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STUDY OF PLASTIC DEFORMATION IN BINARY ALUMINUM ALLOYS
BY INTERNAL-FRICTION METHODS

By E. C. Olson, R. E. Maringer, L. L. Marsh,
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SUMMARY

The damping capacity of several aluminum-copper alloys has been investigated during tensile elongation. This damping is shown to depend on strain rate, strain, temperature, alloy content, and heat treatment. A tentative hypothesis, based on the acceleration of solute atom diffusion by deformation-produced vacancies, is proposed to account for the observed behavior. Internal-friction maxima are observed in deformed aluminum and aluminum-copper alloys at -70° and -50° C. The peaks appear to be relatively insensitive to frequency and alloy content, but they disappear after annealing at temperatures nearing the recrystallization temperature.

INTRODUCTION

Internal friction, or damping capacity, has seen increasing application in the field of physical metallurgy because of its great sensitivity to relaxation processes occurring in metals. Application of this method to the study of diffusion, precipitation, recovery, and other such phenomena has proven highly successful and has led to a deeper understanding of such phenomena on an atomic scale.

One of the basic problems in the physics of metals concerns the nature of the interaction between foreign atoms and crystalline defects, such as vacancies or dislocations. Strain aging, the discontinuous yield point, the serrated stress-strain curve, and a number of other mechanical properties are believed to result from such interactions. This investigation was initiated on the premise that interactions which affect the gross mechanical properties should affect, to a greater degree, the more sensitive internal friction of a material. Previous work on aluminum-copper and aluminum-magnesium alloys (ref. 1) demonstrated that a number of phenomena were observable. The present work was therefore restricted primarily to the aluminum-copper system so that some of these phenomena might be studied more closely.

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SYMBOLS

C	concentration of vacancies
D	diffusion coefficient
H_f	heat of formation of vacancies
H_j	activation energy for motion of vacancies
Q^{-1}	damping or internal friction
R	universal gas constant
T	temperature, °K or °C
ϵ	plastic strain
$\dot{\epsilon}$	strain rate

EQUIPMENT

Internal-friction measurements were made using two torsional pendulums. These were patterned after that of K&E (ref. 2) and consisted of a rigid upper grip, a wire specimen, and a free-swinging lower grip supporting an inertia bar. A concave galvanometer mirror on the lower grip reflected the image of a hairline to a ground-glass scale several meters away. Internal-friction measurements were taken by observing the time required for the amplitude of vibration to decay a given amount.

The bottom-grip assembly and the inertia bar exerted a tensile force between 120 and 290 psi on the wire specimen, depending upon the inertia bar used. The maximum shear strain to which the specimens were subjected was about 4×10^{-5} , while the average shear strain during most measurements was held to 2×10^{-5} or less.

The apparatus in which damping was measured during straining, described in detail in reference 1, was altered by the substitution of a V-shaped wedge to modulate the light intensity in phase with the specimen oscillation. In essence, this machine consisted of a torsional pendulum mounted in a tensile machine. A constant strain rate was applied to the pendulum, and thus to the specimen, by the tensile machine. Load, the oscillation of the pendulum, strain, and time were recorded on a Sanborn twin-pen recorder. The recorded data were then analyzed to obtain internal-friction values. In later stages of the work the pendulum was inverted to protect the pickup system from heat.

MATERIALS

High-purity aluminum (see table 1) and copper were used in the preparation of these alloys. The melts were made by vacuum induction at a pressure of 1 micron of mercury and cast into graphite molds. The initial melts were alloyed at 1,100° C, cooled to 700° C, and cast. The ingots were then remelted at 700° C and recast. (See table 2 for alloy analysis.)

The resulting ingots were forged to 1 inch square at 315° C, grit-blasted, and cropped to remove shrinkage and other defects. The ingots were then rolled to a $\frac{5}{8}$ -inch diameter at 315° C and again grit-blasted. These rods were then rolled to a $\frac{1}{4}$ -inch diameter at 315° C and cold-swaged to a 0.187-inch diameter. After a 1-hour anneal at 380° to 480° C, the rods were cold-drawn to a 0.080-inch diameter. The resulting cold reduction in area was about 81 percent. These wires were heat-treated as shown in table 2 to obtain a common grain diameter of about 0.4 millimeter. In some of the tests commercial 2S aluminum was used.

EXPERIMENTAL RESULTS

Simultaneous Internal-Friction and Tensile Testing

Specimens of pure aluminum and aluminum-copper alloys were straightened, cleaned, and mounted in the internal-friction measuring apparatus described earlier. The specimens were heat-treated in place to avoid handling strains. Unless otherwise specified, the heat treatment was a recrystallization for 1/2 hour at 475° to 500° C, depending on the alloy. This produced a mean grain diameter of about 0.4 millimeter. The specimen was furnace-cooled from the recrystallization temperature to 300° C, held for 1/2 hour, and then furnace-cooled to the testing temperature.

Load measurements were recorded electronically from the distortion of a cylindrical load cell. Load readings could be estimated to within about ± 0.1 pound or about ± 20 psi. For the aluminum alloy containing 0.2 percent copper, over the entire temperature range investigated (-100° to 100° C), the stress at 1-percent plastic deformation was $2,450 \pm 300$ psi, with no increasing or decreasing trend observable. The flow stress at 1-percent deformation for pure aluminum and the alloys containing less than 0.2 percent copper appeared to decrease with increasing temperature, as would be expected (ref. 3). Because of the scatter and the auxiliary nature of the stress-strain curves, no analysis of them has been attempted.

All internal-friction measurements were made at frequencies of 1.4 to 1.5 cycles per second unless otherwise specified. The maximum strain amplitude on the surface of the specimens was 4×10^{-5} . Unless otherwise specified, all tests were run at a constant platen speed. This resulted in an approximately constant strain rate of 0.8×10^{-5} per second. Strain measurements were made from platen displacement corrected for the elastic response of the machine.

At the initiation of plastic deformation, the damping always rose very rapidly to a comparatively high value. The manner in which the internal friction varied with increasing strain depended upon the alloy and the temperature. In some cases, while the specimens were actively deforming, the scatter in the data points was rather severe. Two examples of this scatter are shown in figure 1. This scatter appears to be due at least in part to a real scatter in the damping of the specimen and not to experimental error. Because of the number of points involved, the points are not shown in the remainder of the graphs of experimental data.

It was shown in reference 1 that damping during tensile straining was dependent upon alloy content and test temperature. To determine this dependence more accurately tests were run on the series of aluminum-copper alloys previously described. The aluminum alloy containing 0.2 percent copper was examined in more detail than were the other alloys. The results of the examination are discussed at length later in the paper.

Figure 2 summarizes the temperature-dependence data. The apparently important features are as follows:

- (1) The damping rises to a high value at the initiation of plastic deformation.
- (2) The magnitude of the damping decreases with increasing copper content.
- (3) The decrease in damping after the initial rise is temperature dependent.

(4) At higher temperatures and in the alloys with higher copper contents, the damping first goes through a minimum and then begins to increase with increasing strain.

The initial rise in damping is undoubtedly associated with the onset of plastic deformation. It has been shown (ref. 4) that, during elastic deformation under conditions similar to those involved here, the increase in damping with strain is quite small for metals with no secondary relaxation mechanism. The previously observed (ref. 1) direct dependence of the internal friction on the strain rate and on the period of oscillation also suggests that the internal friction is sensitive to the amount of plastic deformation taking place per cycle.

The effect of impurity content, in this case copper, is quite appreciable. This is made more apparent in figure 3(a) where the damping at 1 percent strain is plotted against the copper content at several different temperatures. The ability of a given amount of copper to reduce the internal friction increases as the temperature increases. At 40° C there is, within experimental error, a linear relationship between the amount of solute and the logarithm of the internal friction. It is clear that the damping of the alloys does not decrease due to a decrease in the amount of deformation taking place, since the strain rate is constant. The effect seems to be much more intimately connected with the nature of the deformation than with the amount of deformation.

Above approximately 0° C the effect of impurity content is more pronounced. The abruptness of this effect is shown in figure 3(b) where the damping at 1 percent strain is shown as a function of temperature. With particular reference to the curve for the aluminum alloy containing 0.2 percent copper alloy, the sharp break is shifted either to higher or lower temperatures as the strain at which the internal-friction values are chosen is decreased or increased. There is, apparently, a strain-temperature-impurity-content relation which affects the nature of the deformation and its associated damping which occurs at a constant strain rate.

At temperatures above 15° C the internal friction of the aluminum alloys containing 0.1 and 0.2 percent copper begins to increase at higher strains (beyond the initial decrease in damping). Presumably, this might also occur at some of the lower temperatures if these tests were carried out to greater strains. The mechanism by which the damping is decreased seems, therefore, to build up to maximum effectiveness, and then as strain increases, to have a reduced effectiveness.

Since the damping during tensile elongation is directly proportional to the period of the oscillation and to the strain rate (see ref. 1), it is apparent that the amount of deformation being done during the period of measurement is an important parameter. However, if this were the only

important parameter, the curves of Q^{-1} versus $\dot{\epsilon}$ would be the same for all the alloys since a constant strain rate was used. This is clearly not true. A second important parameter may be the way in which the deformation is occurring. The data can then be interpreted as showing the influence of temperature, strain, and impurity content on the nature of the deformation. There is a hypothesis which might account for the experimental observations. Cottrell has proposed in reference 5 that with plastic deformation vacancies will be created and will accelerate the diffusion rate of solute atoms. When this diffusion rate is too slow or too fast, the effect of the interaction of solute atoms and moving dislocations is small. There should, however, be an optimum value of the diffusion rate at which the interaction is at a maximum. At this point, one should observe discontinuities in the stress-strain curve, the so-called Portevin-Le Chatelier effect. On the basis of some rather rough assumptions, Cottrell has suggested that the order of magnitude of the optimum value of the diffusion coefficient D is about $10^{-7} \dot{\epsilon}$ square centimeter per second, where $\dot{\epsilon}$ is the strain rate.

