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INVESTIGATION OF PLASTIC BEHAVIOR OF BINARY ALUMINUM

ALLOYS BY INTERNAL-FRICTION METHODS

By R. E. Maringer, L. L. Marsh, and G. K. Manning

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SUMMARY

Internal-friction tests have shown that nominal additions of 0.1, 1.0, and 5.0 weight percent copper and magnesium to aluminum have significant effects on damping behavior. Small additions of copper to aluminum raised the internal friction and lowered the shear modulus above 200° C. Large additions of copper and of magnesium raised the shear modulus and lowered the damping.

An "anomalous" internal-friction peak at about 125° C has been observed in aluminum containing 1.00 percent copper and is believed to result from the interaction of dislocations and impurity particles. Other small peaks which were observed in high copper and magnesium alloys appear to result from the stress-induced diffusion of impurity atoms. Aging is shown to decrease the internal friction in most cases, probably because of the segregation of foreign atoms at grain boundaries and/or subboundaries.

The internal friction during plastic deformation was sensitive to strain rate, amount of strain, alloy content, heat treatment, temperature, and the frequency at which the measurements are made. The internal friction decreased with increasing temperature for alloys containing 0.08 percent copper and 0.09 percent magnesium but increased with increasing temperature for richer alloys.

INTRODUCTION

The observation that the stress required to produce plastic deformation was less by a factor of about 1,000 than the theoretically predicted stress prompted the development of the dislocation theory. The validity of this theory is now widely accepted, partly because of the striking confirmation of theoretical predictions by physical observations.

A dislocation has been defined as "a line imperfection forming the boundary within the crystal of a slipped area" (ref. 1). The dislocation

is thus a region of strain. According to the theory (refs. 1 to 5) an impurity atom can lower the strain energy of this system upon entering the strained region around a dislocation. Since the dislocation must then increase the strain energy of the system in order to move away from the impurity atoms, it is, in effect, anchored. Atoms which anchor dislocations are said to form atmospheres around the dislocations.

When a stress is applied to a dislocation, the dislocation will tend to move in the direction of the applied stress. Because of the restraining effect of the atmosphere of impurity atoms, the dislocation will tend to bow, somewhat in the manner of a vibrating string. When the stress is sufficiently great, the dislocation will snap free from its atmosphere and proceed through the lattice under a reduced stress. Frank-Read sources can then begin to generate a large number of dislocations. This behavior, the discontinuous yield point, is commonly observed in the mechanical behavior of body-centered cubic metals where it has been attributed to the presence of small amounts of carbon and/or nitrogen.

It was predicted from the dislocation theory (ref. 6) that, below a critical temperature, discontinuous yielding should also be observed in substitutional alloys in other crystal structures. This has been amply verified in a number of alloys such as brass (ref. 6), 6-percent-tin phosphor bronze (ref. 7), and aluminum-magnesium alloys (refs. 8 and 9). Associated with discontinuous yielding in many cases is the serrated stress-strain curve. The serrations are believed to result when dislocation atmospheres are permitted to reform during straining (strain-aging), thus causing, in effect, the reoccurrence of the discontinuous yield.

Sherby, Anderson, and Dorn (ref. 10) have shown that additions of copper and magnesium to aluminum cause the formation of a plateau in the curves of flow stress versus temperature of these alloys. This plateau can be adequately described in assuming the interaction of migrating impurity atoms and dislocations.

These and numerous other examples of mechanical behavior are available which can be described in terms of dislocation and impurity-atom interaction. There are also a number of examples in the literature which indicate that internal friction is also sensitive to this dislocation-impurity interaction.

Internal friction is the ability of a material, completely isolated from its surroundings, to dissipate vibrational energy. Internal friction, or damping capacity as it is often called, is usually observed by setting a specimen into vibration and determining the rate of decay of the amplitude of vibration. This is conveniently done by using the specimen in the form of a wire as the suspension of a torsional pendulum.

In the spectra of internal friction versus temperature of metals, a
number of different types of maximums are to be found. One type of peak
is observed in metals which have first been cold worked. This type of
peak is believed to be related to the same mechanism which gives rise
to the plateau in the curves of flow stress versus temperature of alumi-
num alloys (ref. 10). This is believed to be the dragging along of
anchoring atoms by moving dislocations. This type of peak has been
observed in iron (refs. 11 and 12) and molybdenum (ref. 13) containing
carbon and nitrogen and in aluminum containing copper (ref. 14). Other
types of peaks, such as those due to the stress-induced diffusion of
interstitial or substitutional impurities, and the stress-induced relax-
ation along grain boundaries are also observed in metals and can be of
inestimable help in determining diffusion rates and the location of
various impurities.

While this is a relatively new technique, it appears that much can
be learned by the observation of the internal friction during plastic
deformation. The work which has been done (refs. 15 and 16) indicates
that, again, the internal friction results from the interaction of
dislocations and impurity atoms. It seems possible that most, if not
all, the phenomena which affect the mechanical properties of metals
during plastic deformation have a counterpart in internal-friction
behavior. This investigation resulted from the hope that the remarkable
sensitivity of the internal-friction method might make it possible to
observe and resolve effects which might be masked by other phenomena in
more conventional testing procedures.

This investigation was conducted at Battelle Memorial Institute
under the sponsorship and with the financial assistance of the National
Advisory Committee for Aeronautics. The authors are indebted to
Messrs. J. D. Bray, C. L. Seale, and E. B. Swetnam for their assistance
in the design and construction of the apparatus and in the carrying out
of the experiments and to Professor B. Chalmers for several interesting
discussions on the subject matter.

EQUIPMENT

Internal-friction measurements were made in two torsional pendulums.
These were patterned after that of Kê (ref. 11) and consist of a rigid
upper grip, a wire specimen, and a free-swinging lower grip containing
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
of from 120 to 290 psi, 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 about 2×10^{-5}.

In order to raise the specimen rapidly to temperature for aging tests, the furnace of the internal-friction apparatus was heated to several degrees above the desired aging temperature. The specimens were inserted in the hot furnace and attained within 1° C of the desired temperature within about 6 or 7 minutes depending upon the aging temperature. The temperature gradient along the length of an 8-inch specimen did not vary by more than $\pm 2^{\circ}$ C. The furnace temperature was controlled by a Foxboro controller which was manipulated manually until the specimen had reached the operating temperature. Temperature control was to within 1° C for short periods but varied by several degrees when room temperature changed appreciably. The temperature was always adjusted manually to the aging temperature before readings were taken when the difference from the aging temperature exceeded 1° C.

An apparatus was constructed in order to observe the internal friction of a wire specimen undergoing plastic tensile deformation. A schematic diagram of this apparatus is shown in figure 1. The apparatus consists, in essence, of a torsional pendulum mounted in a tensile-test machine. A load cell to record stress is attached to a rigid plate. Below the load cell a grip holds a dummy wire which in this case is a 4-inch length of cold-drawn brass wire. This wire supports the inertia member of the torsional pendulum. A grip on the lower end of the inertia member supports the upper end of the specimen. The lower end of the specimen is held in a grip which is in turn attached to a movable plate. The movable plate can move at various constant rates so as to subject the specimen to a tensile load at various strain rates. The strain rates used in the course of this investigation varied from 0.00026 to 0.0026 per minute. These rates can be increased or decreased by a simple rearrangement of pulleys. Stress is recorded on one side of a two-channel Sanborn recorder. A microswitch, inserted in the gear train, causes a mark on the record for every 0.0005 inch the movable plate moves. A timer puts a similar mark on the chart every 10 seconds so that strain rate may be computed from the record.

The oscillations of the pendulum are recorded on the second channel of the Sanborn recorder. The physical arrangement for picking up the vibrations is shown in figure 2. The vibration of the pendulum causes a shadow to move in front of the phototubes. This causes an increase of light intensity on one tube and a decrease of light intensity on the other.

The phototubes are connected in a bridge circuit (fig. 3); therefore, the output voltage of the bridge varies as a function of the shadow and

as a function of the pendulum vibration. This voltage is a linear function of the amplitude of oscillation over the range required. A direct-current amplifier follows the phototube bridge circuit to increase the signal strength and match the input impedance of the Sanborn recorder into which it is fed.

The light for the phototubes is furnished by a 32-candle-power automobile headlamp bulb. The lamp is powered by a direct-current source as diagrammed schematically in figure 4. The torsional oscillation is initiated by the use of an electromagnet attracting one end of the torque bar.

The recorded damping curve is analyzed to obtain the time required to decay from a given amplitude to half that amplitude, and the frequency of the vibration is determined. From this, and a knowledge of the damping and frequency of oscillation of the pendulum without the specimen, the internal friction of the specimen can be determined from the relation derived by Boulanger (ref. 17)

$$T^2 \delta'' = (T'')^2 \delta + \delta' \left[T^2 - (T'')^2 \right] \tag{1}$$

where

T   period of pendulum without specimen

T"   period of entire assembly

δ damping of pendulum without specimen

δ' damping of specimen alone

δ'' damping of entire assembly

MATERIALS

Wire for some of these experiments was drawn from specimens used for a previous study of creep. The higher alloys containing nominally 1 percent Cu, 5 percent Cu, and 5 percent Mg were prepared from high-purity aluminum (see analysis, tables 1 and 2), electrolytic copper, and 99.8 percent cell magnesium. The melts were made in a machined graphite crucible and were cast in carbon molds, making sure that a minimum of pipe was obtained in each ingot. The billets were soaked overnight at 510° C in a circulating-air furnace and were extruded to 3/8-inch rounds at a temperature of 480° to 520° C.

The richer alloy wires were drawn directly from the extrusion, using 10-percent reduction with each pass. An intermediate anneal of 1/2 hour at 343° C was necessary to complete drawing down to 0.060-inch diameter.

The lower alloys, containing nominally 0.1 percent Cu, 0.1 percent Mg, and 1 percent Mg, and the high-purity aluminum Al, were drawn to wire directly from 1/2-inch-diameter creep specimens, again using a 10-percent reduction per pass but with no intermediate anneal.

In order to obtain a common grain size, all specimens were heat-treated as shown in table 3. This grain size was determined by the grain size of the alloy containing 5 percent Cu after 1 hour at 540° C. Heat treatment was carried out in a wire-wound furnace having a temperature gradient of less than ±2° C over the length of the specimen. Specimens were sealed in vacuum in Pyrex, heat-treated, and quenched in water at room temperature. This produced an average grain diameter of approximately 0.2 millimeter. Unless otherwise stated, all specimens were given this heat treatment.

Specimens are designated by a number identifying the machine in which they were tested (1, 2, or 3), a letter signifying the alloy (M - magnesium, C - copper, HP - high purity), a number signifying nominal alloy content (1 for 0.1 percent, 10 for 1.0 percent, and 50 for 5.0 percent), and a specimen number. For example, 1C1-25 is specimen number 25 of a 0.08 percent Cu alloy tested in the number 1 machine.

EXPERIMENTAL RESULTS

Temperature Dependence of Internal Friction and Shear Modulus

Tests of internal friction versus temperature were run on a series of aluminum-copper alloys heat-treated as shown in table 3, water quenched, and aged for 17 hours at 300° C. These data are plotted in figure 5. The internal-friction curves rise and, except in the case of the 0.08 percent Cu alloy, pass through a maximum between 300° to 400° C. This is believed to be a result of the stress-induced relaxation along grain boundaries. At comparatively high temperatures, adjacent grains can slip relative to each other in the plane of their mutual boundary. This relaxation leads to the formation of an internal-friction peak. It is significant to note that, above 200° C, the internal friction of the alloy containing 0.08 percent Cu is higher than that of the high-purity aluminum. This suggests that minute additions of copper to aluminum can actually increase the rate of grain-boundary relaxation. When the copper addition is increased to 1.00 percent, the temperature at which the grain-boundary peak occurs is increased and the magnitude of the peak is decreased. This could

follow the segregation of copper to the aluminum grain boundaries. At about 250° C, there is a smaller peak, the origin of which is uncertain. It is probable, however, that it has as its source relaxation in precipitate particles. A similar peak, although not so large, occurs at about 170° C in the 4.82 percent Cu alloys.

Tests of shear modulus versus temperature of these same specimens show a similar effect of additions of copper (fig. 6). The same grain-boundary relaxation which leads to the formation of the grain-boundary peak in internal friction is evidenced by a discontinuity in the shear-modulus curve. With an addition of 0.08 percent Cu, the shear modulus decreases above about 220° C, supporting the internal-friction observation that small additions of copper can increase the rate of grain-boundary relaxation. Additions of 1.00 and 4.82 percent Cu cause increases in the shear modulus, particularly at the higher temperatures where the additions appear to inhibit grain-boundary relaxation.

A similar series of tests was conducted on the aluminum-magnesium alloys. These alloys were heat treated as shown in table 3, water quenched, and aged for 17 hours at 300° C. The results of tests of internal friction versus temperature on this series of alloys are shown in figure 7. In this case, the internal friction is affected by an addition of 0.09 percent Mg only below about 200° C. Above this temperature, the curve is very close to that for the high-purity aluminum. Additions of 0.78 percent and 5.45 percent Mg depress the grain-boundary peak, as do the copper additions, except that for magnesium the depression is more gradual. Again, in the 5.45 percent Mg alloy there is an unexplained peak superimposed upon the grain-boundary peak at about 320° C. This peak is probably also the result of a relaxation process in precipitate particles. In both the Al-Cu and Al-Mg alloys, some small peaks occur at lower temperatures and will be discussed later.

The curves of shear modulus versus temperature of the aluminum-magnesium alloys are shown in figure 8. The shear modulus appears to be lowered below 200° C by an addition of 0.09 percent Mg. Above this temperature the shear modulus for the alloy is equal to that for the high-purity aluminum. Increasing the magnesium content increases the shear modulus, especially above 200° C, but not so drastically as do similar additions of copper.

These data are consistent with those of Starr, Vicars, Goldberg, and Dorn (ref. 18), which show a slight lowering of the ratio of the observed shear modulus to the unrelaxed shear modulus in aluminum alloys containing 0.029, 0.054, and 0.101 atomic percent Cu. The data indicate that the rate of relaxation of the grain boundaries can be increased by small additions of copper to the grain boundaries. Austin, St. John, and Lindsay (ref. 19) show that additions of less than 1 percent cobalt to ferrite lower the stress required to produce 0.1 percent creep in

1,000 hours at 427° C. Similarly, Sherby, Anderson, and Dorn (ref. 10) show that the flow stress of an aluminum alloy containing 0.082 atomic percent germanium is lower at 27° C than the flow stress of an aluminum alloy containing 0.033 atomic percent germanium. It is possible that these are examples of the mechanical effects of such a grain-boundary weakening.

The activation energy for grain-boundary relaxation in the high-purity aluminum as obtained from internal-friction data was found to be 37,400 calories per mole. This is in good agreement with the previously observed 38,000 calories per mole (ref. 18) calculated from shear-modulus relaxation.

Aging Characteristics of Alloys

A number of alloys were heat-treated for 1 hour at 500° C, water quenched, and inserted in the internal-friction apparatus, and readings were taken while the wire was being heated at approximately 4° C per minute. These data should then represent the internal friction of a relatively unaged specimen. Following this, the specimens either were cooled and the run was repeated or they were aged at about 250° C over-night after which the run was repeated. This latter test should then represent the internal friction of a partially aged specimen. Data of this type for aluminum alloys containing 0.08 percent Cu and 1.00 percent Cu are plotted in figure 9. For the 0.08 percent Cu alloy, the damping decreased over the entire temperature range investigated, while, for the 1.00 percent Cu alloy, the internal friction increased below 220° C and decreased above this temperature. It appears that the aging decreased the magnitude of the grain-boundary peak. Isothermal-aging curves of similar alloys are shown in figures 10(a) to 10(c).

