $f=f_e$ giving ϵ_e in ϵ_t and the constituent non-linear anelasticity associated with ϵ_i in ϵ_t , where $\epsilon_{i,t}$ is the saturation value of ϵ_i found at $\epsilon_t=\epsilon_A$, here $\epsilon_{i,t} \approx 0.11\epsilon_A$. The abrupt increase in f for ϵ_t above ϵ_A seen in Fig. 3(a) is explained as the abrupt decrease in ϵ_i in Fig. 3(b). f_e is slightly higher than f calculated for a vibrating reed with E_s (see [20] for details). Figures 4(a) and 4(b) are similar to Fig. 3(a) and 3(b), but here the results observed for a-Cu₅₀Zr₅₀ [23] are shown. The characteristic ϵ_t dependence of f and the steep increase in the internal friction for ϵ_t above ϵ_A are commonly observed for a-Cu₅₀Zr₅₀ [23], a-Cu₅₀Ti₅₀ [21,22] and a-Pd₈₀Si₂₀ [20].

In the tensile tests in the pre-yield microstrain range [14], the appearance of hysteretic strain is detected in the higher strain range in which the non-linear increase in anelastic strain is observed. On the other hand, in the vibrating reed measurements for the ϵ_t dependence of f , the steep increase in the internal friction is observed for ϵ_t above ϵ_A , suggesting that expansion in the size of the deformable units reported in the computer simulation study [1,15,16] starts for ϵ_t beyond ϵ_A , in other words, the size of the deformable units remains unchanged for ϵ_t below ϵ_A . Then, the characteristic anelastic responses found under alternating strain can be explained as follows: For ϵ_t below ϵ_A , the LSD in every deformable unit is enhanced due to a resonant motion of a certain cluster of many atoms as a whole under alternating strain, and shows saturation with increasing stress. For ϵ_t beyond ϵ_A , expansion in the size of the deformable units starts, and gives rise to the complementary decrease in the size of the above clusters, resulting in a decrease in the resonant anelastic-process. This model will further be discussed in the section 4.

3. EFFECTS OF PASSING ELECTRIC CURRENT

As mentioned in the section 1, both the structural relaxation and the crystallization processes of various amorphous alloys are found to be accelerated under PEC with the current density i_d of 10^7 A/m² [24-27,31,32]. For the crystallization process, the homogeneous nucleation process is accelerated under PEC [27]. Meanwhile, it has been reported that the homogeneous nucleation process is accelerated under uniaxial stress [33,34]. The combination of these results suggests that PEC induces internal stress in amorphous alloys, the order of 0.1 GPa under PEC with i_d of 5×10^7 A/m², and predicts that E_d can be modified under PEC when the vibrating reed measurements are carried out at $f \approx 10^2$ Hz with $\epsilon_t \approx 10^{-6}$.

Figure 5 shows an example of the effect of PEC on the resonant frequency f which was measured at $f \approx 200$ Hz with $\epsilon_t \approx 10^{-6}$ for an a-Cu₅₀Zr₅₀ reed specimen (see Fig. 1 in [28] for the measurement setup). f under PEC was measured in a He gas atmosphere of 10^2 Pa to minimize the effect of joule heating (the open symbols in Fig. 5). An increase in the specimen temperature under PEC was measured separately and then, the change in f due to the increase in the specimen temperature under PEC (the filled symbols in Fig. 5) was estimated from the f vs. temperature data which were also separately measured without PEC. The net increase in f , Δf , under PEC was estimated after the subtraction of the change in f due to joule heating as shown in Fig. 5.

Figure 6 shows examples of $\Delta f/f_0$ observed at $f \approx 10^2$ Hz with $\epsilon_t \approx 10^{-6}$ under PEC for the various a-alloys, where f_0 is f observed without PEC. In the same figure, the i_d vs. f data observed for the crystalline Ni-Cr alloy is also shown, where no net-change in f was observed. In contrast, the considerable net-

increase in f , $(\Delta f/f_0)_{\text{PEC}}$, is commonly found for all the amorphous alloys but the magnitude of $(\Delta f/f_0)_{\text{PEC}}$ depends on the chemical composition of the a-alloys. $(\Delta f/f_0)_{\text{PEC}}$ diminishes after crystallization (not shown here)

Further, for every a-alloy shown in Fig.6, the i_d dependence of $(\Delta f/f_0)_{\text{PEC}}$ observed under PEC is found to show good agreement with the ϵ_t dependence of $(\Delta f/f_0)_\epsilon$ after scaling the ϵ_t axis with a proportional constant α as $\epsilon_t = \alpha i_d$, where $(\Delta f/f_0)_\epsilon$ denotes an increase in f with increasing ϵ_t observed at $f \approx 10^2$ Hz without PEC (see Fig.3(a) and Fig.4(a)). Figure 7 shows an example of the scaling made between the i_d dependence of $(\Delta f/f_0)_{\text{PEC}}$ and the ϵ_t dependence of $(\Delta f/f_0)_\epsilon$ observed for a-Cu₅₀Zr₅₀.

