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SOME OBSERVATIONS RELATING TO RECOVERY OF INTERNAL FRICTION DURING FATIGUE OF ALUMINUM

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California Institute of Technology

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SUMMARY

Recovery of internal friction during periods of rest in specimens subjected to fatigue stresses in torsion has been studied experimentally with high-purity aluminum. The effective heat of activation for the process is found to be about 10,000 calories per gram molecule or less. The idea is presented that the dislocations responsible for the recovery of internal friction are the same as those responsible for fine slip which, according to one existing theory, is the mechanism responsible for fatigue failure. Experimental results indicate that obtaining a recovery factor independent of stressing history may possibly be associated with installing a cyclic process in which the subgrain structure is well established. The basic process then occurring may be one of indefinite to-and-fro motion of some free dislocations within a framework of immobilized arrays of dislocations responsible for the general plastic flow and the substructure.

INTRODUCTION

A considerable amount of experimental evidence is available which shows that when the internal friction of a freshly cold-worked material is measured during a subsequent period of rest there is a definite tendency for the internal friction to decrease as the length of the rest period increases. The extent of recovery is variable and is dependent on the composition of the material, the amount of cold-work, the temperature, and time. This effect is generally denoted as recovery of internal friction.

The first known study of recovery was made by Kelvin (ref. 1). He showed that sustained periods of vibration gave rise to a considerable increase in the internal friction and that subsequent periods of rest resulted in a large decrement in its value. Apparently, the next systematic investigation was conducted by Köster and Rosenthal (ref. 2). Nowick (ref. 3), while classifying the various kinds of internal friction observed in cold-worked metals, named that particular part which is characterized by a substantial decay in value during a period of rest the "Köster effect." Reference 4 briefly describes the recovery effect observed by the present author while conducting some work on 1100 aluminum and the purpose of the present report is to describe the observations made during a study of this problem.

The present investigation was conducted at the Guggenheim Aeronautical Laboratory of the California Institute of Technology under the sponsorship and with the financial assistance of the National Advisory Committee for Aeronautics. It is with pleasure that the author acknowledges the helpful suggestions received from Professor Sechler.

SYMBOLS

B, B_1, C, C_1	constants
Q	heat of activation for recovery of δ_2 damping
R	gas constant
T	absolute temperature at which process is taking place
t	time variable
γ	recovery factor, ratio of damping at end of 1-minute rest to damping at end of 30-minute rest
$\delta = \delta_1 + \delta_2$	
δ_1	damping due to grain boundaries, subgrain boundaries, and slip interfaces
δ_2	damping due to some relatively free dislocations found within subgrains
η	number of free dislocations responsible for δ_2 damping
σ	external stress acting on material
σ_d	back stress developed by piled array of dislocations
σ_i	internal stress opposing movement of dislocations

EQUIPMENT

The test equipment, which has been described in detail in reference 4, basically consisted of a torsional vibrator from one end of which was suspended a test specimen mounted on the top of an inertia bar. This is shown schematically in figure 1. The damping of the test specimen was measured by the decay of low-amplitude free vibrations in torsion and a photocell electronic counter system. The strain corresponding to the maximum amplitude of such vibrations in the decay measurements was 2.7×10^{-5} radians and the axial stress on the test specimen was less than 100 psi. The internal friction was calculated in a manner described in reference 4.

TEST MATERIAL AND SPECIMENS

The test material was high-purity aluminum of the following composition (the material and the composition being supplied by the Aluminum Co. of America): 99.996 percent aluminum, 0.003 percent zinc, and 0.001 percent silicon. The yield point and the ultimate strength in tension of the material as received were 6,750 psi at 0.2-percent-offset strain and 9,800 psi, respectively. The test material was supplied in the form of 1/2-inch-diameter rods 12 feet long and was described as hand-forged. There was evidence of a considerable amount of preferred orientation of the grain along the length of the rod.

The test specimens, which were of the geometrical form shown in figure 2, were machined directly from the material as received, the machine cuts in the final stages being of the order of 0.002 inch. The specimens were polished first with 600A grit emery paper and then in an electrolytic bath.

ANNEALING PROCESS

Preliminary experiments revealed that, depending upon the annealing time and temperature, two entirely different kinds of failures can be obtained. The low-temperature, short-time anneal was found under fatigue stressing to give rise to a crack formation and separation like the breakage of a piece of dried fibrous wood when subjected to twist perpendicular to the fiber direction. For a high-temperature, long-time anneal, the failure was characterized by a partition surface consisting of roughly concentric rings around a starlike formation obviously caused in the final stages of failure by the tensile load of 100 psi acting on the specimen. These two types of failures are shown schematically in figure 3.