The 0.08 percent Cu specimens aged at 195° and 157° C (fig. 10) both show a plateau or incubation period lasting about 10 minutes and then a decrease in the internal friction which is continuing after 1,300 minutes. Other than increasing the magnitude of each curve, the 2-percent tensile-strain has had a relatively small effect on the rate at which the decrease in internal friction occurs. The incubation period is decreased in both cases, and the initial, more rapid, dropoff of internal friction seems to be hastened by a factor of about 2. Between the two aging temperatures, there seems to be again only a small difference other than magnitude. It is difficult to evaluate this difference when the final value of the internal friction is unknown. However, it does not seem probable that the difference in aging rates exceeds a factor of 3 and, for an activation energy of about 25,000 calories per mole, this factor would have to be about 26.

Aging the specimens containing 1.00 percent Cu at or below 200° C presents, in part, a different picture, as might be inferred from the data in figure 9. The data (fig. 10(b)) do not present a consistent picture, although they have some similarities. Both curves show a slight decrease at the beginning of aging. A 2-percent tensile strain magnifies this drop and indicates that the drop may result from the recovery of either applied or handling strains. There appears to be an increase in the internal friction in both cases near the end of the aging test of both strained specimens. This might be expected if some peak is being built up. At 250° C (fig. 10(c)) the internal-friction aging curve looks much like that for the specimen containing 0.08 percent Cu at 195° C. This is a possible indication that the same process is occurring in both cases, but the rate is slower (requiring a higher temperature) for the aging process in the richer alloy. This is an indication, assuming that the aging process is one of diffusion, of concentration-dependent diffusion.

A specimen of the 4.82 percent Cu alloy was heat-treated for 1 hour at 542° C and water quenched. It was then mounted in the internal-friction apparatus and readings were taken while the wire was being heated at approximately 4° C per minute as with the previous specimens. This resulted in small peaks at about 115° and 185° C (fig. 11(a)). After heating to 260° C, the specimen was cooled in the apparatus and the same test was rerun. This run showed only a single peak at 180° C. After an anneal of 18 hours at 250° C, a similar test showed that the entire curve was raised. It appears that the peak at 180° C is still present, but the higher background is beginning to mask its presence. In a specimen of the alloy containing 5.45 percent Mg which was heat-treated for 1 hour at 500° C, water quenched, and tested in a similar manner, a parallel reaction seemed to occur (fig. 11(b)). The quenched specimen showed a peak at about 110° C, while, after aging, a different peak occurred at about 165° C.

It is possible that the lower temperature peaks in both these alloys are due to the stress-induced diffusion of copper atoms in solid solution. Annealing permits the atoms to move to dislocation sites in which a similar stress-induced diffusion results but with an increased relaxation time, resulting in a peak at a higher temperature.

Childs and LeClaire (ref. 20) suggest that a reasonable estimate for the diffusion rate D is given by

$$D = a^2/12\tau \qquad (2)$$

where a is the lattice parameter and τ is $\frac{1}{\omega}$ or $\frac{1}{2\pi f}$ at the temperature of the internal-friction maximum.

In the case of the alloy containing 4.82 percent Cu, when the lower
peak occurs at about 115° C at a frequency of 1.5 cycles per second, D
must be about 10^{-15} cm²/sec. The same calculation for the peak at 180° C
gives a value of D of about the same magnitude if the lattice parameter
is considered to be the same. It must be remembered that this is a gross
approximation in that the validity of equation (2) is being assumed at
low concentrations, and the results assume the correct interpretation of
the internal-friction peak. However, if these assumptions are moderately
accurate, and if the foreign atoms generally diffuse in the stress field
of a dislocation, then the calculated activation energies are considerably
lower and the diffusion rates obtained from extrapolation to lower tem-
peratures are considerably higher than might be anticipated from standard
diffusion data (see fig. 12). These data would suggest activation ener-
gies of from 20,000 to 29,000 calories per mole and room-temperature
diffusion rates of from 10^{-18} to 10^{-21} cm²/sec for the diffusion of
copper and magnesium in aluminum. Most reported activation energies
for the diffusion of copper and magnesium in aluminum, with the excep-
tion of the value given by Beerwald (ref. 21), are in the range from
32,600 to 38,500 calories per mole (ref. 22). Beerwald's value for the
diffusion of magnesium in aluminum is 28,600 calories per mole. A
Dushman-Langmuir calculation of the activation energies from conventional
diffusion data gives 31,400 for magnesium in aluminum and 29,000 for
copper in aluminum (ref. 23). Relating the temperatures at which the
peaks occur to their activation energies by means of the relation pub-
lished by Wert (ref. 24) gives about 25,000 calories per mole for the
115° C peak and about 29,000 calories per mole for the 180° C peak in
the alloy containing 4.82 percent Cu. The measured activation energy
for the 165° C peak in Al-Mg alloy is 27,800 calories per mole. The
data suggest that the published activation energies are high, but a
clarification of the data must await a clarification of the mechanism
of these damping peaks. The data are suggestive, however, of a concen-
tration dependence of the diffusion rate. The diffusion rate appears
to be faster in a random solid solution than in the regions of high
concentration around a dislocation.

"Anomalous" Internal-Friction Peak

Kê (ref. 14) observed what he termed an "anomalous" peak in an
alloy of aluminum and 0.5 percent Cu. This peak was obtained by par-
tially recrystallizing the cold-worked alloy at 300° C. The peak
displayed a rather uncommon dependence on the strain amplitude of
testing, decreasing in magnitude as the strain amplitude increased.
Kê also observed that, at room temperature, this peak first increased
and then decreased in magnitude, disappearing in several days. After
its disappearance, the peak could be reestablished by subsequent
annealing at 300° C. An anneal at 400° C completed the recrystallization

and removed the peak completely. Many attempts were made during this investigation to repeat this experiment using the 0.08 and 1.00 weight percent Cu alloys, but no peak was observed. A somewhat different procedure, however, produced a somewhat similar peak. After a 1-hour anneal at 500° C in the internal-friction apparatus, a specimen of the 1.00 percent Cu alloy was cooled down slowly in the apparatus, internal-friction readings being taken as the temperature decreased. A peak of the type K$\hat{e}$ reported was observed at about 125° C.

The amplitude dependence of this small peak was quite pronounced and went through what appears to be a considerable change (fig. 13). At room temperature there was virtually no amplitude dependence. As the temperature rose, the amplitude dependence became more and more pronounced, going through a maximum and then decreasing with increasing strain amplitude. At around 200° C, however, this decrease disappeared and the internal friction appeared to increase as a linear function of strain amplitude.

This amplitude dependence can be explained by a mechanism consistent with that postulated by K$\hat{e}$. If a dislocation is dragging along a trapped copper atom during its oscillation, it follows that the farther the atom is dragged, the greater is the energy dissipation. However, if the stress is high enough, the temperature low enough, or the strain rate fast enough, then the dislocation can break free of its anchor. In the torsional pendulum, both the maximum stress and the maximum strain rate are raised at the same time. Therefore, a condition is approached with increasing amplitude when the dislocation breaks free of its anchoring atom before the atom has been dragged very far. Hence, the low internal friction at low temperature means little dragging, the reversion at moderate temperatures indicates a dragging and a breaking free, while the lineal high-temperature amplitude dependence indicates complete dragging along of the copper atoms.

In view of the rapidity with which, as K$\hat{e}$ reports, this peak disappears upon aging at room temperature, it seems odd that the slow cooling from 300° C would have not long since completed the aging. It seems odd, also, that, if these are copper atoms at dislocations, the 180° C peak in the 4.82 percent Cu alloy does not exhibit similar behavior, for it is believed to arise from atoms at the dislocations. It is possible that this peak is the result of the dragging of small groups of copper atoms rather than single atoms, but it seems rather unlikely that such groups could migrate at such low temperatures.

A further oddity is that, while K$\hat{e}$'s peak occurred at about 25° C, the peak in the 1.0 percent Cu alloy occurred at about 125° C. This is strongly indicative of a concentration-dependent peak temperature if the mechanism of energy dissipation is the same in each case. If the relaxation mechanism is controlled by diffusion, this might also be interpreted as evidence of concentration-dependent diffusion rates, with the rate decreasing with increasing concentration.

It was hoped that aging solution-treated and quenched specimens at various temperatures would produce the anomalous peak. This was to be determined by following the amplitude dependence of the internal friction as aging proceeded. However, in no case was the amplitude dependence, after aging, significantly different from that early in the aging, nor was it nearly so pronounced as that shown by the anomalous peak.

If the previous discussion on diffusion is essentially correct, it is possible to make use of an expression derived by Cottrell (ref. 5, p. 149) to determine the time t for a certain fraction f of the original solute to segregate to dislocations. This expression is

$$f = \alpha\rho(ADt/KT)^{2/3} \tag{3}$$

where

α constant, $3\left(\frac{\pi}{3}\right)^{1/3}$

ρ dislocation density

A constant, approximately 5×10^{-21} dyne cm^2 (ref. 25)

K Boltzmann constant

D diffusion rate

T absolute temperature

From this expression, it is possible to make some rough estimates of the time it would take for half the solute to segregate to dislocations ($f = 0.5$). These data are given in table 4.

It is clear from these values that, if the assumed diffusion rates are at all accurate, and if the dislocation densities fall below 10^{12} per square centimeter, then very little segregation will be possible at room temperatures. The aging reported by Kê (ref. 14) and by Sherby, Anderson, and Dorn (ref. 10) at room temperature in all probability does not involve segregation of atoms to dislocations. These calculated times are, however, of the right order of magnitude to explain the disappearance of the low-temperature peaks in 4.82 percent Cu and 5.45 percent Mg alloys after annealing at temperatures at or below 250° C.

These calculations suggest a strong temperature-dependent aging, since the rate of segregation is controlled by the strongly temperature dependent diffusion rate. The aging curves of the previous section do

not show this temperature dependence and, hence, are probably not indic-
ative of the rate of segregation of impurity atoms to dislocations.
This segregation may be affecting the internal friction but to a degree
which has not yet permitted the effects to be resolved from other con-
current effects.

<center>Variation of Internal Friction During</center>

<center>Tensile Deformation</center>

Wires of the various alloys (excluding the 4.82 percent Cu and
5.45 percent Mg alloys) were heat-treated as shown in table 3 and were
pulled in tension at constant strain rates in the internal-friction
machine previously described. The internal friction was measured by
superimposing a small-amplitude torsional oscillation at about 2.5 cycles
per second on the specimen and observing the decay of the amplitude of
the oscillation. In general, it was found that the internal friction
rose rapidly upon the initiation of plastic deformation but decreased
continuously after passing through a maximum. This decrease appears to
be a function of strain rate and alloy content. Figures 14(a) and 14(b)
show the variation of internal friction with strain for four different
strain rates for high-purity wire and for an aluminum alloy containing
1.00 percent Cu and indicate the effectiveness of copper in depressing
the damping. For the most part, the dependence of the internal friction
on strain rate appeared to be linear. In the high-purity wire, however,
there appeared to be a slight curve to the data. These data are summa-
rized for all the alloys tested in figure 15 and again show the ability
of foreign additions in depressing the internal friction. As an approx-
imation, it appears that the logarithm of the internal friction is pro-
portional to the square root of the alloy content. The meaning of this
is unknown, and the verification of this point must await the testing
of intermediate alloys. Aging the 0.08 percent Cu and the 1.00 percent Cu
alloys for 17 hours at 300° C caused the internal friction to drop an
additional amount, varying from about 10 to 30 percent (fig. 16).

It was observed that the temperature dependence of the internal
friction of high-purity aluminum was virtually zero in the range from
25° to 150° C. The 0.08 percent Cu alloys showed an appreciable decrease
in internal friction with increasing temperature (fig. 17). The 0.09 per-
cent Mg alloys showed what appeared to be a decrease, but it was not
nearly so pronounced. The 1.00 percent Cu alloys (fig. 18), however,
showed just the opposite temperature dependence, with the internal fric-
tion increasing with increasing temperature. The 0.78 percent Mg alloy
showed this same increasing internal friction with increasing temperature.
It appears from this that the temperature dependence is not strictly a
property of the pure metal but is more concerned with the solute content.

The observed, damping or energy loss may come about from three causes. These are:

(1) The oscillation of large numbers of dislocations present in the dynamic conditions of straining

(2) The interaction of anchoring particles with the oscillating dislocations

(3) The release of some of the elastic energy stored in torsion by the plastic deformation initiated primarily by the tensile load

It is believed that the oscillation of dislocations can dissipate considerable energy. At the low stresses and frequencies of the present tests, however, it is doubtful that there can be any lag of the dislocation motion behind that of the stress unless small amounts of plastic deformation occur and absorb energy. It is interesting to note, however, that the internal friction of cold-drawn, high-purity aluminum wire at room temperature is more than 20 times that of the same wire in a recrystallized state. This would indicate a large amount of reversible plastic deformation on a local scale, perhaps similar to the microstrain recently discussed by Clark (ref. 25) where yielding was believed to occur within a grain but without sufficient energy to pass a grain boundary. This internal friction is virtually independent of strain amplitude which indicates that it does not result from an interaction mechanism with impurity particles.

The interaction forces between dislocations and anchoring particles are temperature dependent and must decrease with increasing temperature. This temperature dependence apparently leads to the formation of the "anomalous" internal friction peak referred to previously. Indications are that the temperature at which this peak occurs decreases with decreasing solute content. The same temperature dependence would apply for an interaction mechanism observed during tensile straining and might account for the reversed temperature dependence of the internal friction during the tensile straining of the 0.1- and 1.0-percent-copper alloys. It is probable that the internal friction observed during the tensile straining of these alloys goes through a maximum at some temperature. The temperature dependence of the internal friction will therefore depend upon the temperature range and the alloy content being studied.

The rapid drop in internal friction with increasing alloy content is a reflection of the increased rate of strain-hardening brought about by the alloy additions and the consequent increasing inability of the dislocations to produce even local deformation. That the stress required to maintain a constant strain rate is closely connected with the internal friction is shown by the fact that, in some cases, a plot of stress versus

internal friction is a straight line. This relation appears to fall off at very low strain rates, probably because of the change in the static background damping which cannot at present be estimated and subtracted away. A change in alloy content changes the slope of this line.

The third source of energy loss which is anticipated is that due to the plastic deformation itself. When a small stress is applied in torsion, elastic energy is stored in the specimen. If, however, a second large stress causes plastic deformation, the torsional stress will contribute to this deformation and some of the stored energy will thereby be lost. This energy loss will be, to a first approximation, proportional to the time during which the torsional stress is applied or to the period of oscillation. This expectation was checked and found to hold true as shown in figure 19. The internal friction, extrapolated to zero, gives the damping independently of this "mechanical" loss. The effects of other variables such as strain rate, alloy content, or temperature on this loss have not been investigated.

DISCUSSION

The decrease in the shear modulus in the 0.08 percent Cu alloy above 200° C appears to be real, although the effect is small and the reason for such weakening is unclear. There is evidence in the literature of what might readily be interpreted as a mechanical effect of this weakening in other alloy systems (refs. 10 and 19). There is contrary evidence, both theoretical and experimental (ref. 26), that an increase in the shear modulus should be noted in very small additions to a high-purity metal. There is also reason to believe that the relaxation effects observed in the curves of shear modulus versus temperature are not entirely the result of grain-boundary slip but might involve the rearrangement of subboundaries or of the internal-dislocation network. The action of the small additions may not be to weaken an existing structure but to hold in place many more dislocations which would ordinarily have annealed out. In other words, small additions might preserve a weaker structure. It may well be that, with a change in heat treatment to produce a different grain size or after a different aging treatment, this decrease will disappear.