To apply the Cottrell mechanism to these data, it seems logical to assume that the lowest value of the internal friction corresponds to the optimum value of D . Why this should be so is not clear. It may be that the measuring stress can be relaxed more easily by a free dislocation than by one hindered by interaction with moving atoms. Thus, the damping would be lowest at the condition of maximum interaction.

Assuming this to be the case, it is of interest to determine the temperature and strain dependence of the minimum in the damping and, if possible, to extrapolate such a curve to zero strain where the diffusion rate can be determined from known data. The dependence between temperature and the strain at the damping minimum is shown as the line to the left in figure 4(a) for aluminum-0.2 percent copper alloy. Plotting reciprocal temperature versus the square root of strain produces, within experimental error, straight lines. The significance of the square-root relationship is not known, but it is interesting to note that, when the individual curves for damping during strain are plotted versus the square root of strain, the initial drop and the later increase in damping become symmetrical within experimental error.

It is also to be noted that an internal-friction peak is normally symmetrical when plotted versus reciprocal temperature. That is, the peak is symmetrical with respect to the logarithm of the diffusion rate. Empirically, this suggests, from figure 4(a), that the logarithm of the diffusion coefficient is proportional to $\sqrt{\dot{\epsilon}}$.

The extrapolated value of the reciprocal temperature is 0.0017, or 315° C. Using this value of the temperature and substituting into the equation for the diffusion of copper in aluminum (ref. 1)

$$D = 10^{-2} \exp \frac{-28,000}{R(588)}$$

$$= 4 \times 10^{-13} \text{ cm}^2/\text{sec}$$

On the basis of Cottrell's assumption, with $\dot{\epsilon}$ being approximately 0.8×10^{-5} , the optimum D is

$$D = 10^{-7} \times 0.8 \times 10^{-5}$$

$$= 8 \times 10^{-13} \text{ cm}^2/\text{sec}$$

Considering the assumptions made both in Cottrell's analysis and in the extrapolation procedure, this agreement is uncomfortably close. However, it does stand in support of the hypothesis that the internal-friction behavior can be interpreted in terms of a vacancy-accelerated diffusion and interaction of solute atoms with moving dislocations.

Cottrell went further in his analysis and, by using Mott's assumption (ref. 6) that the concentration of vacancies C is given by

$$C = 10^{-4} \epsilon$$

where ϵ is the plastic strain, arrived at the following expression for D :

$$D = 10^{-6} \epsilon \exp\left(-\frac{H_j}{RT}\right)$$

where H_j is the activation energy for the motion of vacancies. Actually, this is the increment in D with the onset of strain since the diffusion rate cannot be zero when the strain is zero.

Before applying this equation to the experimental data, some additional refinements to the original curves can be made. If one assumes, for the moment, the validity of the vacancy-accelerated strain-aging hypothesis, then a change of damping from a maximum to a minimum ΔQ^{-1} during straining is the result of the acceleration of the diffusion rate from an ineffective, slow rate to a higher rate of maximum effect. The damping at some constant fraction of ΔQ^{-1} should then represent a constant diffusion rate somewhat lower than that observed at higher fractions of ΔQ^{-1} . Various fractions of ΔQ^{-1} have also been plotted as $\sqrt{\epsilon}$ versus reciprocal temperature (fig. 4(a)) and are also seen to be linear within experimental error.

Holding to the assumption that the straight lines in figure 4(a) represent lines of constant diffusion rate, it is possible to calculate the diffusion rates by extrapolating the lines to zero strain and substituting these values of the reciprocal temperature into the solid-state diffusion equation. This leads to the conclusion that interaction of copper atoms with moving dislocations is already observable when the diffusion rate is of the order of 3×10^{-23} square centimeters per second. This value is 10 orders of magnitude less than that for maximum interaction, and it hardly seems likely that such a diffusion rate could produce a measurable fraction of the internal-friction effects observed at the most effective diffusion rate. Therefore, the physical significance of figure 4(a) is somewhat debatable.

These same data are plotted as $\log \epsilon$ versus $\frac{1}{T}$ in figure 4(b) and again appear linear within experimental error. It is apparent that figures 4(a) and 4(b) cannot each be an adequate description of the data, but the data are insufficient to establish either relation. In figure 4(b), the slope of the lines determines H_j in Cottrell's equation if the lines each represent constant diffusion rates. It is seen that the slope of the lines is not constant, but varies from about 2,800 to 17,600 calories per mole for values of ΔQ^{-1} from 100 to 10 percent. It is difficult to perceive how strain could alter the activation energy for vacancy motion to such an extent. Indeed, the experimental work of Wintenberger reported in reference 7 indicates that, although vacancies disappear more rapidly with increasing amounts of strain, their activation energy is unchanged. Wintenberger measured this activation energy H_j and found it to be 13,000 to 13,500 calories per mole in pure aluminum. This is somewhat higher than the number arrived at by Cottrell from McReynold's data (ref. 8), which was 7,500 calories per mole, but is in fair agreement with the data of Bradshaw and Pearson (ref. 9), who measured 10,000 calories per mole. Cottrell's calculation was made by determining the temperature dependence of the strain required to initiate discontinuous yielding (Portevin-Le Chatelier effect) in aluminum alloys. If the Portevin-Le Chatelier effect and the effect of Q^{-1} versus ϵ are truly related, then Cottrell's number is valid for only the initiation of the effect and not for the maximum interaction at which the equation was aimed.

While it does not seem possible at the moment to proceed further with the above analysis, there are other experimental data which support a mechanism of vacancy-accelerated diffusion. An increase in the strain rate would cause dislocations to move more rapidly. In order for solute atoms to diffuse along with the moving dislocations, the atoms must have a higher diffusion coefficient. Two examples of tests run at a higher strain rate (4×10^{-5} per second) are shown in figure 5. The damping is higher, and the minimum in the damping curve is shifted to higher strain

than that for the tests at a lower strain rate. While these data are insufficient to determine accurately the interdependence of strain rate and temperature, the data available indicate they may be related by an activation energy of approximately 12,500 calories per mole. This is in good agreement with H_j as determined by other means.

Assuming that the Q^{-1} versus ϵ phenomena are the result of vacancy-accelerated aging, the strain at which the damping begins to drop is dependent upon the number of vacancies present. In addition to their production by plastic deformation, vacancies can be produced by high temperatures. Therefore, quenching a specimen from a high temperature should permit the retention in the metal at low temperatures of a nonequilibrium number of vacancies. A test of the effect of Q^{-1} on ϵ in such a specimen should show that the strain required for the initiation of the drop in damping should be less than that in a slow-cooled or aged specimen because of the greater number of vacancies present.

Specimens of an aluminum alloy with -0.2 percent copper were quenched in ice water from temperatures between 340° and 488° C. Measurements of electrical resistivity at liquid-nitrogen temperatures were made by a standard potentiometer arrangement for comparing two nearly equal resistances. The incremental increase in resistivity was assumed to be due to an increase in the concentration of vacancies. A plot of the resistivity increment versus the reciprocal temperature is given in figure 6. The slope of this line should be H_f , the heat of formation of vacancies.

While the scatter in the data is comparatively severe, it can be seen that the slope is consistent with a value of H_f of about 12,000 calories per mole. This does not agree with the more refined measurements of Bradshaw and Pearson reported in reference 9, which probably indicates that the quenching rate was not completely effective. Nevertheless, the curve indicates the retention of thermally produced vacancies.

Mounting the specimen and starting a test for Q^{-1} versus ϵ required about 2 minutes. It was observed that not more than half of the incremental resistivity annealed out in this period, so that a significant proportion of the quenched-in vacancies were retained at the initiation of plastic deformation. Curves of Q^{-1} versus ϵ for commercial 2S aluminum specimens cooled at various rates are shown in figure 7. The drop in damping with increasing strain appears to shift toward lower strains as the cooling rate increases and has apparently disappeared entirely for the quenched sample. A series of curves showing the effects of room-temperature aging on quenched 2S aluminum specimens are shown in figure 8. Here, the drop in damping appears at greater strains as aging time at room temperature increases. This presumably is due to a decrease in the number of vacancies with time at room temperature. Therefore, these data are also in agreement with a mechanism involving vacancy-accelerated diffusion.

Damping Below Room Temperature

In view of the rather pronounced effects of straining on the damping of aluminum alloys near room temperature, a survey of the low-temperature damping was made to determine the presence of any deformation-produced phenomena.

Specimens of as-drawn wire (which had been drawn several months previously) were examined over the temperature range of 40° to -190° C. Two small peaks were observed, one at -70° C and the other at -50° C. A series of tests involving aging at increasing temperatures led to the gradual disappearance of the peaks. A series of curves for aluminum alloy containing 0.2 percent copper under these conditions are shown in figure 9. It can be seen that aging for 1/2 hour at temperatures up to 200° C has relatively little effect other than to resolve what is apparently a double peak into a single peak which reaches its maximum near -70° C. After an anneal for 1/2 hour at 250° C, the peak has decreased to about one-half of its original magnitude, and the peak has almost entirely disappeared after 1/2 hour at 275° C.

The decrease in the magnitude of the peak after an anneal at 250° C coincides with the metallographic appearance of the first recrystallized grains. That the disappearance of the peak is controlled by a time-temperature relationship is demonstrated in figure 9(e), where a 1/2-hour anneal at 200° C has only slightly decreased the magnitude of the peak while a 15-hour anneal at the same temperature has eliminated the peak.

Some attempts were made to establish the time-temperature relationships which control the annealing-out process of these peaks. Unfortunately, these were not successful. The major difficulty encountered was that the small size of the peaks and the character of the background damping did not allow sufficient resolution of changes in the peaks themselves. In addition to this, the two separate peaks behaved differently and increased the difficulty of resolution. Nevertheless, some of the pertinent characteristics of the peaks have been established.

These peaks are observed in freshly drawn wire as well as in material which has been aged at room temperature for some months. The damping peak observed in freshly drawn (75-percent reduction of area) aluminum alloy containing 0.1 percent copper is shown in figure 10 along with the curve for the same alloy which had been drawn about 18 months previously. The large peak at -50° C is apparent. However, the peak in the older wire has greatly decreased in magnitude (along with the background internal friction) and now seems to occur at -70° C. The behavior of the high-purity aluminum used in this study is remarkably similar (fig. 11). Evidently there are two separate peaks which, from all indications, do not depend upon the presence of the copper in the alloy for their existence. Indications are that the peak at -50° C is produced by deformation, but that it decays out with time at room temperature and is eventually

replaced by a peak at -70° C. As can be seen in the curves of figures 9(d), 9(e), and 9(f), this may not be true for annealing above room temperature. Again, the resolution of the data is insufficient to clarify this point.