It was also found that the high-temperature, long-time anneal (6 hours at 900° F) gave rise to a rather large grain size, 2 to 3 millimeters or larger; in some cases the grains occupied the whole diameter of the specimen (1/8 inch) and were up to 1 centimeter in length. In addition, it was found that, if the inertia bar end was not restrained, such specimens exhibited large permanent deformation under fatigue, the process apparently being one of displacement across the grain boundaries with well defined large slip surfaces. Normally no restraint is necessary since the inertia bar is stationary due to the loose inertial coupling offered by the test specimen to the vibrator. Under these conditions a nonoscillating, irreversible deflection of the inertia bar is a measure of the creep deformation which the test specimen may undergo under fatigue conditions. A plot of this torsional deformation against time gave the characteristic creep-curve features, the amount of deformation being a function of temperature. It is to be noted that this creep deformation was not observed when the specimen was not being subjected to torsional fatigue stressing.

Since it was inconvenient to work with the large grain size, lower temperature and shorter annealing times were tried; the chosen criterion for the selection of time and temperature was the removal of the preferred orientation structure, resulting in grain refinement but no grain growth. It was found that an anneal of 1/2 hour at 600° F may be applied to achieve this condition successfully, and this combination was used for annealing all specimens used in this work. As may be expected, this also resulted in much smaller permanent torsional deformations. The average grain size obtained under these conditions of annealing was 0.04 millimeter.

TEST PROCEDURE

The specimens as obtained from the electrolytic bath were carefully mounted on the testing machine and the temperature of the furnace enclosing the test specimen was raised to 600° F in about 10 minutes and kept at that temperature for 1/2 hour. At the end of this period the specimen was cooled to room temperature with the furnace opened. Subsequently, the temperature was raised to the desired value before starting a test. This procedure was found desirable since there was evidence of grain growth if the test specimen was cooled at a lower rate with the furnace closed.

The damping of the specimen thus prepared was measured at the test temperature several times before subjecting it to fatigue stress. In general it was found that there was a variation about the mean value of the order of 5 percent. After this measurement, the specimen was subjected to torsional fatigue at 236 cycles per second for a specified time and the damping was measured during a subsequent period of rest of about 1/2 hour or more. In the room-temperature test a minor variation was adopted. At the end of 1 minute of fatigue stressing, a rest period of 20 hours or more was given during which time the damping was determined; the subsequent periods of rest after further fatigue stressing were again of 1/2-hour duration. In addition, another 20-hour rest period was given during the latter stages of the stressing history of the same series of specimens, the damping again being determined to find if there was any correlation between stressing history and short- and long-time recovery.

The critical temperature for stress relaxation across the grain boundaries, as defined by the relation of internal friction to temperature, was found to be 575° F. The recovery tests were restricted to temperature equal to or less than 300° F, this temperature being in general the minimum recrystallization temperature for aluminum. The chosen maximum amplitude for measurement of decay was found to be just within the range at which the decay was amplitude independent. There is a possibility that the period of rest during which the recovery of internal friction was measured will change the fatigue life of the test specimen. In view of the difficulty in correlating the results from different specimens, it was decided to use the same specimens for any one test involving the same stress and the same temperature. The following torsional stress levels and test temperatures were chosen for this investigation:

(1) maximum shearing stress: 4,000, 5,000, 6,000, and 7,000 psi (nominal)

(2) temperature: 75°, 200°, 250°, and 300° F

A test of 150° F was performed to get additional information at the 7,000-psi stress level.

DISLOCATION MODEL FOR ANALYSIS AND INTERPRETATION OF RESULTS

The model developed for analyzing the results closely parallels that proposed by Kuhlmann and Nabarro and discussed by Cottrell (refs. 5 and 6) for explaining the elastic aftereffect and the recovery in cold-worked metals; the recovery process here is understood to involve a reduction of the internal stresses during annealing below recrystallization temperatures.