The decrease in the grain-boundary peak magnitude, both with the addition of impurity atoms and with an isothermal aging after quenching from a solid-solution anneal, indicates that the impurity atoms are segregating at grain boundaries and subboundaries. The lack of a strong temperature dependence or of an appreciable effect of prestrain indicates that the decrease in internal friction with isothermal aging is not a

function of ordinary lattice diffusion but is a result of the channeling
of foreign atoms along dislocations to preferential sites. The actual
mechanism by which the internal friction decreases appears to be by the
reduction of the rate of grain-boundary relaxation. The subsequent rise
in the internal friction after some annealing (for example, in the
5.45 percent Mg alloy, fig. 11(a)) is apparently due to an energy loss
occurring in some precipitate particles. The actual change of internal
friction due to the segregation of impurity atoms to dislocations may
be too closely tied up with the decrease in the magnitude of the grain-
boundary peak to be resolved. In the richer alloys, however, the decay
of the small, low-temperature peaks may be a result of this segregation.

The peak observed in the 0.5 percent Cu alloy by Kê (ref. 14) and
in the 1.0 percent Cu alloy in this investigation has not been satis-
factorily explained. It is not apparent why a slow-cooled specimen
should produce a peak which itself disappears with aging, while a quenched
specimen does not. The mechanical effects of the internal-friction
testing may have an effect in the production of this peak. The two peaks
observed may be due to two similar but different mechanisms or may indi-
cate a concentration-dependent diffusion rate.

The general characteristics of the behavior of internal friction
during plastic deformation agree with those observed by Alers (ref. 16)
in zinc single crystals. The results of the test of aluminum alloys
seem to indicate that at least some of the internal-friction phenomena
observable are due to a mechanism similar to that responsible for the
anomalous peak in Al-Cu alloys. Whether this is a dragging along of
single atoms or the interaction of a dislocation with a coherent precip-
itate particle is not at once apparent, although the latter seems to be
a more favorable mechanism to account for the observed behavior. The
decreased internal friction during deformation after an aging treatment,
when it is suspected that more small coherent particles are present,
would support this latter mechanism. Decreasing the size and number of
the particles would increase the misfit of the particle and increase the
work required for a dislocation to move through the particle. This would
increase the energy loss through interaction. A higher temperature peak
might then occur where the more developed particles have been allowed to
form. This might be particularly true for richer alloys where it is less
difficult for a particle to grow.

It is probable that this internal friction is also a reflection of
the same mechanism which leads to the plateau in the curves of flow
stress versus temperature of many alloys (ref. 10) and to the serration
in the stress-strain curves (Portevin-LeChatelier effect) of some of
these alloys.

CONCLUSIONS

An investigation of the internal-friction behavior of binary aluminum alloys has indicated the following conclusions:

(1) Small additions (0.08 weight percent) of copper to high-purity aluminum affect the grain boundaries of the material in such a way as to increase the rate of relaxation along these boundaries.

(2) The decrease of the internal friction during quench-aging of aluminum-copper alloys can be interpreted as the segregation of copper atoms to preferential sites such as grain boundaries or subboundaries.

(3) The two small peaks observed in both the 4.82 percent copper and the 5.45 percent magnesium alloys may be useful in determining the rate of segregation of solute atoms to dislocations and in the determination of solid-state diffusion rates.

(4) Calculated diffusion rates based upon internal-friction data and published data indicate that published observations of room-temperature aging involve either the motion of dislocations to impurity atoms or the movement of impurity atoms along dislocation lines rather than the random movement of impurity atoms to dislocations.

(5) The difference in the temperature at which the anomalous peak occurs indicates either a concentration-dependent type of anchoring particle or a concentration-dependent diffusion rate.

(6) The internal friction undergoes large changes during plastic deformation and appears to be affected by strain rate, temperature, degree of strain, alloy content, and prior heat treatment.

(7) The temperature-dependent internal friction during deformation and such mechanical-testing effects as the Portevin-LeChatelier effect are probably the result of the same mechanism, the interaction of dislocations with some impurity particle.

(8) Internal friction is a very sensitive tool for use in resolving the various effects of alloying, heat treatment, and temperature on the basic processes of plastic deformation.

Battele Memorial Institute,
    Columbus, Ohio, July 31, 1955.

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TABLE 1.- SPECTROGRAPHIC ANALYSIS

OF HIGH-PURITY ALUMINUM

Element	Weight, percent
Si	0.001
Fe	.005
Cu	.002
Mg	.001
Mn, ND[a]	<.001
Zr, ND[a]	<.01
Cr, ND[a]	<.001
Ni, ND[a]	<.002
Ti, ND[a]	<.0001
B, ND[a]	<.001
Pb, ND[a]	<.001
Ca	<.0005
Ga, ND[a]	<.005
Sn, ND[a]	<.005

[a]ND, not detected.

TABLE 2.- ANALYSIS OF EXPERIMENTAL ALLOYS

Alloy	Weight percent solute addition		
	Cu	Mg	Nominal
1	0.08	----	0.1
2	1.00	----	1.0
3	4.82	----	5.0
4	----	0.09	.1
5	----	.76	1.0
6	----	5.45	5.0

TABLE 3.- HEAT TREATMENT OF EXPERIMENTAL

ALLOYS TO PRODUCE A COMMON GRAIN SIZE[a]

Alloy	Time, min	Temperature, °C
High-purity Al	10	400
Al - 0.08 percent Cu	5	450
Al - 1.00 percent Cu	20	450
Al - 4.82 percent Cu	60	540
Al - 0.09 percent Mg	5	450
Al - 0.76 percent Mg	5	500
Al - 5.45 percent Mg	60	500

[a]All alloys were water cooled.

TABLE 4.- CALCULATED TIMES FOR SOLUTE TO

SEGREGATE TO DISLOCATIONS UNDER

DIFFERENT CONDITIONS

f	Dislocation density/cm^2	Temperature, °C	D, cm^2/sec (a)	t, sec (a)
0.5	10^8	27	10^{-21}	4×10^{14}
.5	10^{12}	27	10^{-21}	4×10^8
.5	10^8	120	10^{-17}	6×10^{10}
.5	10^{12}	120	10^{-17}	6×10^4
.5	10^8	250	10^{-13}	8×10^6
.5	10^{12}	250	10^{-13}	8

[a]Approximate value.
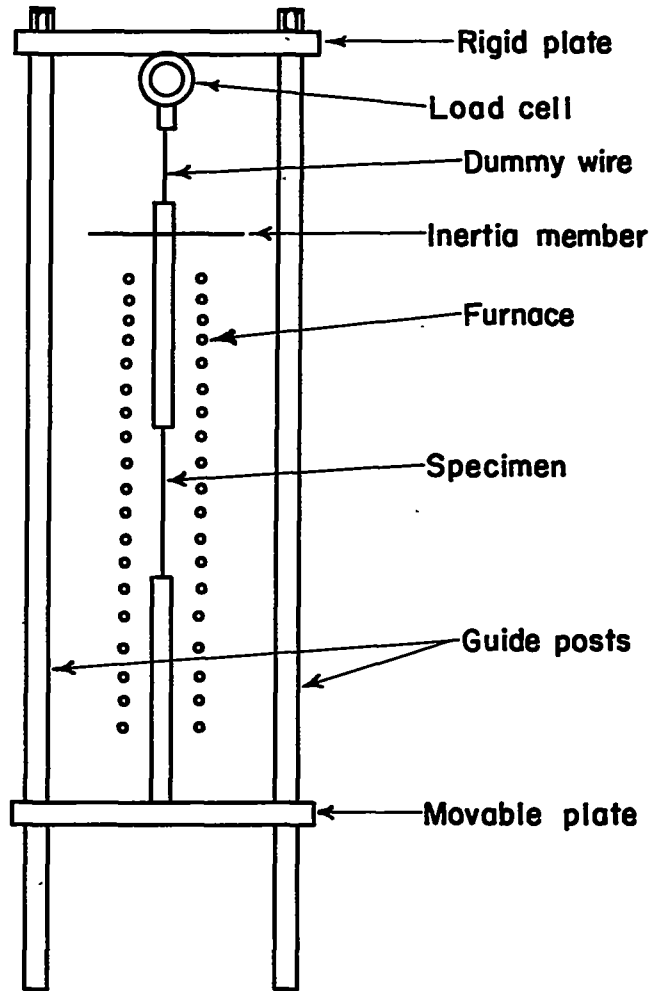
Figure 1.- Diagram of internal-friction testing apparatus.
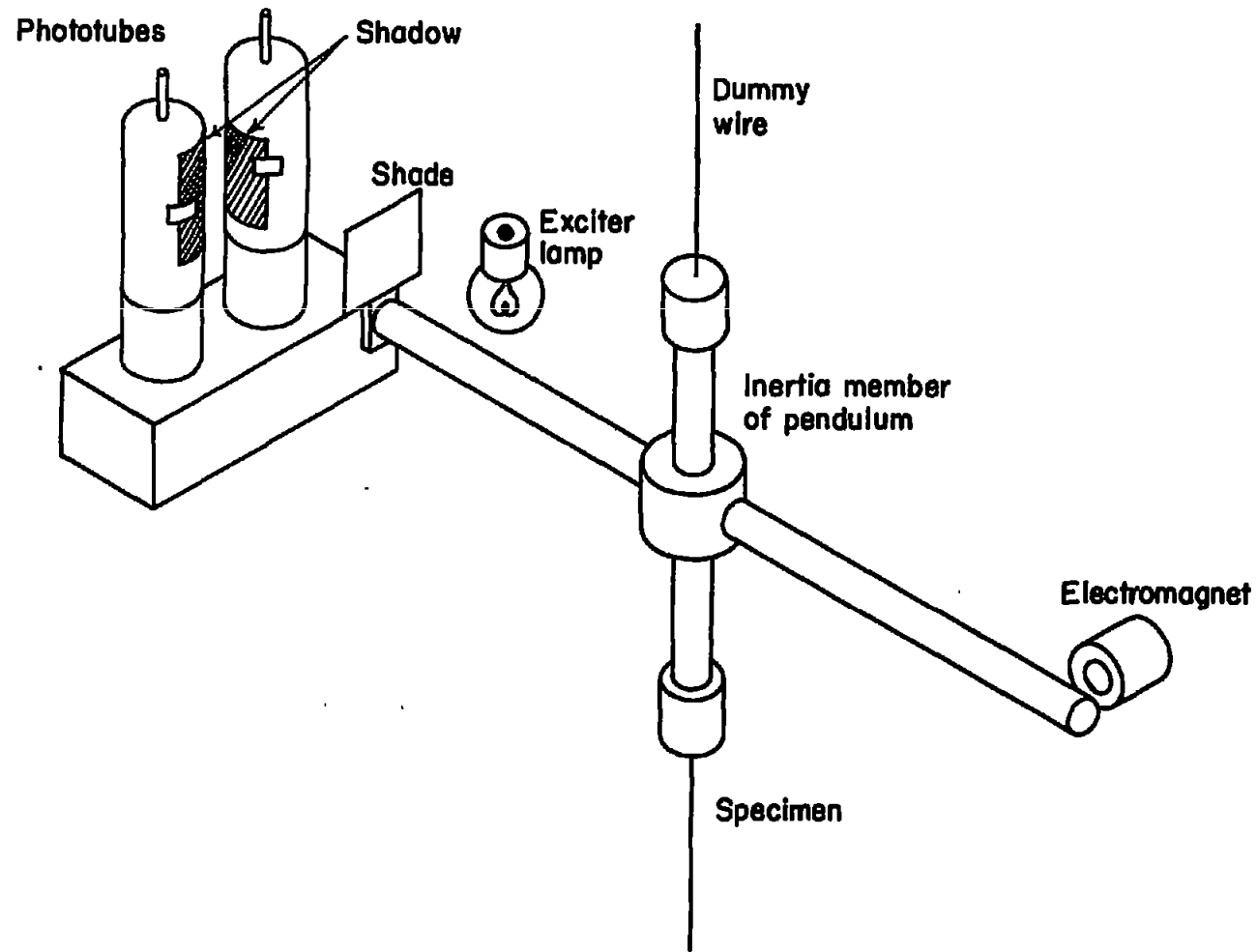
Figure 2.- Physical arrangement for sensing torsional vibrations.
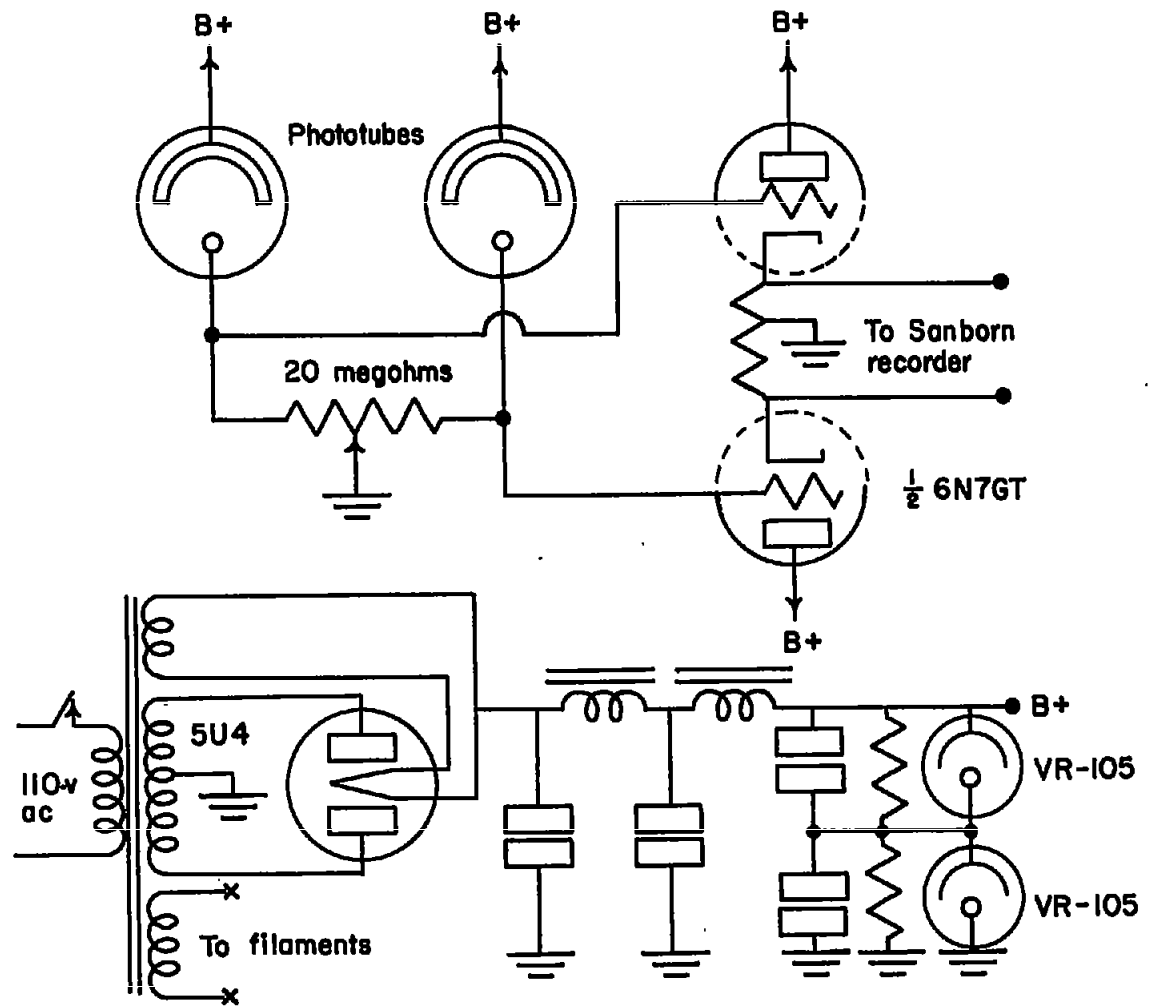
Figure 3.- Circuit diagram for oscillation-sensing device.