Using the empirical relation of Wert and Marx (ref. 10), it is possible to estimate the activation energy associated with the relaxation process responsible for the peaks. This is about 13,000 to 14,000 calories per mole. This value of the activation energy is of immediate interest since it is close to the energy observed for the motion of vacancies H_j in aluminum.

Activation energies of relaxation peaks can be determined experimentally by measuring the change in peak temperature with different frequencies of oscillation. This was done on the aluminum alloy containing 0.5 percent copper over a range of 0.19 to 1.6 cycles per second with the surprising result that practically no temperature shift was observed. The observed shift, for an activation energy of about 13,000 calories per mole, should have been about 15° C for the frequency range employed. The shift actually amounted to less than 4° C, which is approaching the experimental error in the resolution of the peak maximum. Apparently the process responsible for the peaks does not conform to the usual form of relaxation analysis.

An estimate of the activation energy can also be made from the shape of the damping peak. The scatter in the data is again severe, but the peak shapes, in general, are close to what one might expect for an activation energy of about 13,000 calories per mole.

Because this activation energy is similar to that for the motion of vacancies in aluminum, some attempts were made to observe the peaks in freshly quenched specimens which contained an excess number of thermally produced vacancies. Some indications of a peak at -50° C were seen, but the results were ambiguous because of a marked instability in the damping. Thus, the results were inconclusive.

The peaks, therefore, have at least four distinguishing characteristics. These are:

- (1) Plastic deformation is required to produce them.
- (2) Temperatures approaching those for recrystallization are required to remove them.
- (3) Their presence does not depend upon impurity content.
- (4) They do not respond to ordinary relaxation analysis.

That deformation is required to produce the peaks and that the peaks seem relatively unaffected by solute (copper) concentration indicate that

the mechanism producing the peaks is associated with deformation-produced defects. Ignoring stacking faults (since the stacking-fault energy is comparatively high in aluminum), it seems probable that either vacancy or dislocation motion is essential to the mechanism. Though attempts to produce the peaks by quenching in a nonequilibrium number of vacancies were inconclusive, they would not have disproven the vacancy mechanism even if the peaks had not appeared at all. In quenching, vacancies would be produced in volumes of low dislocation density, while the vacancy environment would be substantially different in a drawn wire. It is, however, rather difficult to reconcile the lack of a shift in peak temperature with frequency to a vacancy mechanism. It would be appropriate to test the behavior of these peaks over a frequency range much wider than that available during this work to establish with certainty what relation, if any, the peak temperature has to frequency.

A number of surprisingly similar peaks have been observed in certain face-centered-cubic metals, but usually at much lower temperatures and much higher frequencies. Niblett and Wilks (ref. 11) have observed peaks in cold-worked copper at -200° and -240° C at a frequency of 1,100 cycles per second. Bordoni (ref. 12), using a frequency of about 40 kilocycles, found a single peak at -170° C in plastically deformed aluminum and observed similar peaks in lead, silver, copper, and aluminum. Several theories, all including dislocation motion, have been put forth to account for these peaks. Perhaps the best is that of Seeger (ref. 13), who ascribes the peaks to segments of dislocation lines which under stress form bulges (jogs) which then spread along the length of the dislocation line in a direction parallel to the line.

The peaks described in this paper are similar in each of their distinguishing characteristics, except the last (frequency dependence), to those lower temperature peaks to which Seeger's theory applies. A more thorough investigation of the last distinguishing characteristic may eliminate even this apparent difference. It is probable, then, that a mechanism similar to that proposed by Seeger is responsible for the reported peaks.

It is unlikely, however, that precisely the same mechanism as that of Seeger is involved. These presently reported peaks appear to exist in addition to those peaks specifically discussed by Seeger. Niblett and Wilks (ref. 14) in a more recent paper have reported a peak near -70° C in deformed high-purity copper. They offer no explanation for this peak. Again, this peak is present in addition to the low-temperature peaks discussed by Seeger.

On the basis of the experimental evidence, it does not seem possible to account adequately for the peaks observed near -70° C in both aluminum and copper. It is possible that the peaks result from the interaction of dislocations with vacancies. In effect, the vacancies would have to be

dragged along with the moving dislocations. A somewhat analogous situation exists in iron, where carbon atoms are dragged along with moving dislocations (ref. 2) and produce a damping peak. It is encouraging that the activation energy for this "interaction" peak in iron is significantly higher than that for the interstitial diffusion of carbon in iron. Such a mechanism implies, however, that vacancies (or aggregates of vacancies) maintain their existence up into the recrystallization-temperature range.

CONCLUSIONS

Damping during plastic deformation of aluminum and aluminum-copper alloys has four distinguishing characteristics. These are:

1. Other things being equal, the damping is directly proportional to the strain rate and to the vibrational period.

2. The presence of copper in the aluminum lowers the damping from that of pure aluminum and introduces temperature-dependent changes.

3. These temperature-dependent changes are altered by an altered strain rate.

4. These temperature-dependent changes are further altered by heat treatment, specifically, by quenching, to retain a nonequilibrium concentration of vacancies.

These observations are interpreted as resulting primarily from the acceleration of the diffusion rate of solute copper by deformation-produced vacancies and the interaction of the copper with moving dislocations.

The damping maxima observed below room temperature in these same alloys also have four distinguishing characteristics. These are:

1. The peaks are produced by plastic deformation.

2. Relatively high temperatures (near recrystallization) are required to destroy the peaks.

3. The presence of copper is not essential to the presence of the peaks.

4. The peaks do not conform to ordinary relaxation analyses.

It is suggested that these peaks result from the interaction of vacancies (or aggregates of vacancies) with dislocation jogs.

Battelle Memorial Institute,
Columbus, Ohio, May 15, 1958.

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TABLE 1.- SPECTROGRAPHIC ANALYSIS
OF HIGH-PURITY ALUMINUM

Element	Content, weight percent
Silicon	0.01
Iron	.002
Magnesium	.001
Copper	.003
Nickel	.001
Manganese	.001

TABLE 2.- ANALYSIS AND HEAT TREATMENT OF EXPERIMENTAL ALLOYS

Alloy	Copper solute addition, weight percent		Recrystallization time, hr	Recrystallization temperature, °C
	Nominal	Special analysis		
High purity	----	See table 1	1/2	475
No. 783	0.05	0.050	1/2	500
No. 787	.1	.090	1/2	500
No. 788	.2	.200	1/2	500

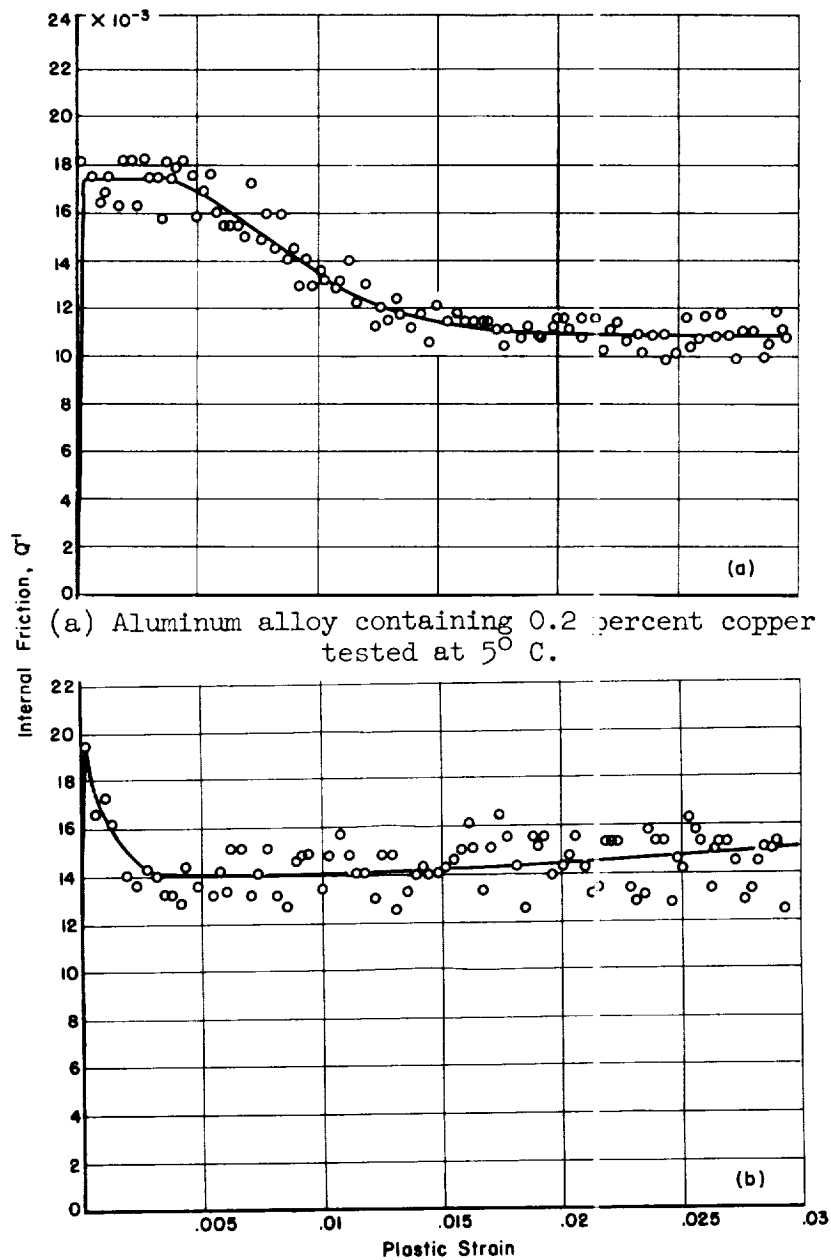
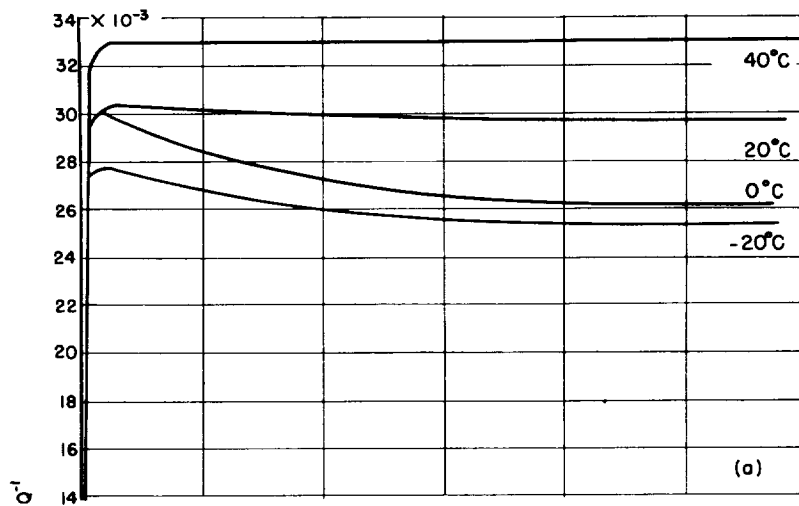
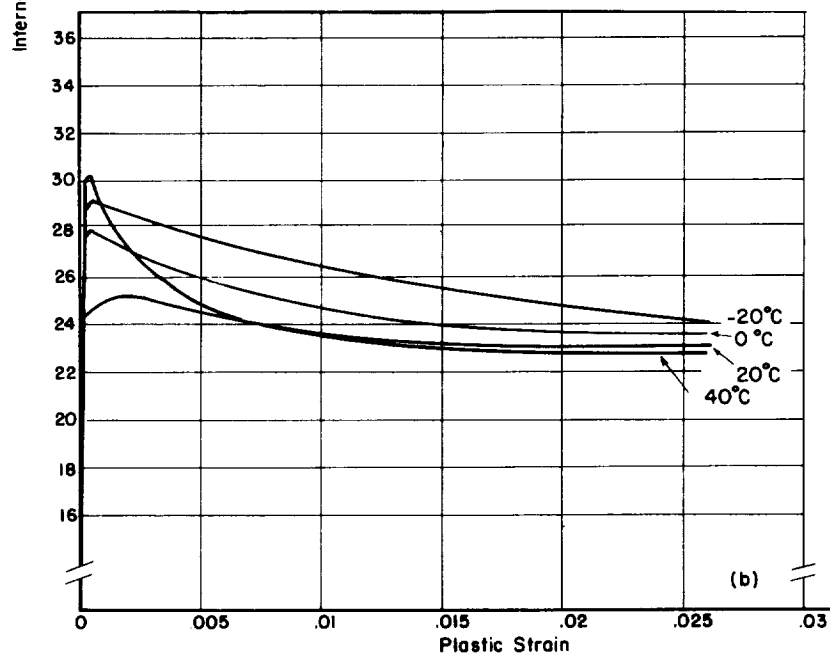