It is assumed that the grain boundaries of an annealed specimen consist of a complex arrangement of dislocations so linked that they are relatively immobile. It is suggested that the energy dissipation by such a configuration in low-amplitude oscillation may be considered to be due mainly to the out-of-phase to-and-fro moment of these grouped dislocations. The detailed mechanism for this is not yet known. In addition, when such a specimen is subjected to repeated stresses, formation of slip on various intersecting planes takes place, eventually leading to polygonization and growth of subgrains (ref. 7 and others). According to the theory presented by Cottrell, it may be imagined that the array of slip planes contains piled dislocations building up a back stress. It is thought that the subgrain boundaries also are similarly arranged groups of dislocations responsible for small changes of orientation on either side. Under the influence of an external stress σ, an internal stress σ_i has to be overcome before a dislocation can be made to move. The dislocations so moved pile against a boundary building up a back stress σ_d. Clearly, according to this theory, plastic flow will stop when $\sigma_d = \sigma - \sigma_i$.

It is now imagined that, in addition to the dislocations discussed above, there exists another class of dislocations which have their sources within the subgrains and so are relatively free and hence more mobile. The stress σ_i to be overcome in order to set them in motion is relatively small compared with that discussed by Nabarro in order to explain mechanical hysteresis. The to-and-fro motion of these dislocations appears to result in the fine slip observed by Wood and others (refs. 8 and 9) and considered by Wood to be the basic mechanism underlying fatigue damage. According to Wood, the to-and-fro motion of the free dislocations can build up intense bands of fine slip resulting in a local deterioration of structure leading to local crack initiation and ultimate fatigue failure.

It is probable that if the internal friction of a test specimen were to be measured immediately after a period of fatigue stressing, the total damping measured would consist of two components, δ_1 and δ_2, where δ_1 is the component due to the out-of-phase movement of the dislocations in the slip interfaces and subgrain and grain boundaries and δ_2 is the result of the motion of the relatively free dislocations having their sources within the subgrains. It seems reasonable to assume that, during a period of rest, these free dislocations spontaneously move toward the boundaries of the subgrains and temporarily get bound to them and thus do not contribute to the energy dissipation of the δ_2 variety. This in effect means that during a period of rest there should be a reduction of damping from $\delta = \delta_1 + \delta_2$ to δ_1 over a period of time. This explanation seems to be substantially the same as that given by Nowick in reference 3 to explain the Köster effect. Since these dislocations are considered to be bound to the boundaries rather loosely, a subsequent period of fatigue stressing is expected to activate them again, almost instantaneously bringing back the damping to the value $\delta_1 + \delta_2$, and to continue to produce additional composite fine slip regions. It is evident that, starting from an annealed specimen, δ_1 itself is variable and in general is expected first to increase with the stressing history and then to become roughly a constant or probably decrease with stressing history as some of the reported investigations indicate. This reduction of damping after a prolonged period of cycling seems to fit the general pattern of the metallographic observations of fatigue test specimens. Thus, Forsyth, in reference 9, and others indicate that fatigue accelerates recrystallization even at ordinary temperatures. Since, in general, δ_1 is completely reduced to the annealed value only by recrystallization, any effort in that direction contributed by fatigue may also be expected to reduce the value of δ_1. Thus it appears that the reduction from the peak value of δ_1, as compared with the annealed value, may be a measure of the amount of recrystallization effected by fatigue. The basic idea that is presented here is that the relatively free dislocations responsible for damping of the δ_2 variety are the same as those responsible for fine slip which, according to Wood, is the basic mechanism underlying fatigue damage.

In passing it may be mentioned that there is really no conflict with the idea that fine slip is restricted in length to the size of the subgrains. The slip lines are not generally absolutely straight and it is suggested that under larger magnifications they may be found to be clusters of slip lines whose lengths are of the same order as those of the subgrains and have the same general direction within the grain as the observed slip lines. Also since intensification of fine slip is observed during progressive fatigue cycling, it is apparent that some system must exist which continuously replenishes the dislocations destroyed in the production of fine slip. As will be shown later this replenishing process seems to stabilize to a constant value after a certain time, the value and the time depending upon the stress level of fatigue cycling.