NACA TN 3681
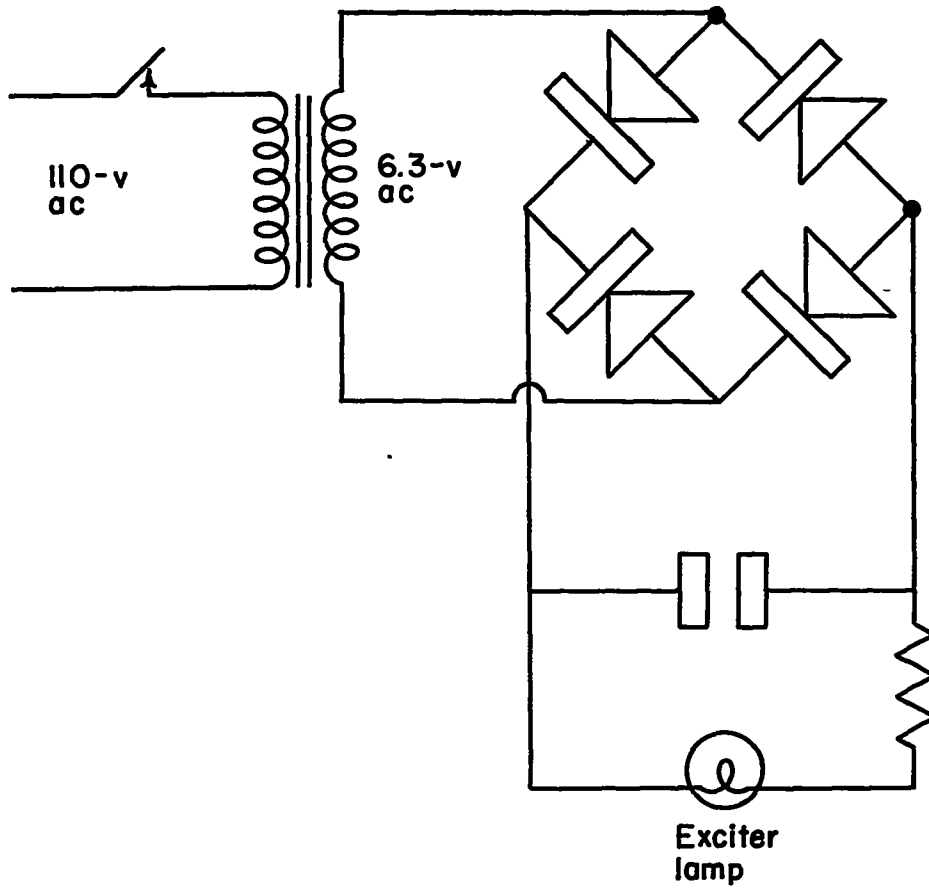
Figure 4.- Circuit diagram of phototube exciter-lamp circuit.
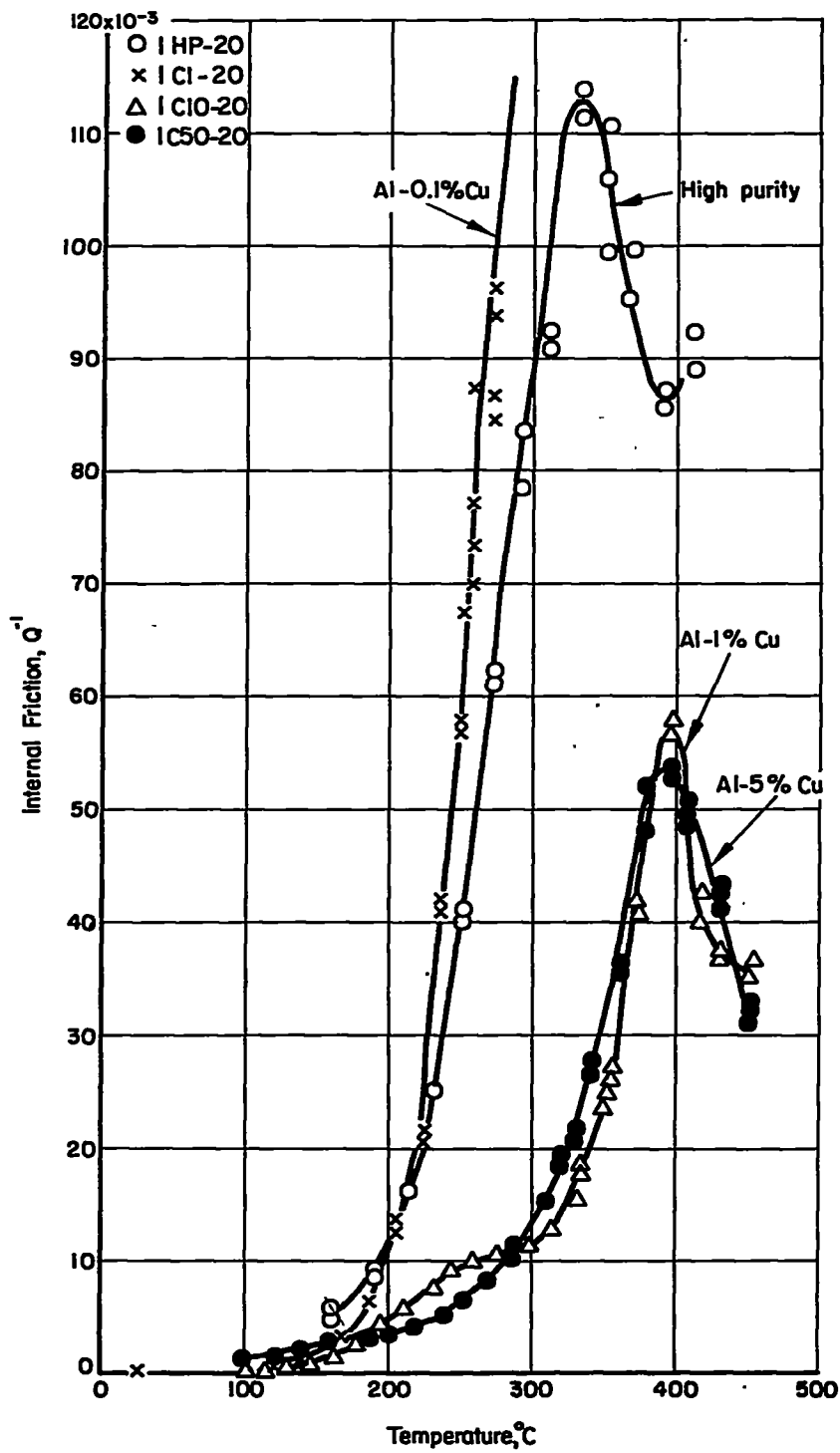
Figure 5.- Variation of internal friction with temperature for various
aluminum-copper alloys.
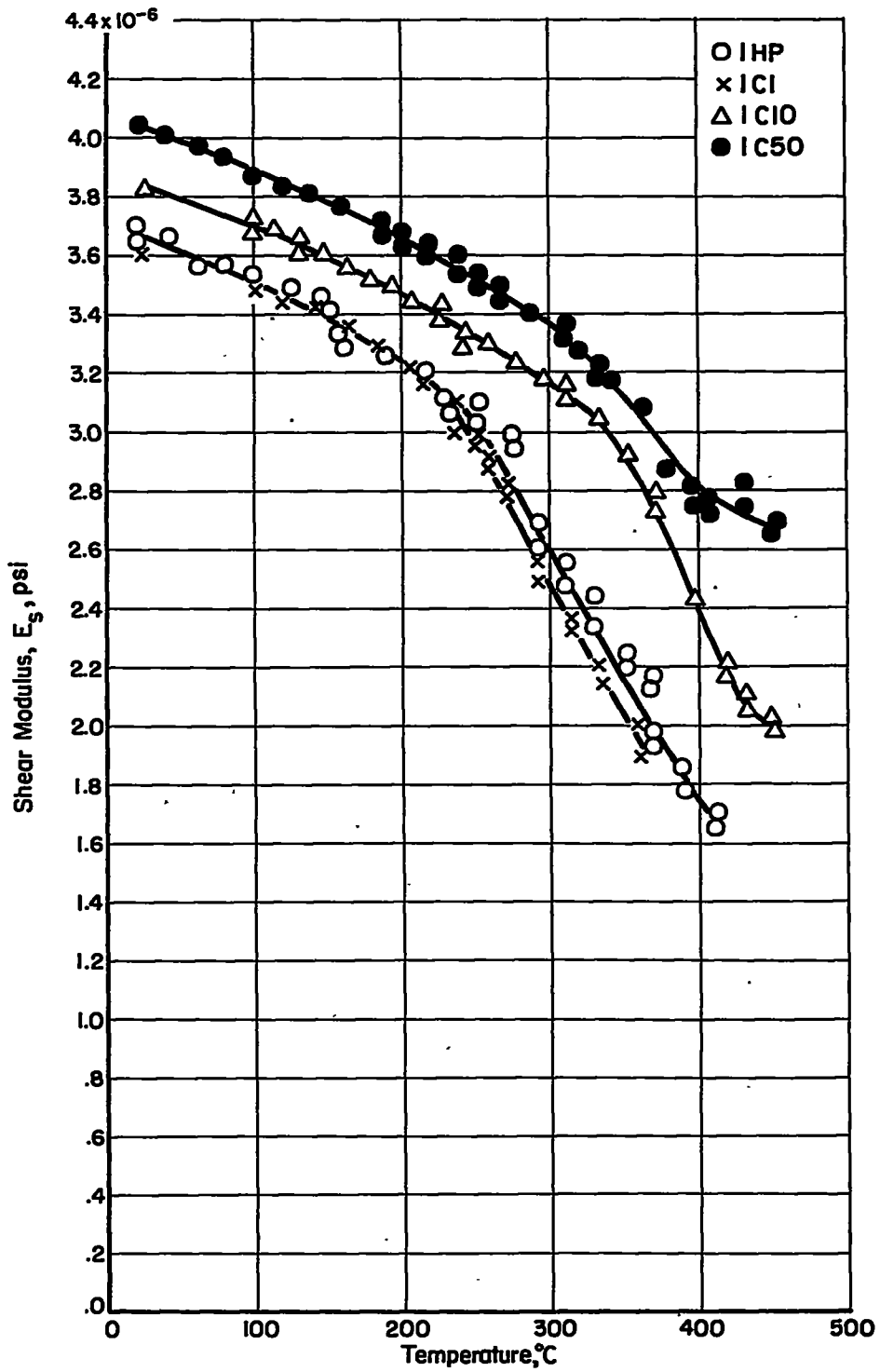
Figure 6.- Variation of shear modulus with temperature for various
aluminum-copper alloys.
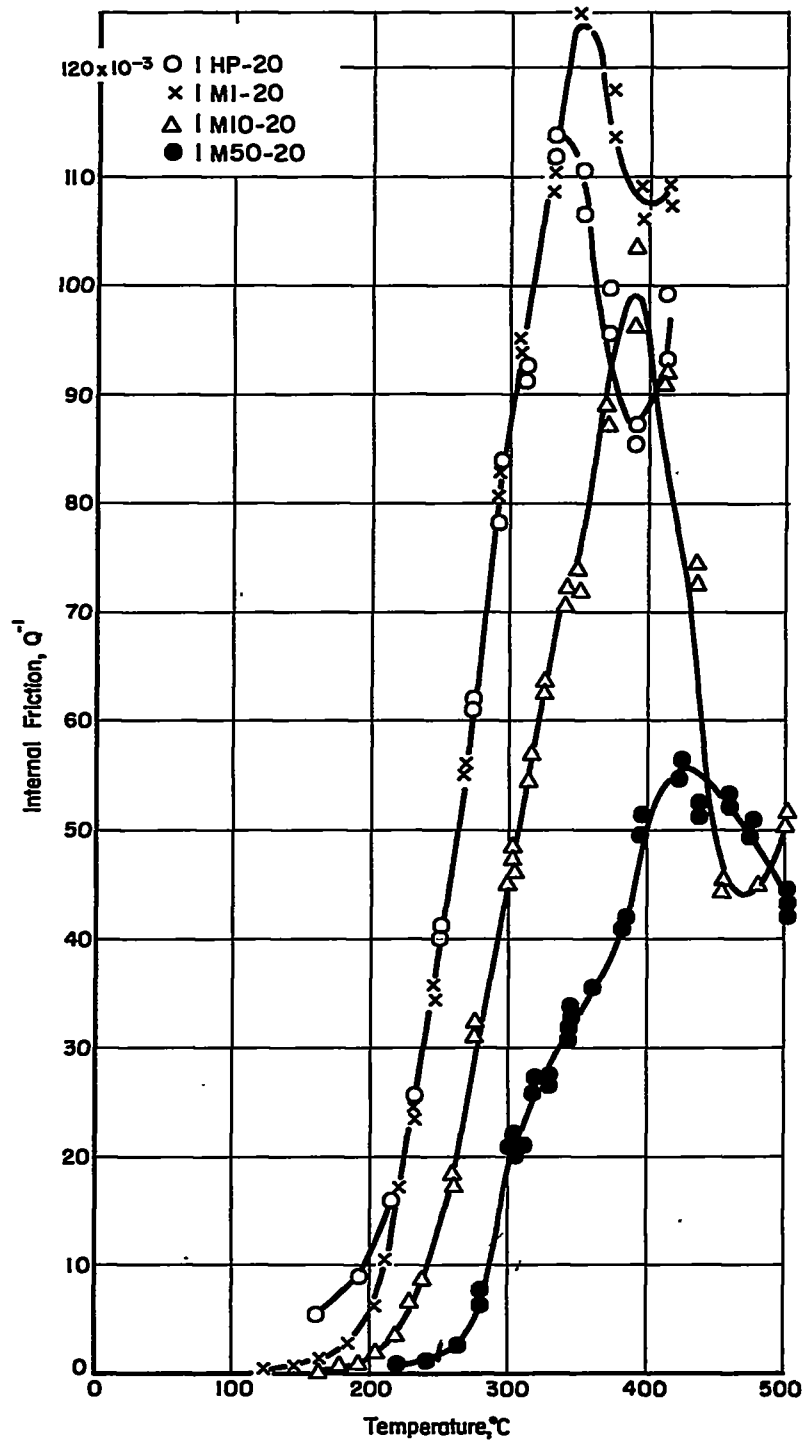
Figure 7.- Variation of internal friction with temperature for
aluminum-magnesium alloys.
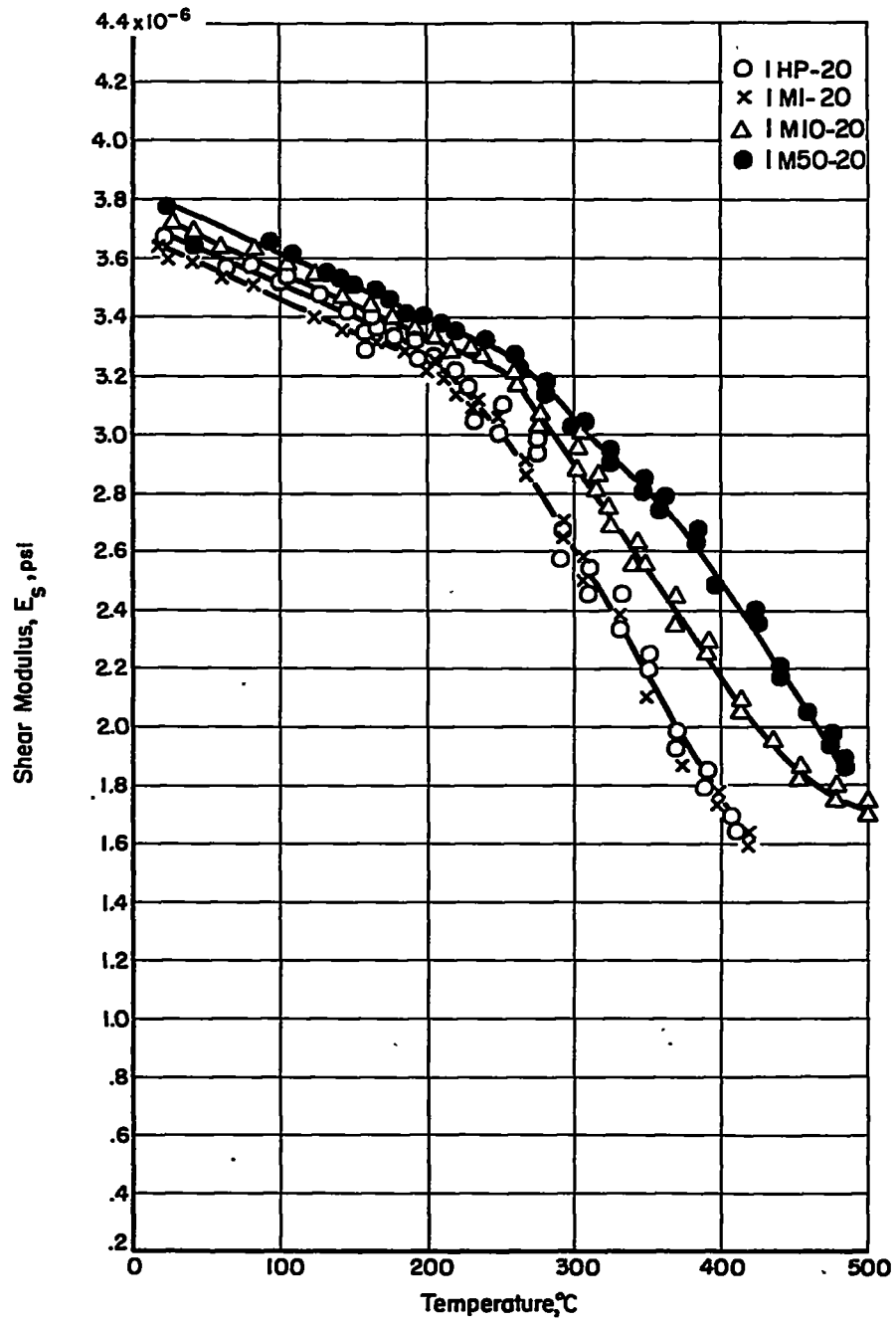
Figure 8.- Variation of shear modulus with temperature for various
aluminum-magnesium alloys.
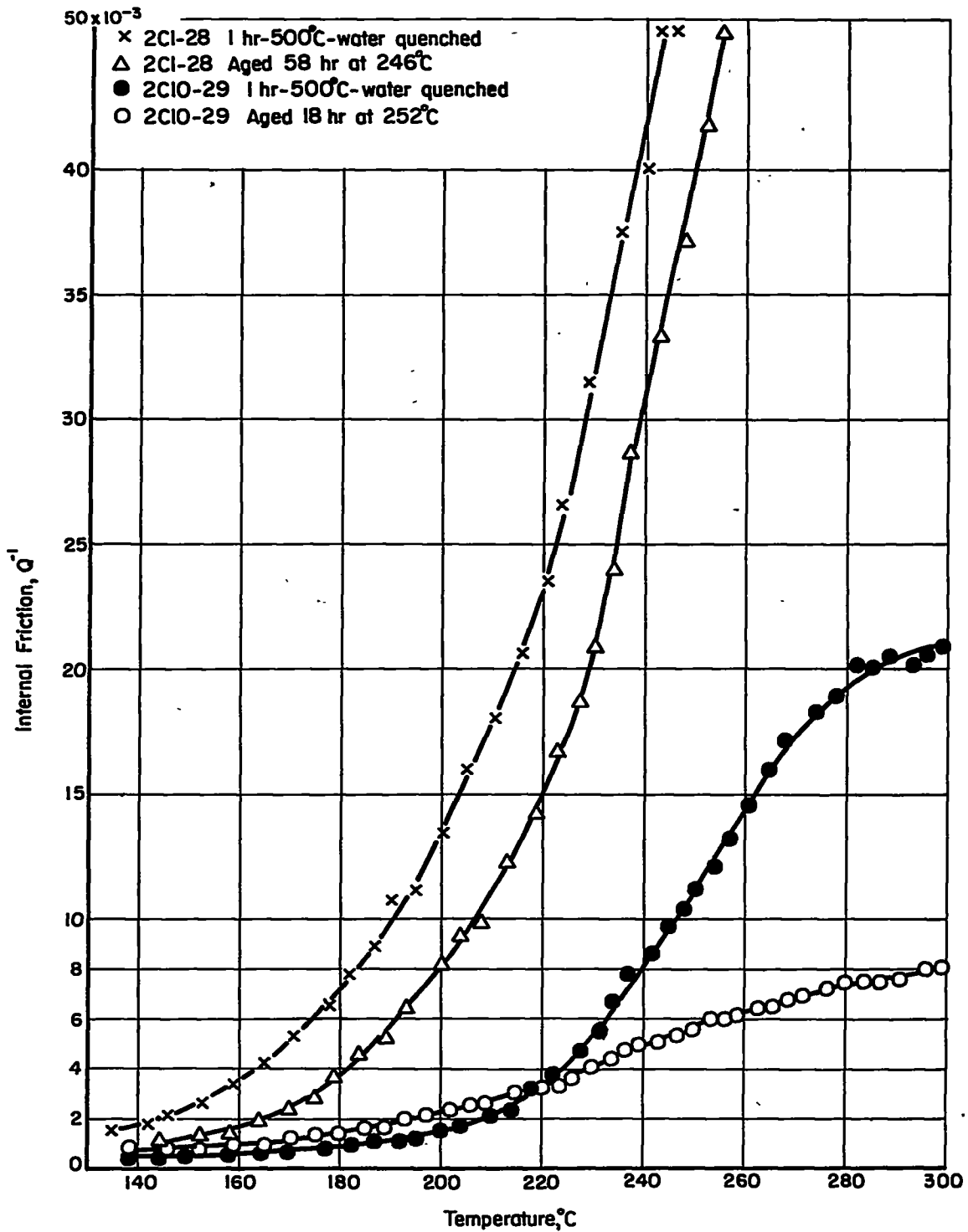
Figure 9.- Internal friction of unaged and aged aluminum-copper alloys.
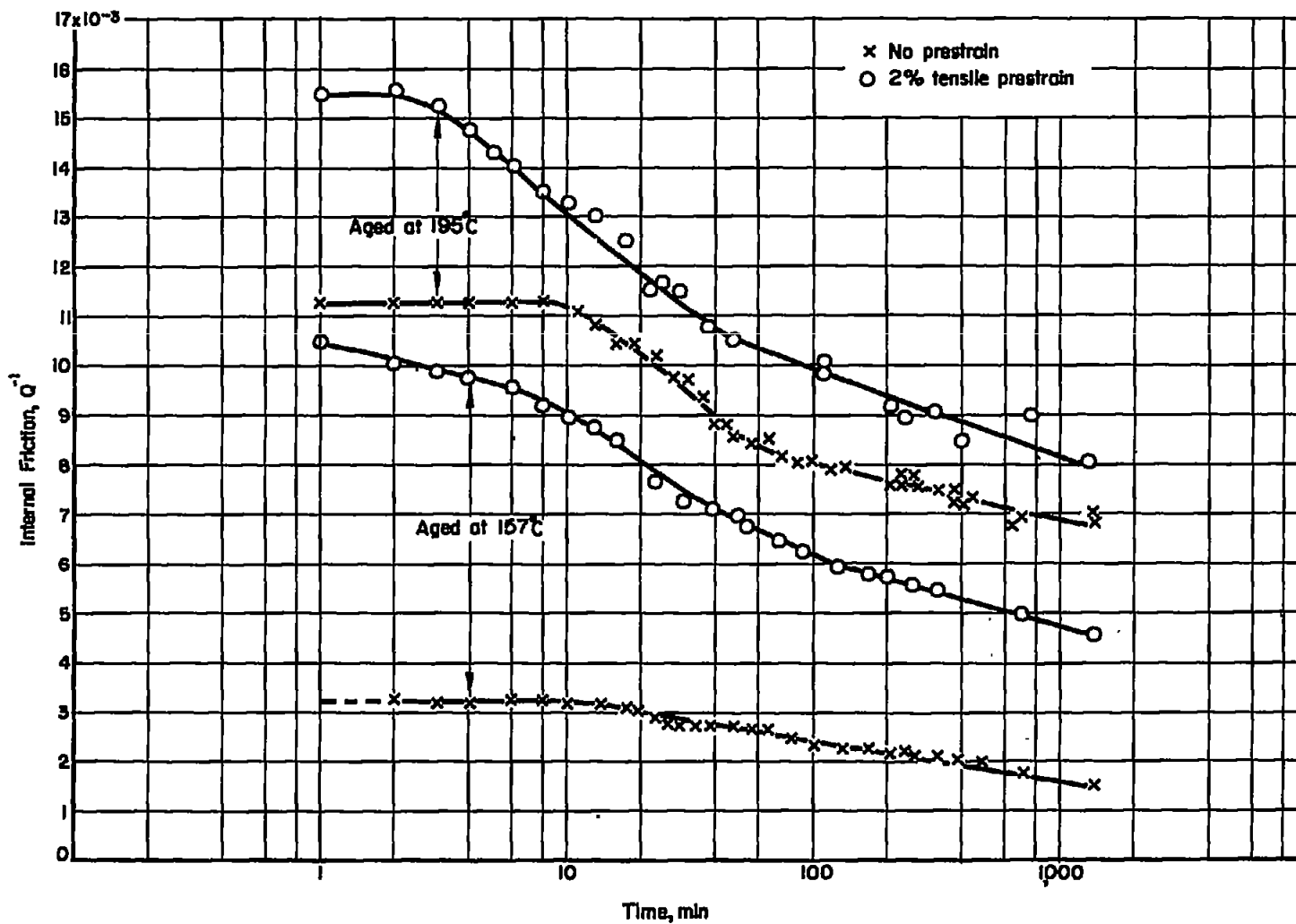
(a) Alloy containing 0.08 percent copper. Aged at 157° and 195° C.