Figure 1.- Examples of scatter observed in simultaneous damping and strain tests for two alloys.

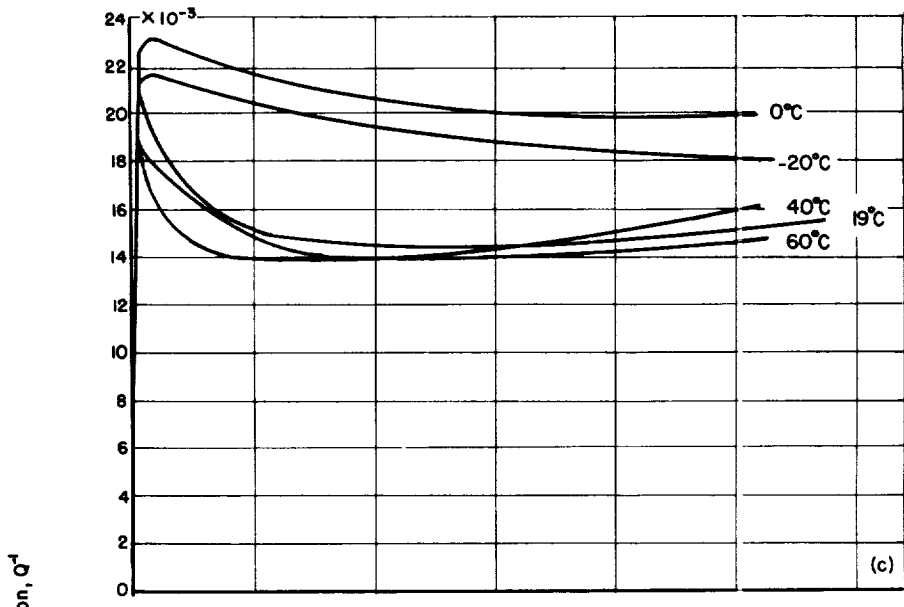


(a) High-purity aluminum tested at temperatures between -20° and 40° C.

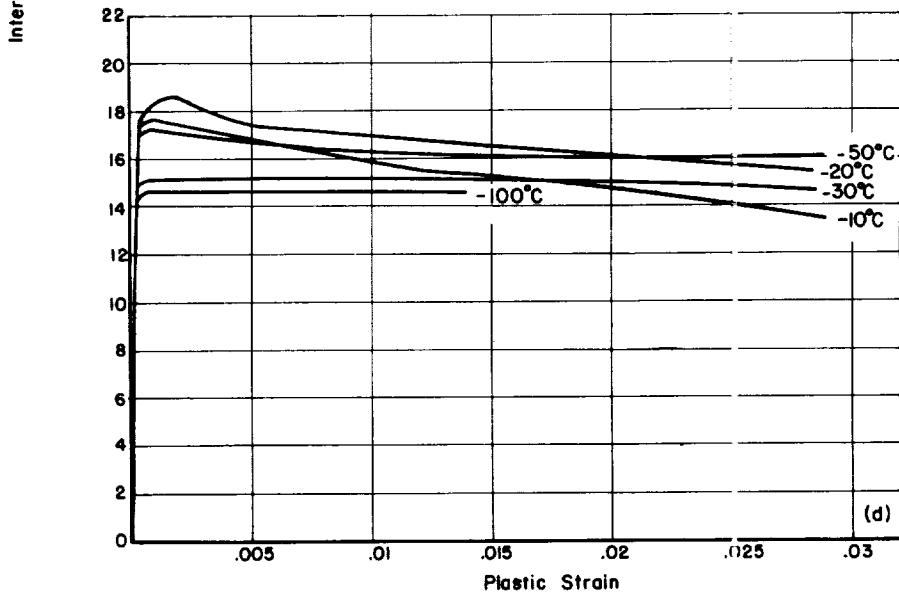


(b) Aluminum alloy containing 0.05 percent copper tested at temperatures between -20° and 40° C.

Figure 2.- Temperature dependence of simultaneous damping and strain measurements for various aluminum-copper alloys.

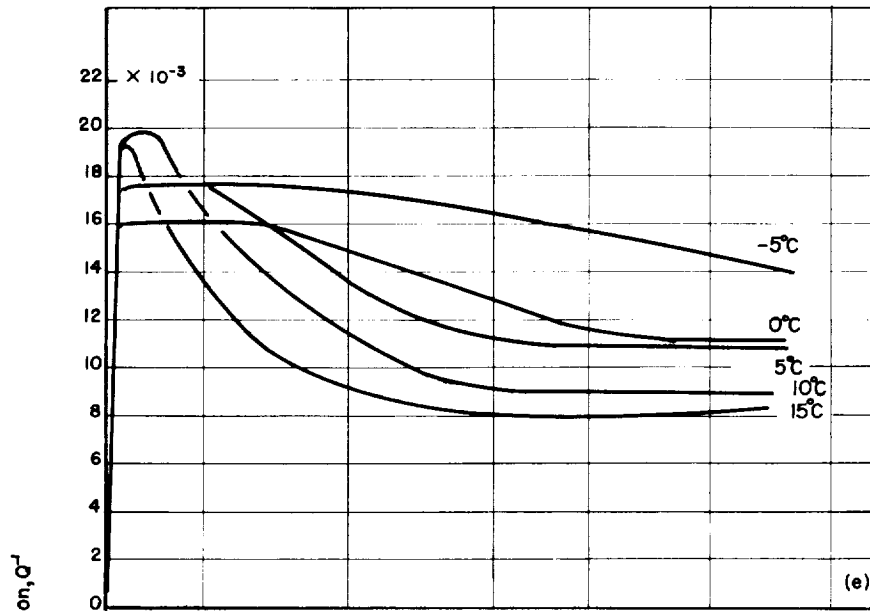


(c) Aluminum alloy containing 0.1 percent copper tested at temperatures between -20° and 60° C.

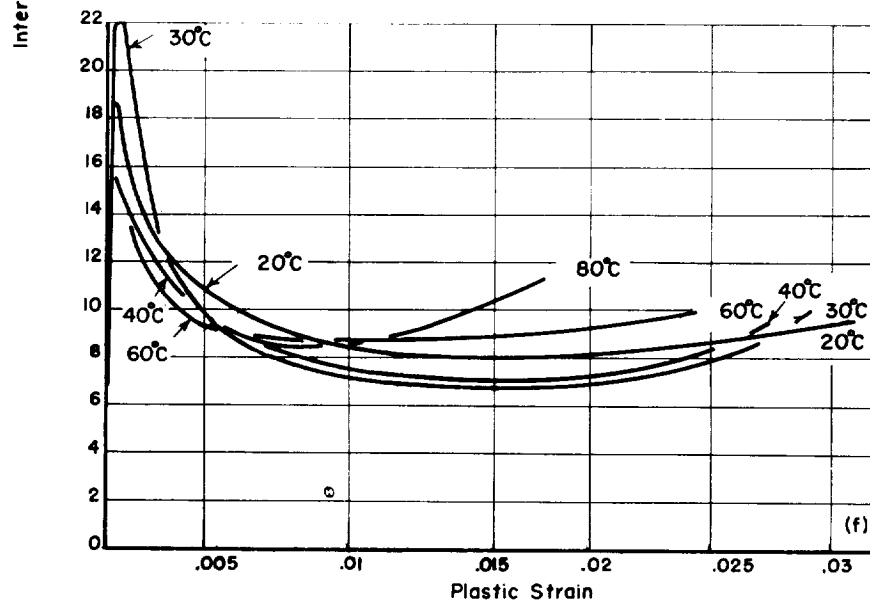


(d) Aluminum alloy containing 0.2 percent copper tested at temperatures between -100° and -10° C.