Let η be the number of relatively free dislocations piled up within the subgrains immediately after a period of fatigue stressing. The back stress σ_d caused by these dislocations is clearly proportional to η. The activation energy needed for spontaneous movement of such dislocations should be proportional to $\sigma_i - \sigma_d$ where σ_i, as mentioned above, is the stress to be overcome to move these dislocations. This stress may be comparable to a kind of frictional force and considered relatively constant. Since σ_i is assumed to be a constant, the activation energy should then be linearly related to σ_d alone, that is, to η alone. Therefore, one may write the rate of spontaneous movement of dislocations

$$\frac{d\eta}{dt} = -C_1 e^{-\left(Q - B_1\eta\right)/RT}$$

where $\left(Q - B_1\eta\right)$ is the effective activation energy. As mentioned above, during a period of rest this spontaneous movement results in a more stable arrangement of these dislocations which are then temporarily unconcerned with the energy dissipation. It is also assumed that the value of δ_2 is proportional to the number of free dislocations available for energy dissipation at any time. That is,

$$\frac{d\delta_2}{dt} \propto \frac{d\eta}{dt} = -Ce^{-\left(Q - B\delta\right)/RT}$$

Since δ_1 is assumed to be a constant after any one particular period of fatigue, it follows that

$$\frac{d\delta}{dt} = -Ce^{-\left(Q - B\delta\right)/RT} \tag{1}$$

This relation then forms the basis for analyzing the results obtained in this investigation. Disregarding the negative sign for the time being, one obtains

$$\log_e\left|\frac{d\delta}{dt}\right| = \log_e C - \left[\left(Q - B\delta\right)/RT\right] \tag{2}$$

or, slightly rearranging terms,

$$\frac{d}{d\delta}\log_e\left|\frac{d\delta}{dt}\right| = \frac{B}{RT} \tag{3}$$

Equations (2) and (3) suggest that a plot of $\log_e\left|\frac{d\delta}{dt}\right|$ against $1/T$ and $\frac{d}{d\delta}\log_e\left|\frac{d\delta}{dt}\right|$ against $1/T$ should give linear relations. From the test results, these functions are determined and plotted in the manner described below.

RESULTS AND DISCUSSION

Recovery factors γ and values of internal friction δ_1 at the end of a 1-minute rest period are presented in table I for all stress levels and temperatures. The results for the 7,000-psi stress level are presented in the form of curves in figures 4 to 7.

The basic experimental results may be summarized as follows: Recovery of internal friction was noticed at all stress levels and temperatures. In addition to being a function of stress level and temperature, recovery also appears to be a function of stressing history. At least at room temperature, the recovery increases with increase of stressing history and eventually becomes a constant. At higher temperatures, the result is not so certain and a tendency toward a decrease of recovery with increase of stress history was observed. There also appears to be some evidence to the effect that recovery is inhibited at a high temperature, (the damping being measured at the same temperature) if fatigue stressing was done at room temperature. This aspect needs further investigation.

The effective heat of activation is computed from figure 5 which gives a roughly linear relation between $\log_{10}\left|\frac{d\delta}{dt}\right|$ and $1/T$. The slope of this line (changing from $\log_{10}$ to $\log_e$) then gives the effective heat of activation as 9,980 calories per gram molecule. The following values were used for this computation: Stress level, 7,000 psi nominal; duration of stressing history, 2.83×10^5 cycles; results analyzed for a value of $\delta = 30 \times 10^{-3}$. This particular value of δ was chosen since for this value there is an intercept of the recovery curves (fig. 4) at all temperatures. The points or curves for 300° F were not included in figures 5 to 7 since there was some uncertainty about the results and since 300° F is too close to the recrystallization temperature.

In figure 6, $\log_{10}\left|\frac{d\delta}{dt}\right|$ is plotted against δ and in figure 7 $\frac{d}{d\delta}\log_{10}\left|\frac{d\delta}{dt}\right|$ is plotted against $1/T$. These curves again give roughly linear relations which suggests that this approach may be a reasonable one. The value of $B\delta$ obtained from figure 7 is found to be roughly equal to 9,140 calories per gram molecule. The effective heat of activation for the process compares favorably with that of Fusefeld for brass as reported by Nowick in reference 10. Since this value is considerably lower than the heat of activation of volume diffusion for pure aluminum (about 37,500 calories per gram molecule, ref. 11), comments by other workers that recovery below recrystallization does not involve volume diffusion seem to be in order. It is more probable, as also suggested by other workers, that during recovery the dislocations responsible for it are not lost but simply get bound either to impurity atoms by seeking them or to subgrain boundaries. The value of $Q = 9,980 + 9,140 = 19,120$ calories per gram molecule does not seem to be related to any of the known heats of activation pertaining to other processes taking place in high-purity aluminum. It is not known to what extent the impurities, 0.003 percent zinc and 0.001 percent silicon, influence the result. The heats of activation for the diffusion of zinc and silicon in aluminum are not known.