Figure 10.- Variation of internal friction during aging of quenched aluminum-copper alloys.
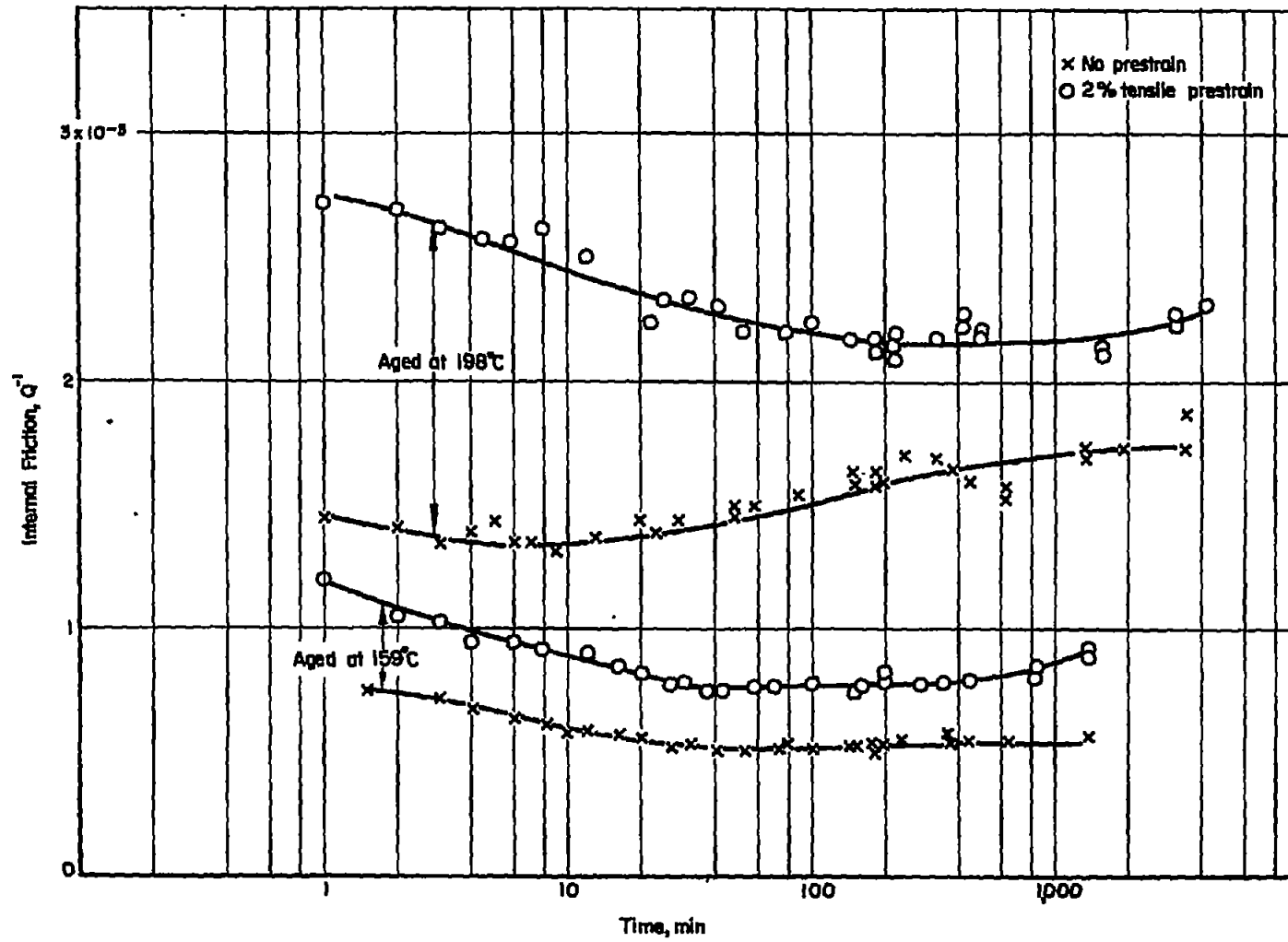
(b) Alloy containing 1.00 percent copper. Aged at 159° and 198° C.