Figure 2.- Continued.

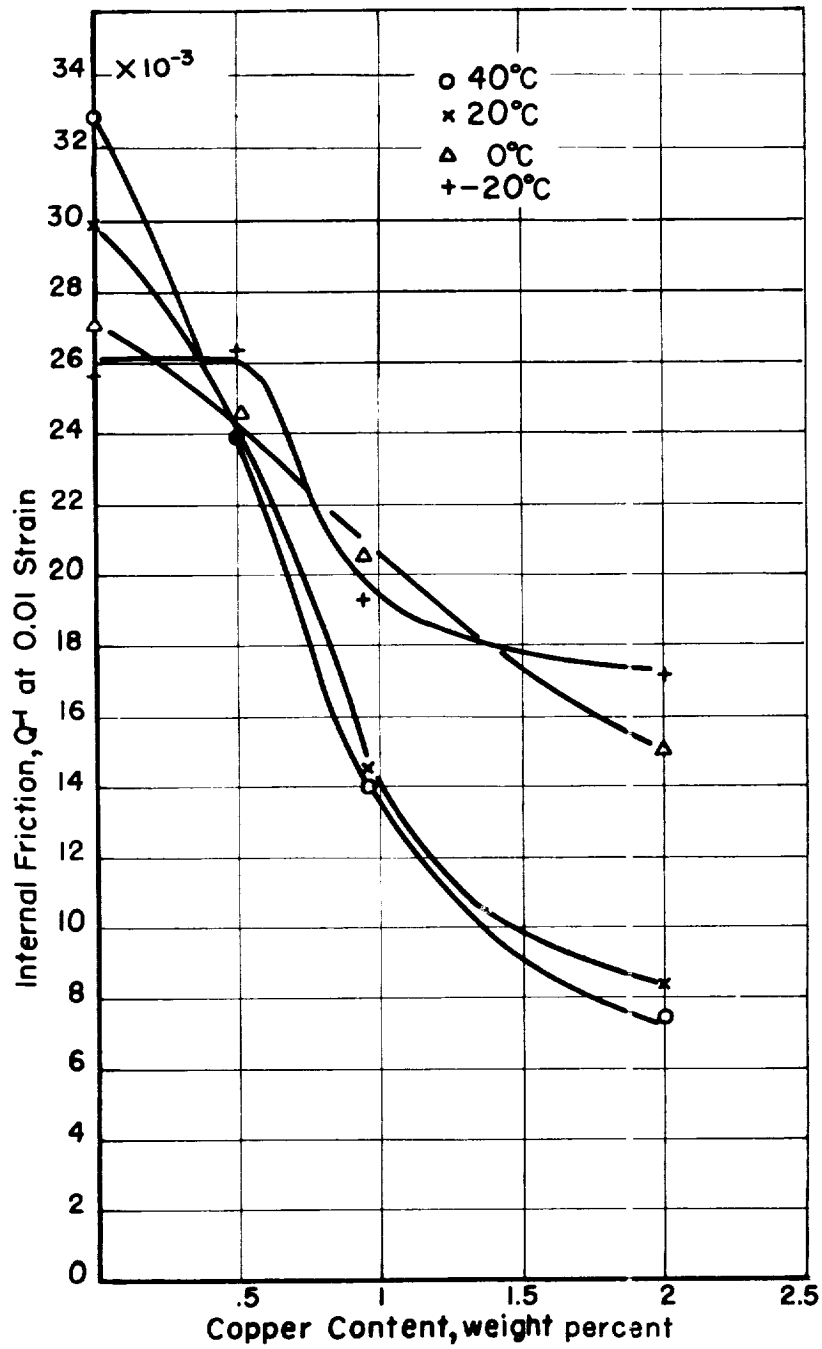


(e) Aluminum alloy containing 0.2 percent copper tested at temperatures between -5° and 15° C.



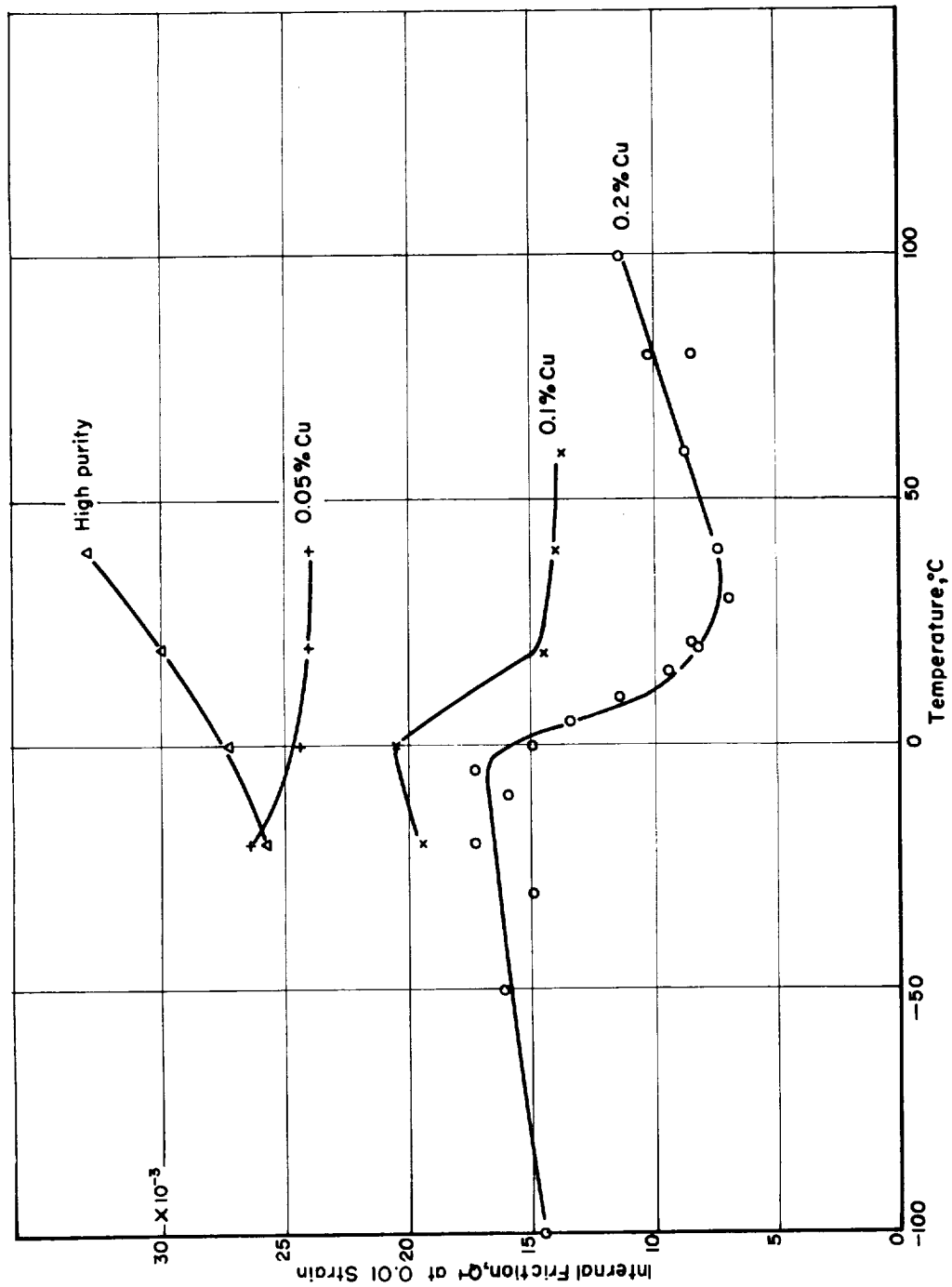
(f) Aluminum alloy containing 0.2 percent copper tested at temperatures between 20° and 80° C.

Figure 2.- Concluded.