After the LSD-model mentioned in the section 2, the LSD in the deformable units is enhanced under alternating strain and shows saturation with increasing applied stress. The i_d dependence of $(\Delta f/f_0)_{\text{PEC}}$ observed under PEC also suggests that the LSD attained under alternating stress of $\epsilon_t \approx 10^{-6}$ decreases under internal stress induced by PEC. Then, as the first step for the evaluation of the effect of PEC, we assume that the force F_{PEC} acting on an atom in the deformable unit under PEC may be calibrated by the corresponding force F_ϵ under increased ϵ_t .

When an dimensional cross-section of an atom is denoted by S , F_ϵ can be given by,

$$F_\epsilon = SE\epsilon_t, \quad (1)$$

where $E\epsilon_t$ measures tensile stress in a specimen and the elasticity theory is assumed to be applicable to account for F_ϵ . We assume that F_{PEC} reflects the concentration of the electromigration force [35,36] through a collective motion of many atoms and can be accounted for using the effective charge number Z^* as

Figure 5. An example of the effect of passing electric current (PEC) on f which was measured for an a-Cu₅₀Zr₅₀ reed specimen at $\epsilon_t \approx 10^{-6}$: ○: the observed changes in f under PEC, ●: the changes in f due to joule heating. Δf denotes the net change in f under PEC.

Figure 6. Examples for the net changes in f under PEC, $(\Delta f/f_0)_{\text{PEC}}$, observed at $\epsilon_t \approx 10^{-6}$ for the various a-alloys, where f_0 is f found without PEC (~ 200 Hz here): 1 at room temperature (RT) and 2 at 80 K for a-Cu₅₀Zr₅₀, 3 at RT and 4 at 80 K for a-Cu₅₀Ti₅₀, 5 at RT and 6 at 80 K for a-(Cu₃₀Zr₇₀)_{92.5}Al_{7.5}, 7 at RT for a-Pd₈₀Si₂₀. The data observed for the crystalline Ni-Cr alloy, 8 at RT, is also shown (see text).

$$F_{\text{PEC}} = Z^* e \Phi, \quad (2)$$

where e denotes the elementary charge, and Φ the strength of the electric field in a specimen under PEC. The combination of equations (1) and (2) gives the relationship,

$$Z^* = \alpha SE / e \rho_R, \quad (3)$$

where ρ_R denotes the resistivity of the a-alloys. The application of equation (3) to the results shown in Fig. 7 or those observed for the various a-alloys gives Z^* listed in Table 2, which is the order of 10^5 .

The nearly free electron model predicts that the effective charge number of an atom in crystalline metals is the factor of ten [37]. The effective charge number which is accounted for from the enhanced dislocation motion under PEC reported in crystalline metals [38] is an order of magnitude larger than the charge number per atom reported in the crystalline metals [35-37], presumably reflecting that a motion of dislocation corresponds to a correlated motion of many atoms. We assume that Z^* measures the concentration of the electromigration force onto the front atom(s) in the probable collected motion of ξ atoms with the charge number z^* per atom as

$$Z^* = \xi z^*. \quad (4)$$

ξ is expected to vary after annealing of the a-alloys for the structural relaxation and the crystallization at elevated temperatures because of a change in the amorphous structure. Examples for the annealing experiments are shown in Figures 8(a) and 8(b), where the change in the length of the specimens $\Delta L/L$ and that in Z^* due to annealing observed for a-Cu₅₀Zr₅₀ are plotted, respectively. $\Delta L/L$ shows a small decrease by about 0.08% at around 550 K reflecting a decrease in the free volumes due to the structural relaxation and a steep decrease by about 0.7 % (see Fig. 7 in [23]) due to the crystallization. Z^* shows an increase by about 30 % after annealing at 500 K and then, a monotonous decrease to zero at higher

Figure 7. Similar to Fig. 6 but for a-Cu₅₀Zr₅₀, the $(\Delta f/f_0)_{\text{PEC}}$ vs. i_d data (the symbols, ● at 290 K and ○ at 80 K) are shown. In the same figure, the $(\Delta f/f_0)_e$ vs. ϵ_t data observed without PEC (the dashed lines) are shown after scaling the ϵ_t axis as $\epsilon_t = \alpha i_d$ with a proportional constant α . See text for details.

Figure 8. Effects of annealing on (a) the length of specimens L and (b) the effective charge number Z^* observed in a-Cu₅₀Zr₅₀. In (a) $\Delta L/L$ was measured at RT. In (b), the symbols ● and ○ were measured at 290 and 80 K, respectively.