Figure 10.- Continued.
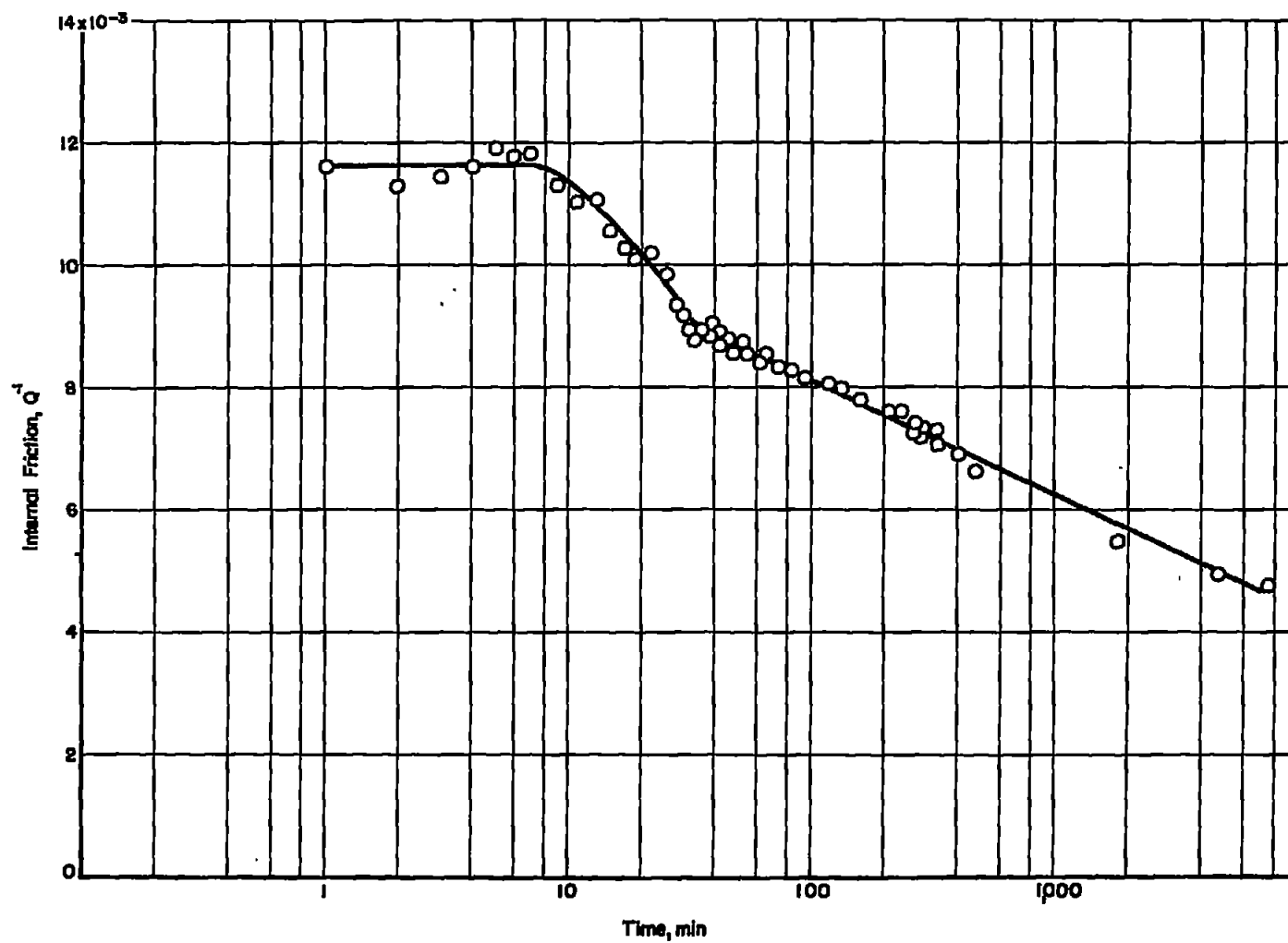
(c) Alloy containing 1.00 percent copper (specimen 2C10-31). Aged at 250° C.

Figure 10.- Concluded.
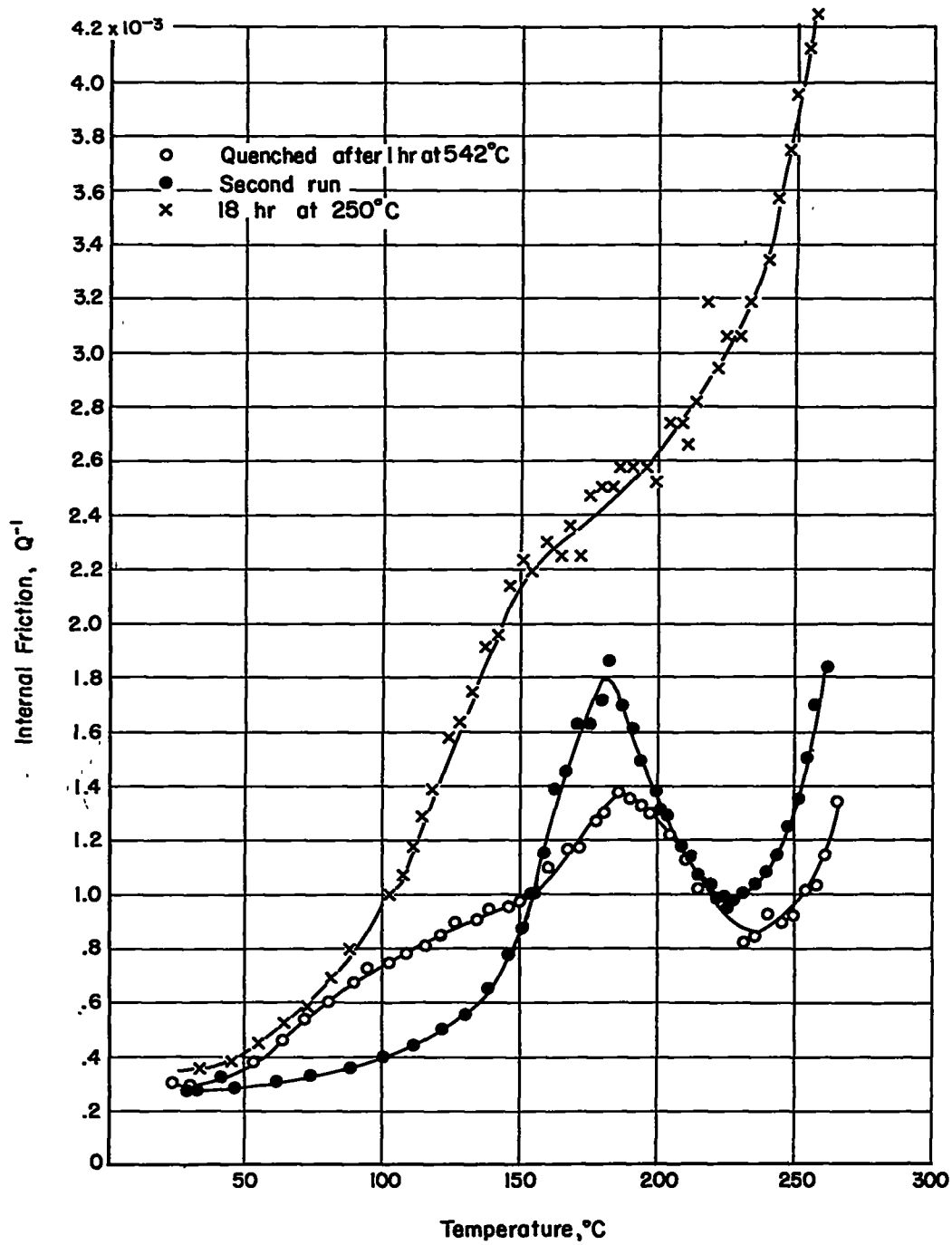
(a) Alloy containing 4.82 percent copper.

Figure 11.- Variation of curve of internal friction versus temperature of
an aluminum-copper alloy before and after aging.
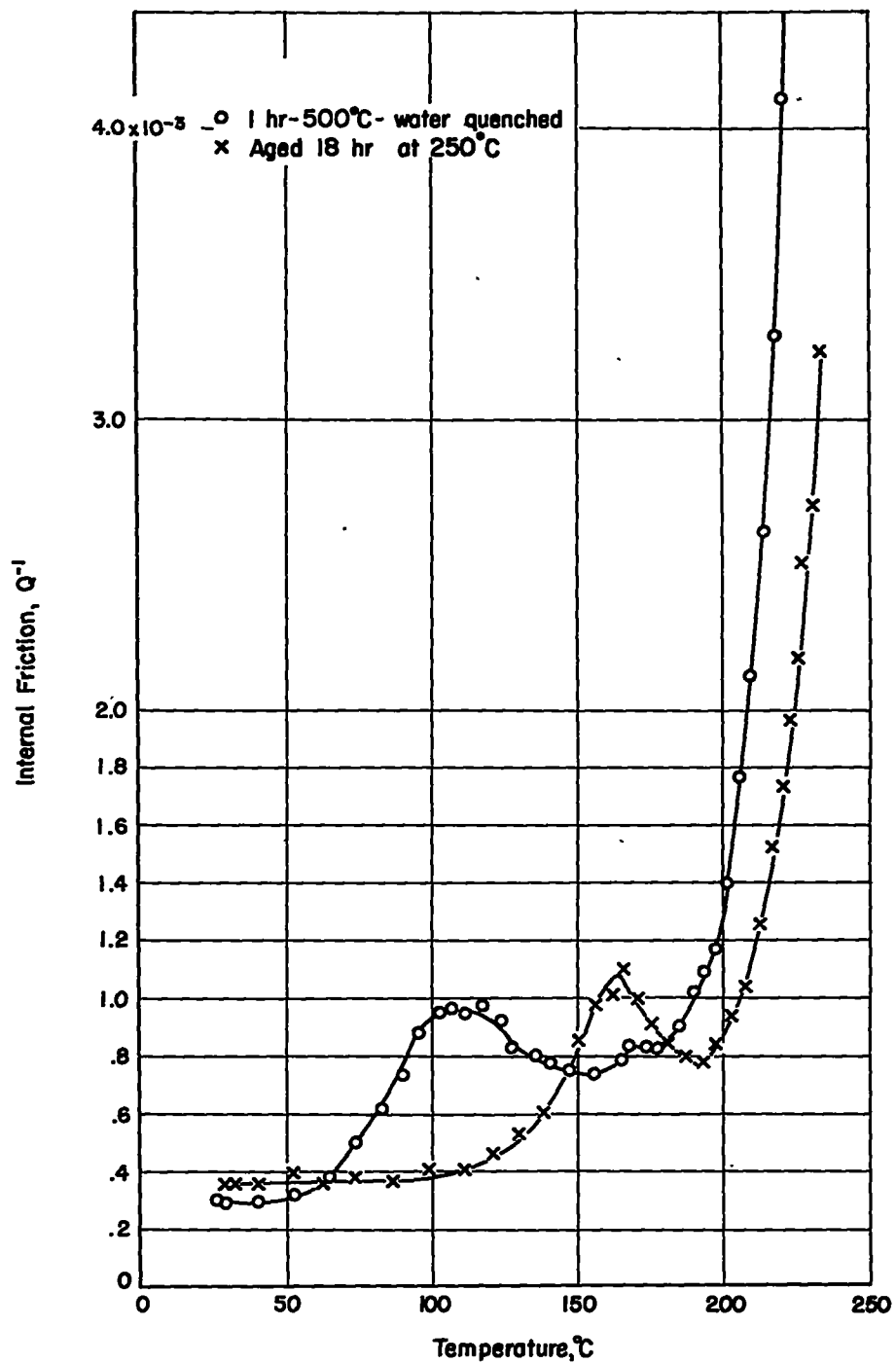
(b) Alloy containing 5.45 percent magnesium.