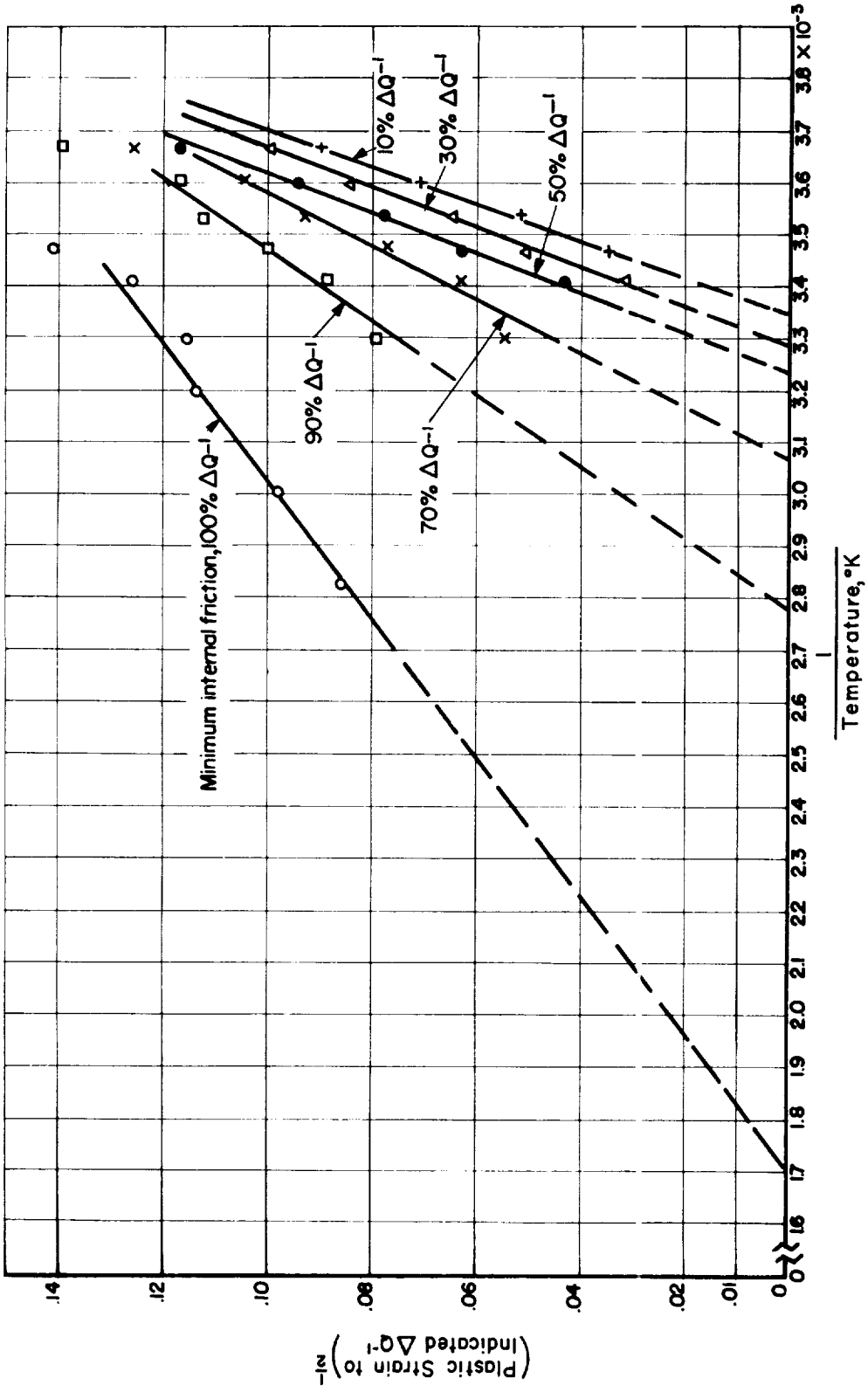
(a) Curves for different temperatures.

Figure 3.- Effect of temperature and alloy content on simultaneous damping and strain measurements.



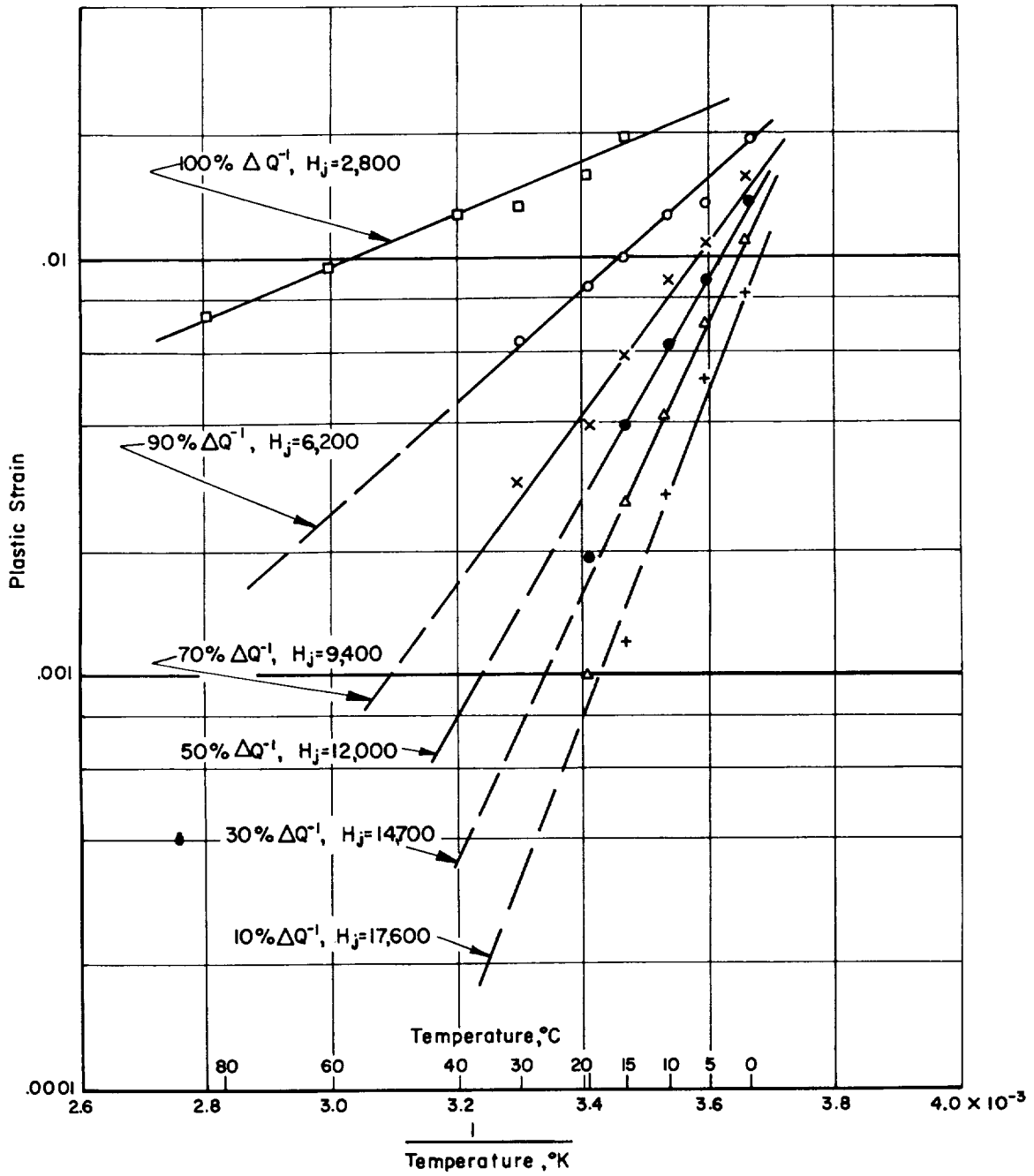
(b) Curves for different alloy contents.

Figure 3.- Concluded.



(a) (Plastic strain to indicated ΔQ^{-1})^{1/2} used as ordinate scale.

Figure 4.- Temperature dependence of stages in simultaneous damping and strain measurements.



(b) Plastic strain used as ordinate scale; replot of data in figure 4(a).

Figure 4.- Concluded.

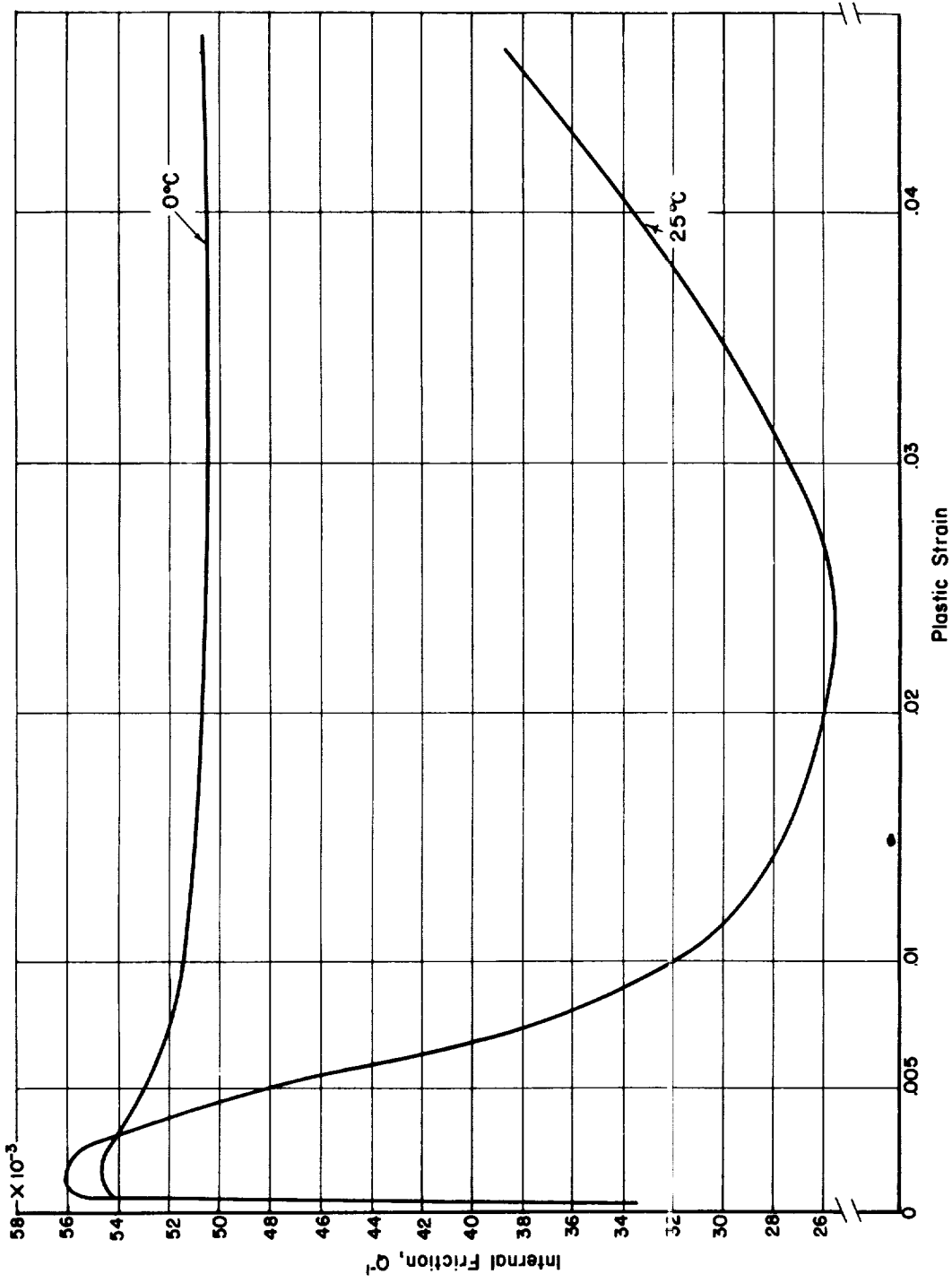


Figure 5.- Effect of temperature on simultaneous damping and strain measurements for increased strain rate of aluminum alloy containing 0.2 percent copper. $\dot{\epsilon} = 4 \times 10^{-5}$ per second.