temperatures. $\Delta L/L$ observed in the early stage of the structural relaxation is surmised to reflect annealing out of the vacancy-like free-volumes, here $(\Delta L/L)_{500K} = 0.015\%$ after annealing at 500 K. When the specific volume of the free-volumes annealed out at 500 K is equal to that of an atomic vacancy, the concentration ΔC_V of the free-volumes annealed out at 500 K can be accounted for as $\Delta C_V = 3(\Delta L/L)_{500K}$. We suppose that the increase in ξ after annealing at 500 K is due to annealing out of the vacancy-like free volumes which have interrupted the probable collected motion. Then, one may have the following relationship between ξ and ΔC_V as

$$\begin{aligned} Z^*_{500K}/Z^*_q &= \xi_{500K}/\xi_q \\ &\sim 1 + \xi_{500K}/(1/\Delta C_V) \\ &\sim 1 + \xi_{500K} \cdot 3(\Delta L/L)_{500K}, \end{aligned} \quad (5)$$

where ξ_q and ξ_{500K} denote ξ 's found in the as quenched state and after annealing at 500 K, respectively, and z^* in equation (4) is assumed to remain unchanged after annealing at 500 K. The application of equation (5) to the observed results of $Z^*_{500K}/Z^*_q = 0.3$ and $(\Delta L/L)_{500K} = 0.015\%$ gives $\xi_q \sim 510$ and $\xi_{500K} \sim 670$ for a-Cu₅₀Zr₅₀. ξ_q and ξ_{500K} found here and those for a-Cu₅₀Ti₅₀ [29] are listed in Table 2, where both ξ_q and ξ_{500K} are several hundreds. These results will further be discussed in the section 4.

Table 2. The effective charge number Z^* which measures the effect of PEC, and the number of atoms ξ associated with the probable collective motion which were accounted for from the annealing experiments, where ξ_q and ξ_{500K} denote ξ 's found at the as quenched state and after annealing at 500 K, respectively.

amorphous alloy	Z^*	ξ_q	ξ_{500K}	reference
Cu ₅₀ Ti ₅₀	2.6×10^5	630	830	[29]
Cu ₅₀ Zr ₅₀	3.1×10^5	510	670	[29]
(Cu ₃₀ Zr ₇₀) _{92.5} Al _{7.5}	(10^5)	(not measured)		[28]
Pd ₈₀ Si ₂₀	2.0×10^5	(not measured)		[30]

4. DISCUSSION

The results mentioned in the section 2 suggest that for ϵ_t below ϵ_A , the LSD in every deformable unit is enhanced by a resonant motion of a certain cluster of many atoms as a whole under alternating strain. The results mentioned in the section 3 also suggest that a certain cluster of many atoms undergoes a collective motion as a whole under PEC. The similarity found between the i_d dependence of $(\Delta f/f_0)_{PEC}$ under PEC and the ϵ_t dependence of $(\Delta f/f_0)_e$ without PEC (see Fig. 7) presumably suggests that the former and the latter probable clusters of many atoms are birds of a feather. Further, a resonant motion of the probable clusters as a whole can be expected when the probable clusters are embedded in the network of the matrix which contains the deformable units. In the above model, we implicitly assume that the LSD in the deformable units takes place at very low stress. Meanwhile, the study on the irradiation induced amorphization [39] reports that homogeneous amorphization is triggered by a shear instability, where the onset of the amorphization is expected to occur when the shear modulus of the irradiated metallic compounds vanishes. Thus, the above implicit assumption may not be unreasonable.

The value of Z^* estimated here (see Fig. 7 and Table 2) depends on the assumptions mentioned for equations (1) to (3), however, the present value may be not so far from the proper value. For example, from the combination of the acceleration of the crystallization process under uniaxial stress [33,34] and

that under PEC, Z^* is accounted for as 10^5 , i.e., the order of Z^* here is the same to that of Z^* accounted for using equations (1) to (3).

A simple account of ξ from equation (4) yields $\xi=(10^4\sim 10^5)$ for the combination of $Z^*\approx 10^5$ found here and $z^*=(1\sim 10)$ reported in crystalline metals, i.e., the calculated value of ξ here is surmised to give the upper bound of ξ . On the other hand, in equation (5), we assume that every free-volume which is annealed out below 500 K serves as an obstacle against the probable collected motion, where the lower bound of ξ is found. Therefore, we tentatively surmise that ξ lies between 10^2 and 10^5 . However, the underlying mechanism for the non-linear anelastic responses under alternating strain and that for internal stress induced under PEC are not known at the present. To clarify these issues, the further study is needed.

5. CONCLUSION

The elastic and anelastic responses measured for various amorphous alloys suggest the followings: A certain cluster of many atoms undergoes a resonant motion as a whole under alternating strain in the frequency range below $(10^3\sim 10^4)$ Hz, giving the model that the above clusters of many atoms are embedded in the network of the matrix which contains the deformable units. Passing an electric direct current (PEC) with 10^7 A/m² induces considerable internal stress through a certain collective motion of many atoms. The cluster of many atoms responsible for the resonant anelastic-process under alternating strain and that responsible for internal stress induced under PEC are surmised to be birds of a feather. The apparent charge number Z^* which measures internal stress induced or the concentration of the electromigration force under PEC is found to be the order of 10^5 . The number of atoms involved in the above cluster is estimated to lie between 10^2 and 10^5 . The underlying mechanism for the resonant anelastic-process under alternating strain and that for internal stress induced under PEC are, however, not known at the present.

Acknowledgements

The authors are grateful to Dr. H. Tanimoto (Univ. of Tsukuba) for his invaluable helps on the course of experiments. This work is partly supported by a Grant in Aid for Scientific Research from the Ministry of Education, Science and Culture of Japan.

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