Figure 11.- Concluded.
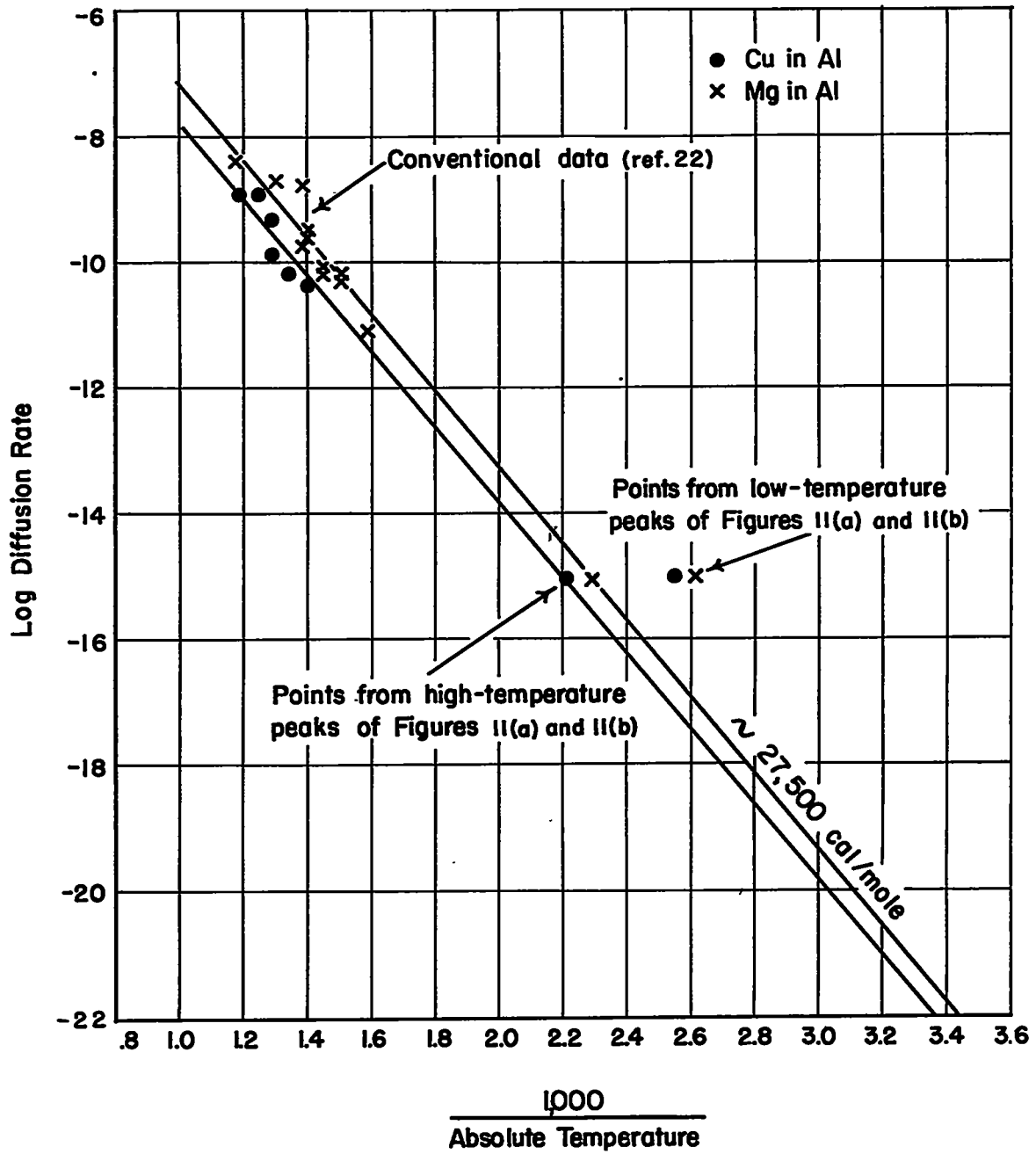
Figure 12.- Estimated extrapolation of diffusion data based on internal-friction points.
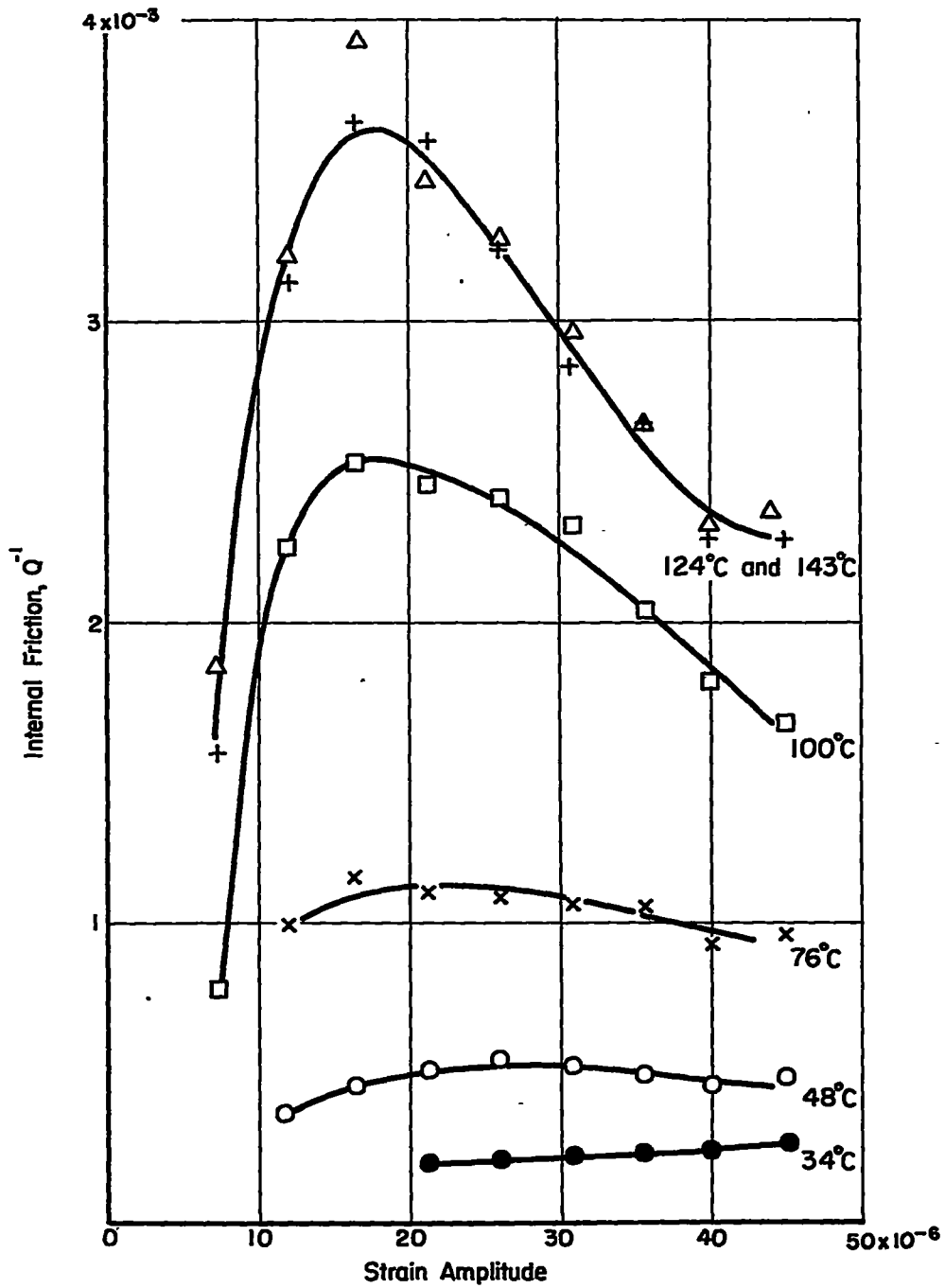
Figure 13.- Variation of dependence of internal friction on strain
amplitude with temperature for anomalous peak. Alloy containing
1.00 percent copper; furnace cooled after 1 hour at 500° C;
subsequent anneal, 1 hour at 260° C.
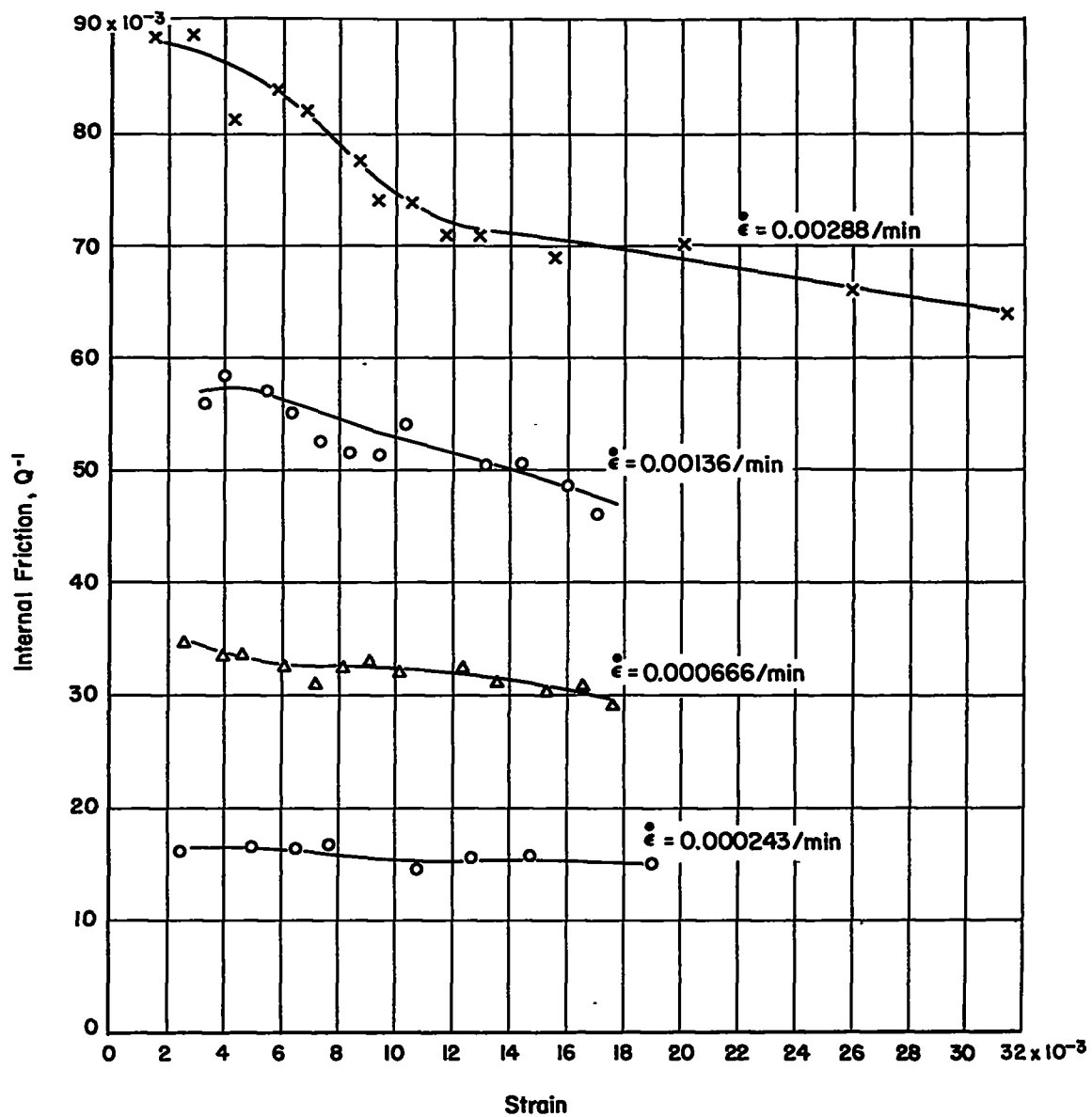
(a) High-purity aluminum.

Figure 14.- Variation of internal friction during plastic deformation for four strain rates $\dot{\epsilon}$.
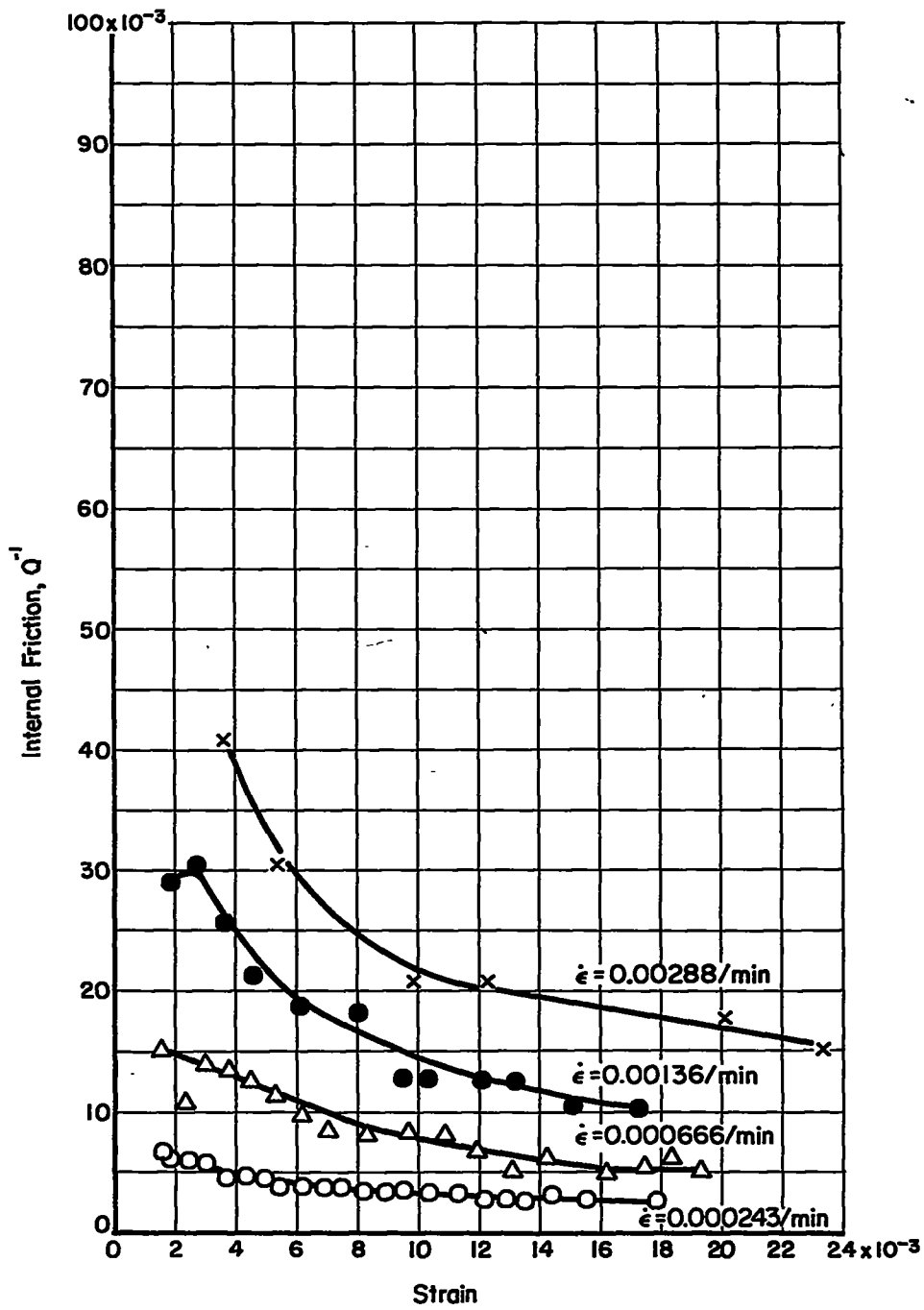
(b) Aluminum containing 1.00 percent copper.

Figure 14.- Concluded.
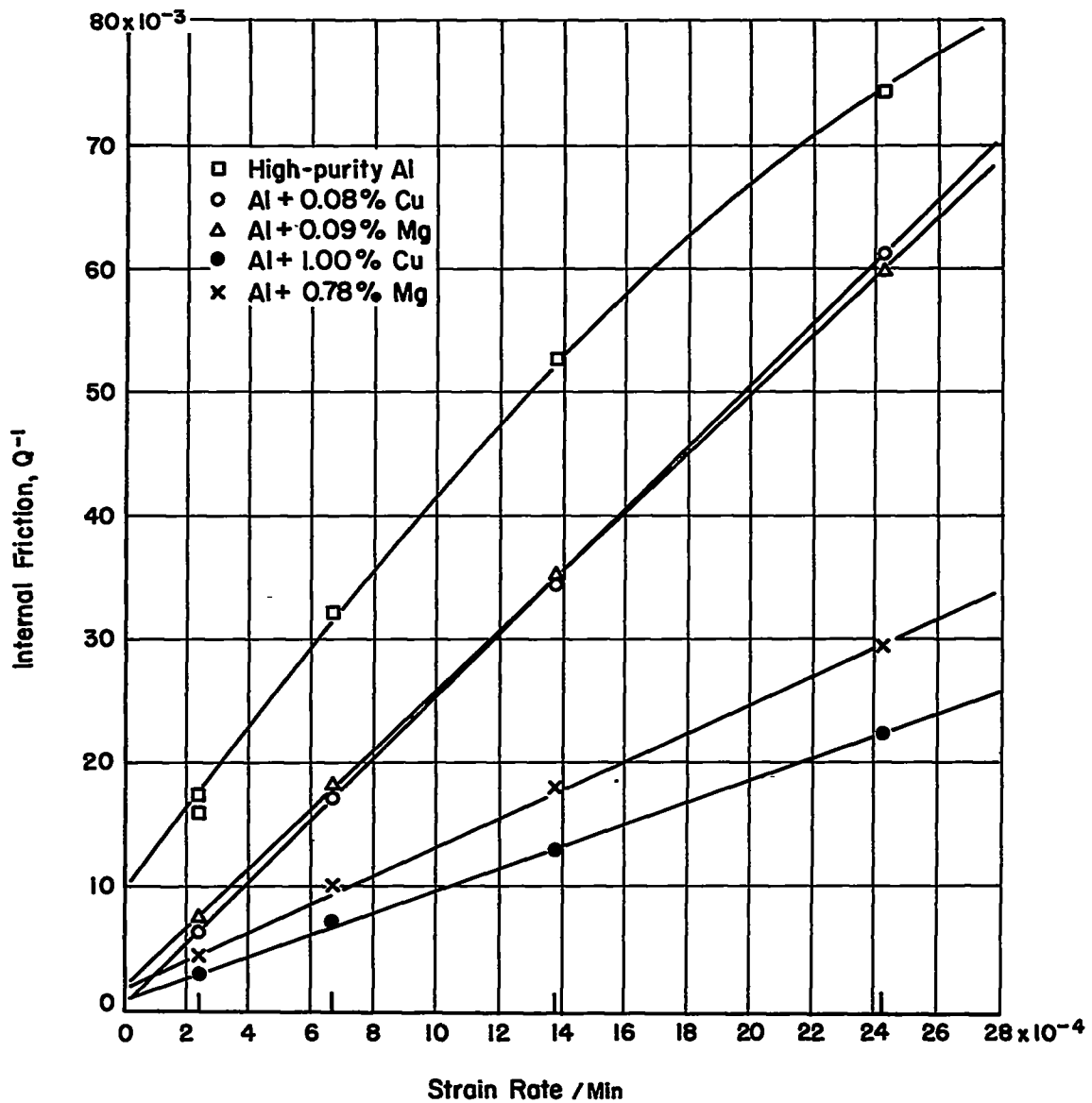
Figure 15.- Variation of internal friction during deformation with strain rate for different aluminum alloys (at 1-percent strain).
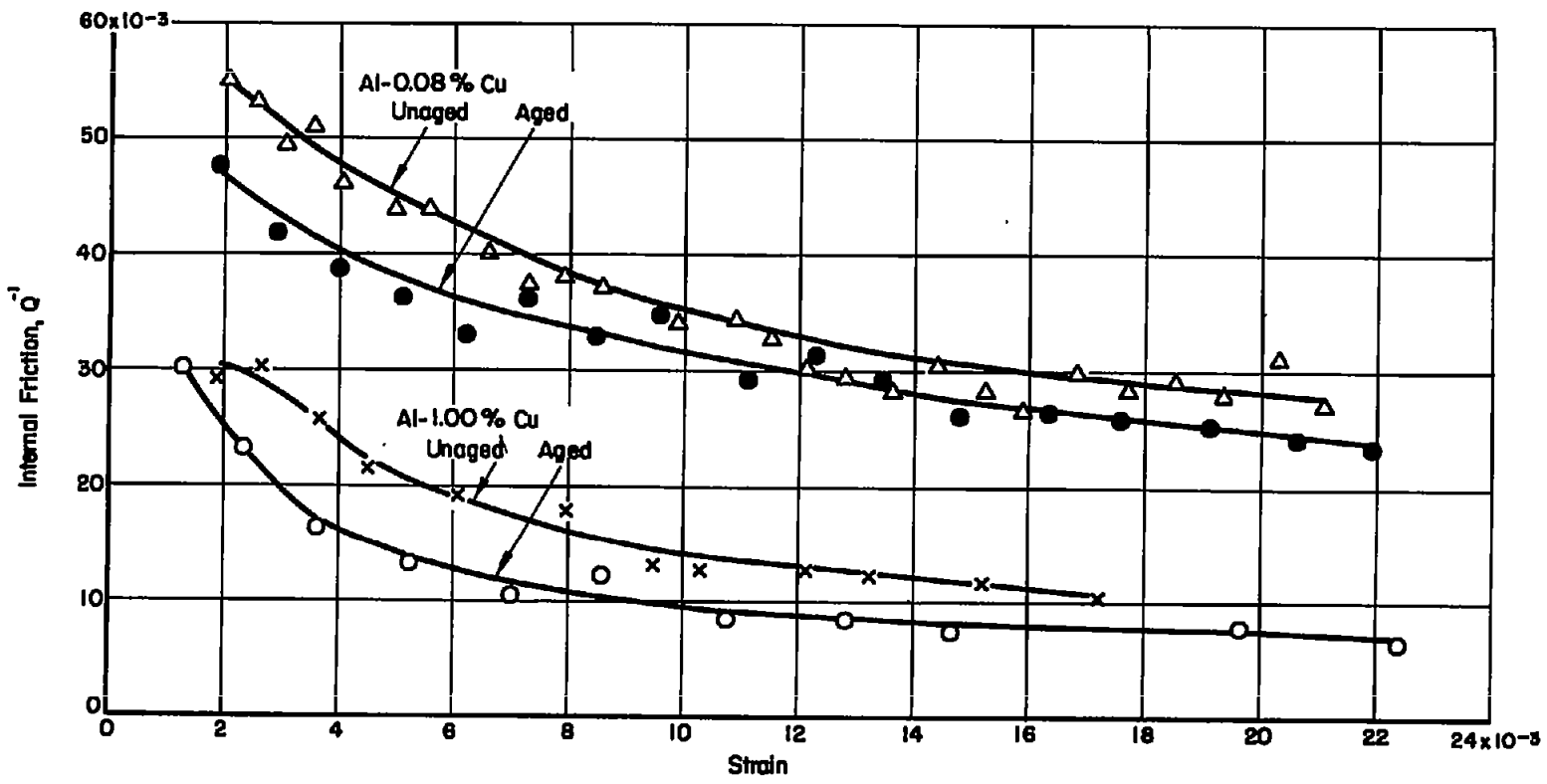
Figure 16.- Effect of aging 17 hours at 300° C on internal friction during deformation of aluminum-copper alloys. Strain rate, 0.00136 per minute.