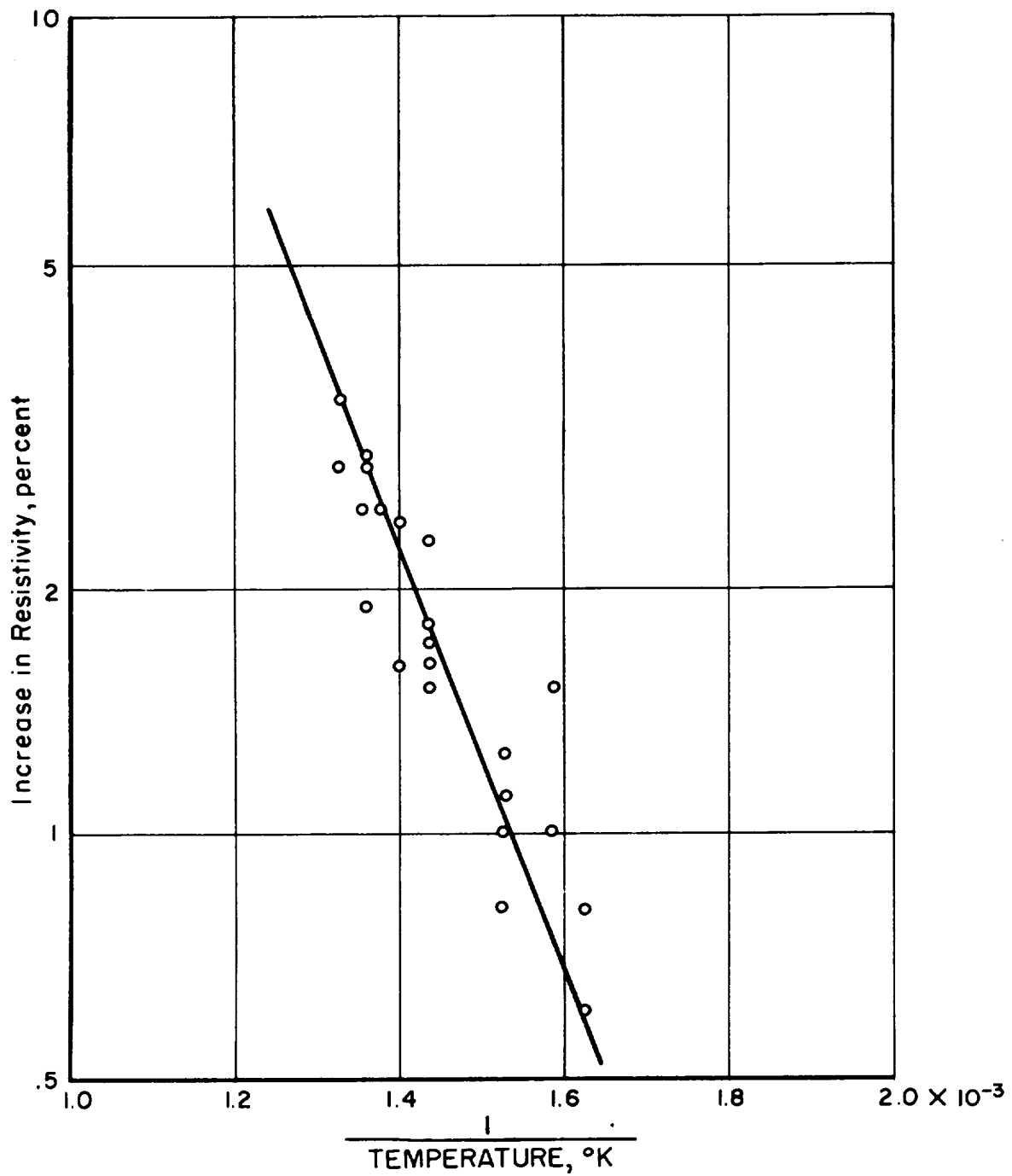


Figure 6.- Dependence of resistivity change on quenching temperature for aluminum alloy containing 0.2 percent copper. Slope of curve is $H_f \approx 12,000$ calories per mole.

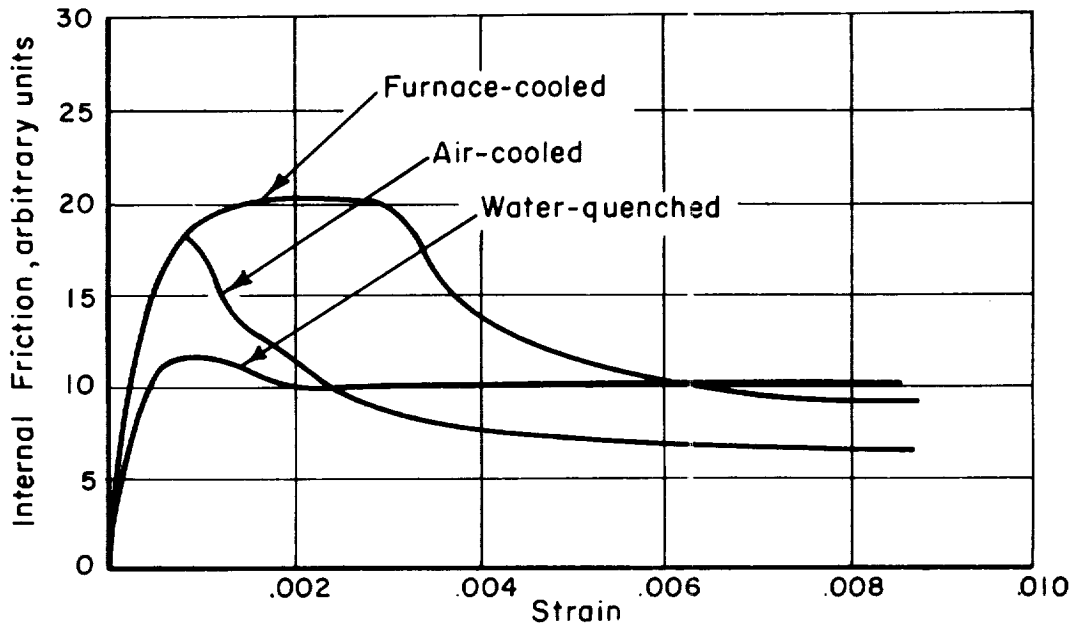


Figure 7.- Dependence of simultaneous damping and strain measurements on cooling rates for commercial 2S aluminum.

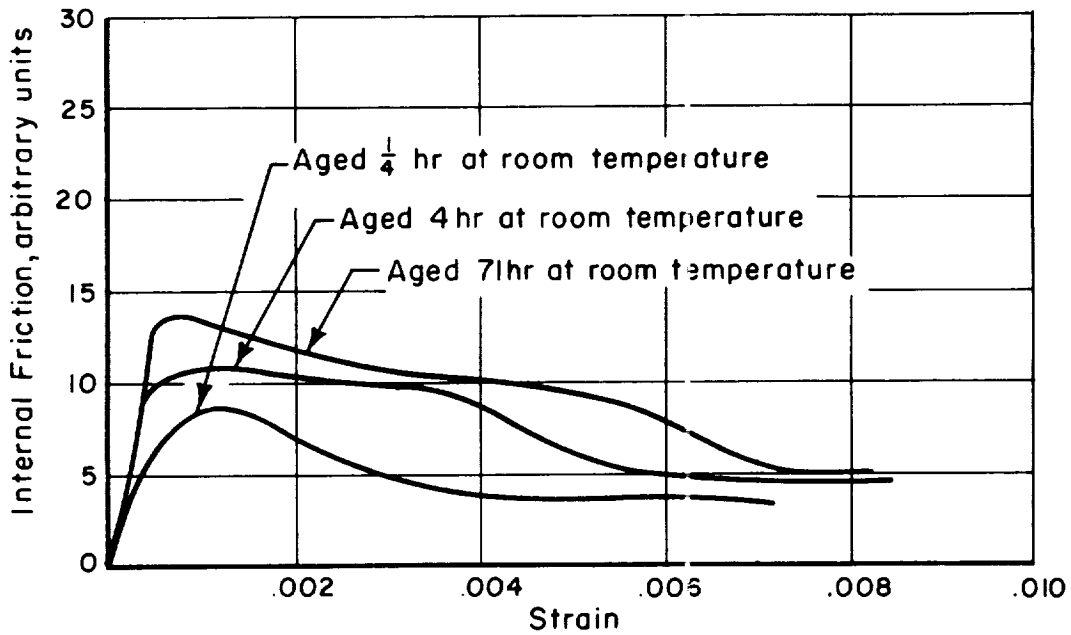
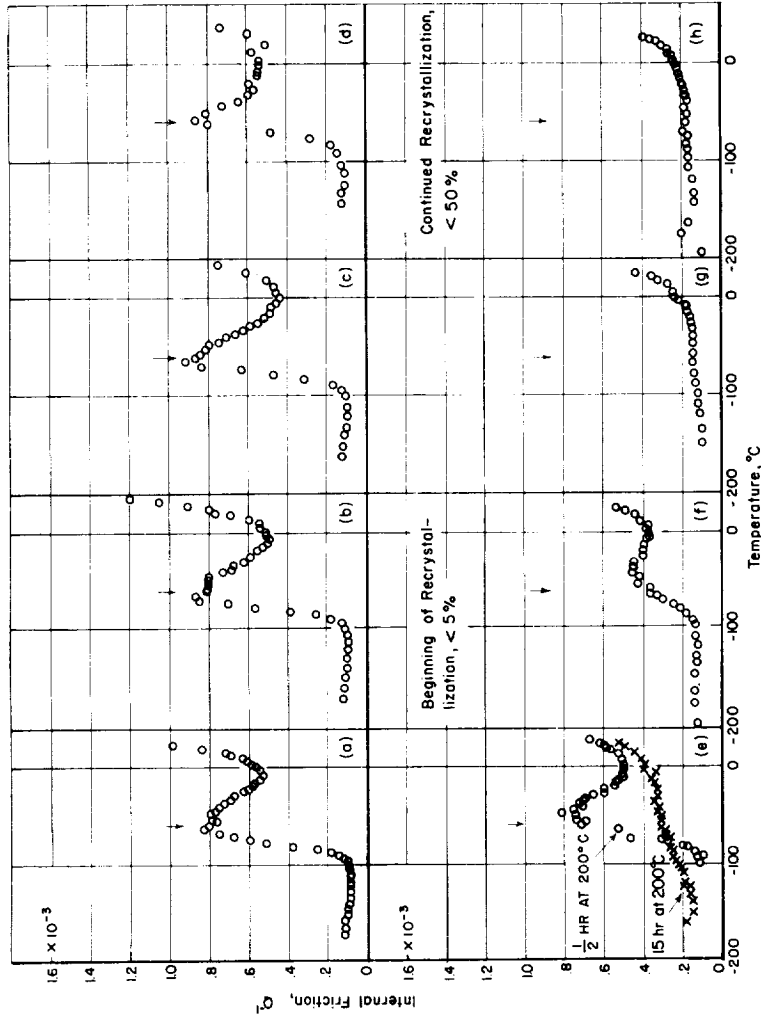