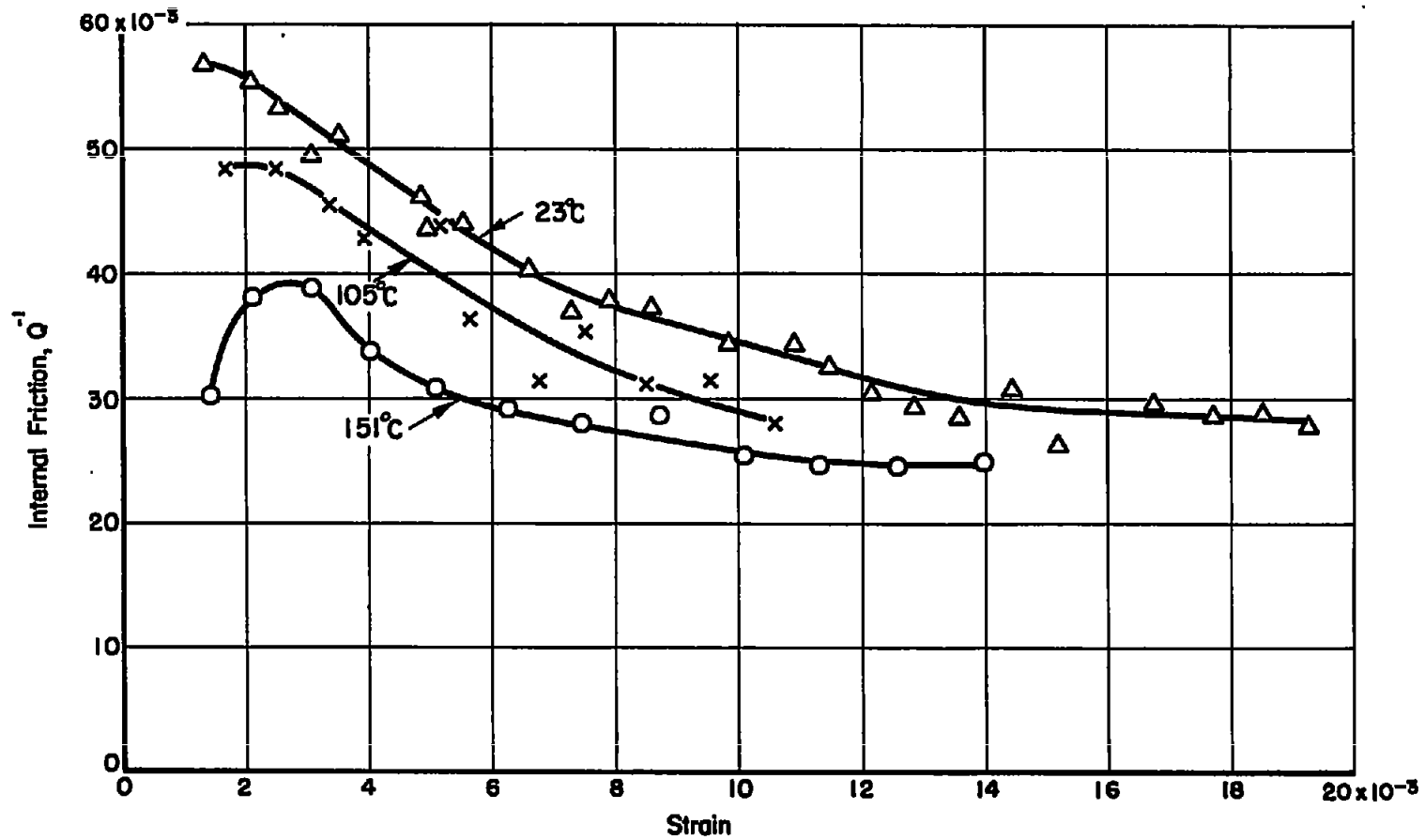Figure 17.- Dependence of internal friction during strain on temperature for aluminum alloy containing 0.08 percent copper. Strain rate, 0.00138 per minute.
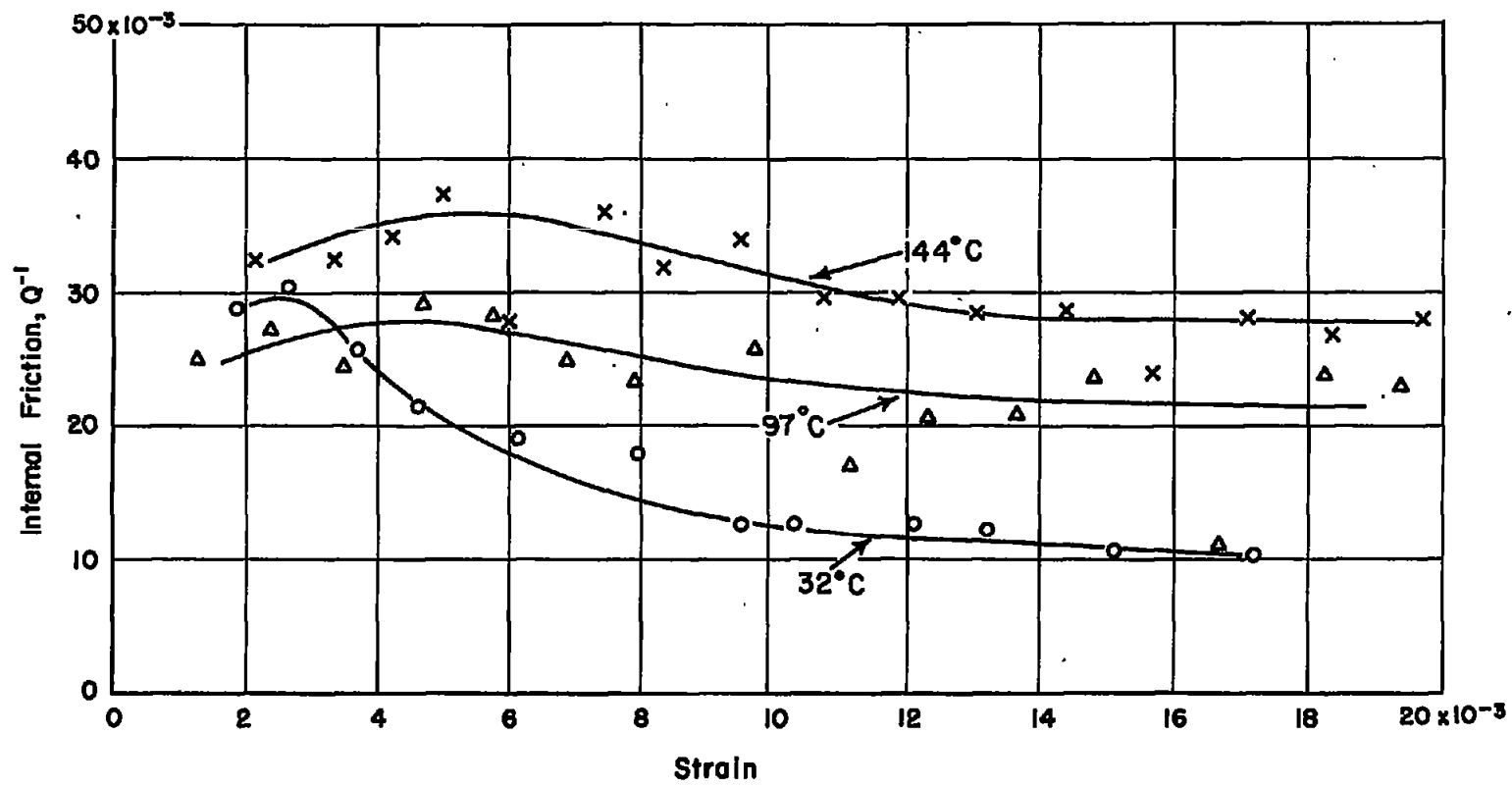
Figure 18.- Dependence of internal friction during deformation on temperature for aluminum alloy containing 1.00 percent copper. Strain rate, 0.00138 per minute.
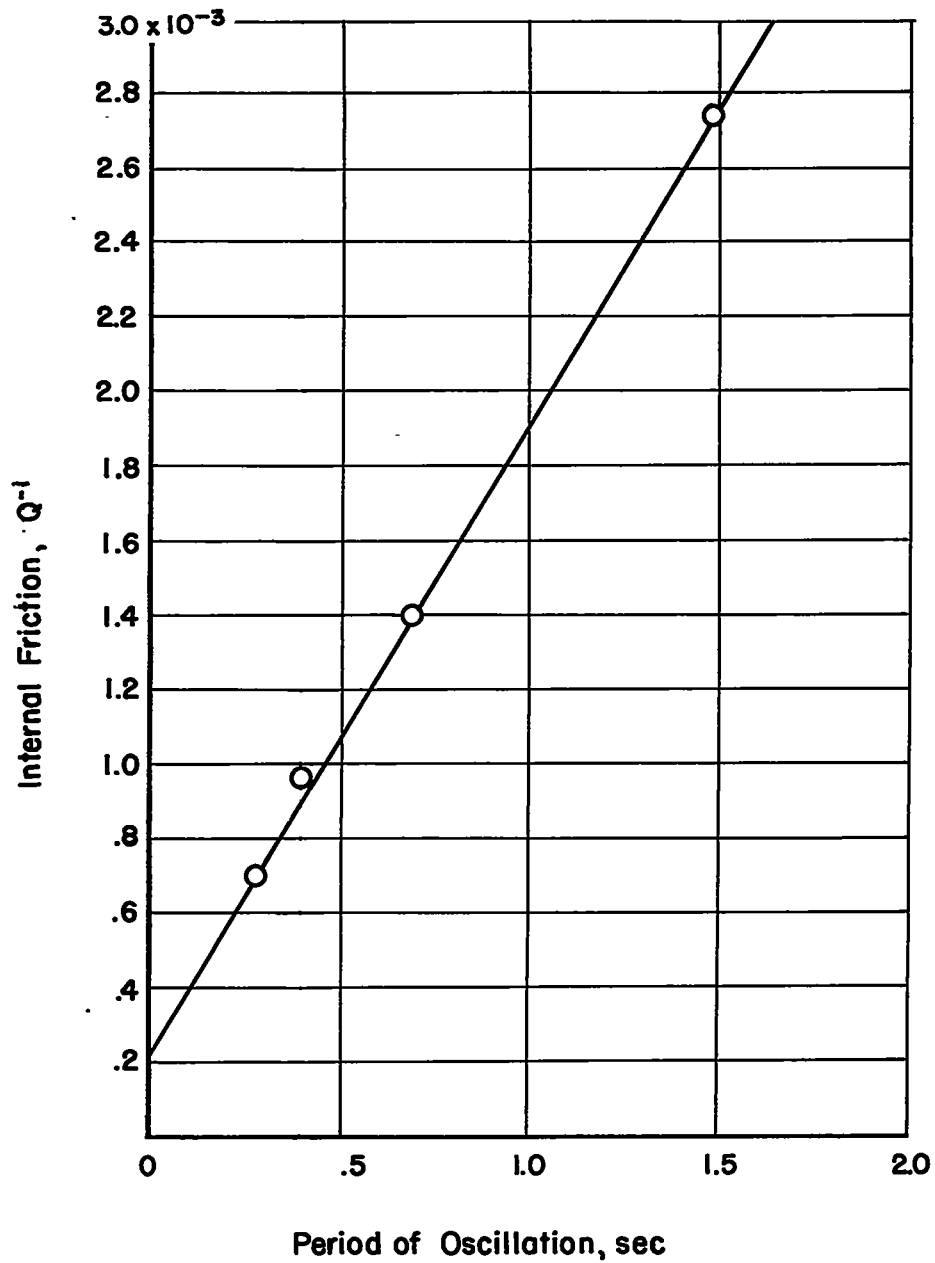
Figure 19.- Variation of internal friction during deformation with period of oscillation for high-purity aluminum. Strain rate