Figure 8.- Dependence of simultaneous damping and strain measurements on room-temperature aging after quenching for commercial 2S aluminum.



- (a) As-drawn.
- (b) Aged 1/2 hour at 50 $^{\circ}C$.
- (c) Aged 1/2 hour at 100 $^{\circ}C$.
- (d) Aged 1/2 hour at 150 $^{\circ}C$.
- (e) Aged 1/2 hour and 15 hours at 200 $^{\circ}C$.
- (f) Aged 1/2 hour at 250 $^{\circ}C$.
- (g) Aged 1/2 hour at 275 $^{\circ}C$.
- (h) Aged 1/2 hour at 300 $^{\circ}C$.

Figure 9.- Effect of annealing on low-temperature peaks for aluminum alloy containing 0.2 percent copper. Arrow is reference mark.

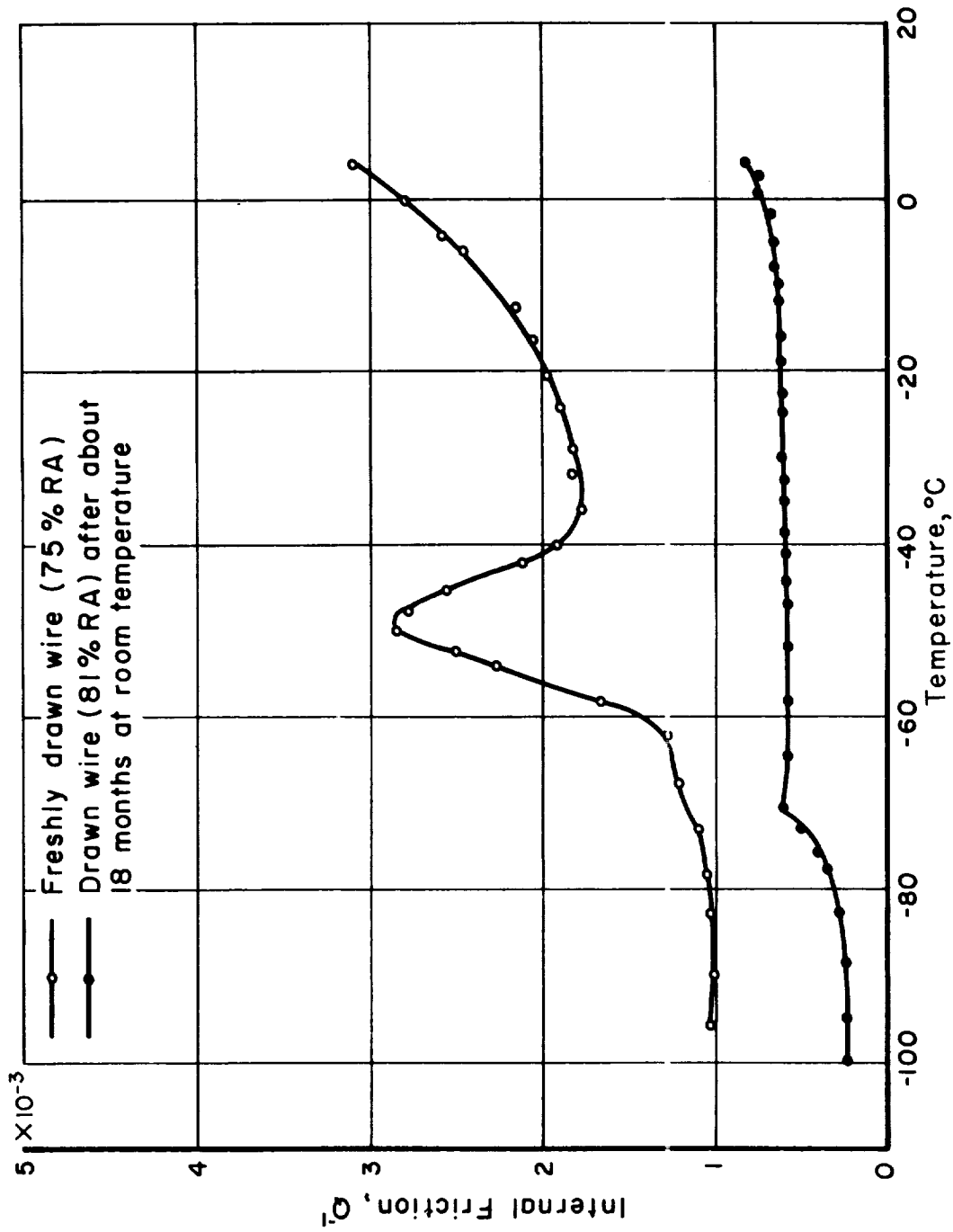


Figure 10.- Effect of room-temperature aging on internal friction of aluminum alloy containing 0.1 percent copper. RA, reduction of area.

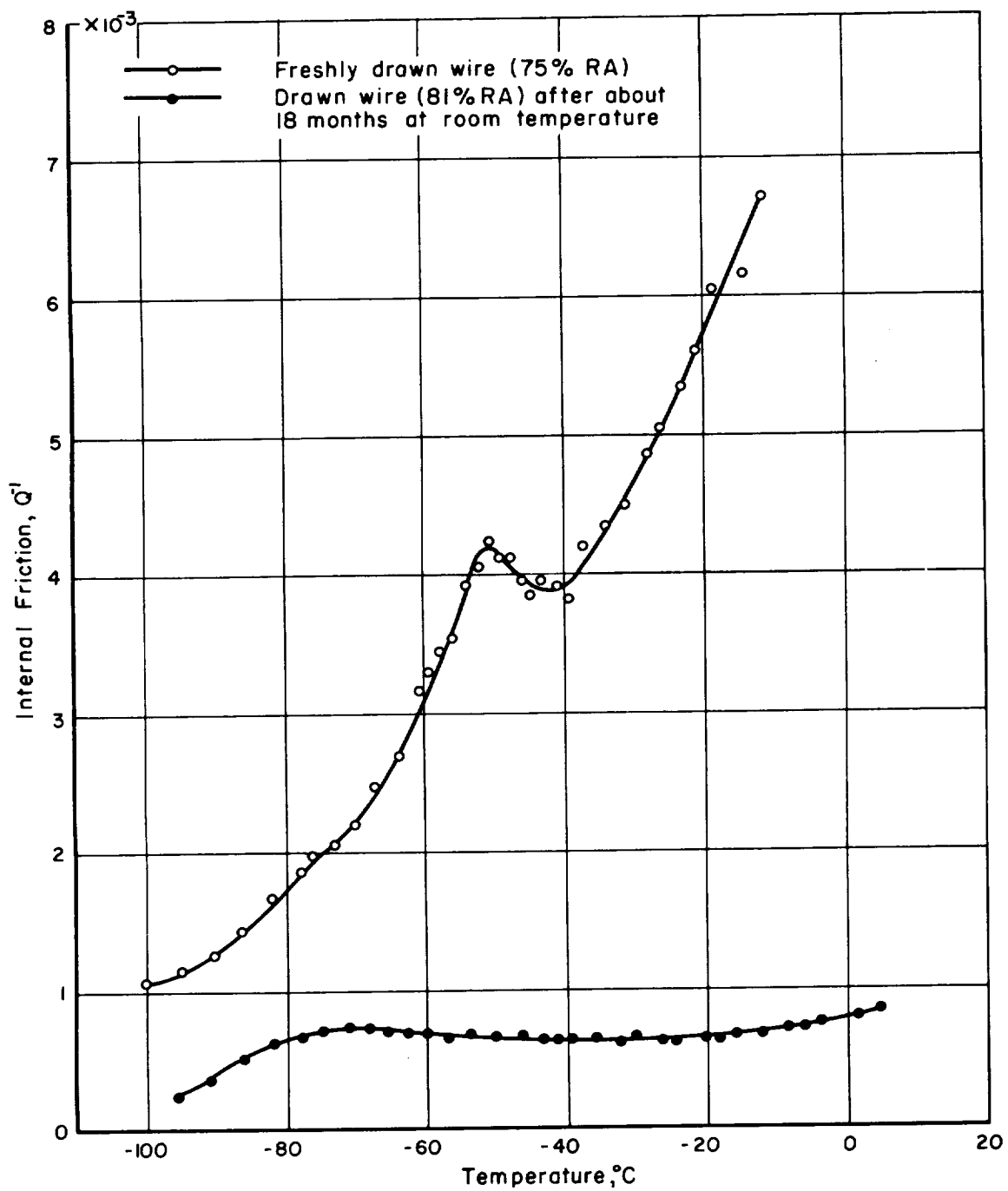


Figure 11.- Effect of room-temperature aging on internal friction of high-purity aluminum. RA, reduction of area.