One of the more interesting phases of these investigations is the variation of recovery as a function of stress level and stressing history. In reference 4 it was suggested that, except in one case, the recovery was not found to be a function of stressing history. This statement needs revision in the light of the present series of tests which are more exhaustive. The negative recovery factor, that is, an inhibition of the reduction of damping or possibly an increase in damping during a rest period, reported in reference 4 was not noticed again and there is a probability that it may be erroneous. What one may call a negative recovery was noticed in this work also but under somewhat different conditions. Before this aspect can be recognized as certain, further tests are needed.

Since it is impossible to extrapolate the recovery curve to the start of the rest period, the ratio of the internal friction at the first minute to that at some other time is adopted as a measure of recovery during that period of rest. In general, the value of the internal friction at the end of a 30-minute interval was taken as the denominator for determining the recovery factor. The recovery factors so determined and the values of the internal friction at the end of the 1-minute rest period are summarized for all stress levels and temperatures in table I. These values are plotted in figure 8 as a function of stressing history. Also plotted in figure 8 is a broken curve joining two points which correspond to recovery factors obtained for room-temperature tests (75° F) for 20-hour rest periods. The variation of damping beyond 20 hours of rest was insignificantly small and hence the value at 20 hours may be assumed to be constant and representative of the values beyond this period of rest.

One striking feature that can be observed from the curves in figure 8 is that at room temperature there is a progressive increase in the recovery factor which eventually tends to a constant with increase of stressing history, whereas at higher temperatures one notices a trend toward a decrease in the recovery factor with temperature and stressing history. However, in no case does one obtain a value less than 1 and as the temperature approaches 300° F, the recovery factor shows a tendency to be relatively independent of stressing history.

At room temperature with an annealed specimen the recovery factor seems to increase gradually, eventually reaching a relatively stable value after some period of fatigue stressing. At least in the beginning of the stressing history, there is a trend toward an increase in the recovery factor with an increase of stress for the same number of cycles of stress.

The relation between long-term recovery factors and stressing history corresponding to 1.42×10^4 cycles and more than 5×10^5 cycles is very interesting. These values are plotted as a function of stress in figure 9. From these curves one notices that for a short fatigue-stressing history the recovery factor increases with an increase of stress level in the test range and for a long fatigue-stressing history the recovery factor decreases with stress level. In fact the two curves tend to intersect in the vicinity of 7,500 psi. Apparently at this stress level the recovery based on long rest periods becomes independent of stressing history even at a stressing history of the order of 1.42×10^4 cycles. A constancy of the recovery factor with stressing history implies that $\left(\delta_1 + \delta_2\right)/\delta_1 = 1 + \left(\delta_2/\delta_1\right)$ is a constant. Since in these cases it was observed that after a long period of rest δ_1 is constant during the later stage of fatigue stressing, one may conclude that δ_2 is also a constant. This implies that the number of dislocations involved in fine slip is constant after a certain period of fatigue stressing. Thus, as the stress level is progressively increased, the period of fatigue stressing needed for a constant recovery factor at any particular stress level gradually decreases; that is, a cyclic pattern is established in which the subgrain structure is well defined, and the basic process then occurring may be one of indefinite fine slip caused by the to-and-fro motion of the free dislocations responsible for the damping of the δ_2 variety. The fact that the higher the stress level, the earlier this pattern is established is further supported by the nature of the fatigue failure, that is, an inverse relation between the stress level and the number of cycles. This result and its interpretation in terms of free dislocations appear to be further evidence in support of Wood's theory of fatigue failure. The explanation of the curves for higher temperatures must await further work.

One series of tests yielded results which as yet are not explained; however, the results are shown graphically in figure 10. In this series, the test specimens were subjected to fatigue at 7,000 psi for 2.12×10^5 cycles and the damping during a subsequent period of rest was determined. In one case the rest period was always at room temperature and the curve typical for the 7,000 psi level for room temperature was again obtained. In the others, after the first measurement was taken at room temperature $1\frac{1}{2}$ minutes after the start of the rest period, the temperature of the specimens was raised to 200°, 250°, and 300° F in about 4 minutes and the damping was measured during the subsequent rest periods at these temperatures. The damping of an annealed specimen showed that within this temperature range the variation of damping was relatively small. It was found, however, that the damping of the other test specimens was inhibited, and, in particular at the higher temperatures (250° and 300° F), there was an increase of the damping as compared with the value determined at room temperature soon after the fatigue stressing was stopped. Cooling the specimens to room temperature revealed that the damping at room temperature is in general less for these heated specimens than for those given a period of rest at room temperature. The room-temperature damping was in general found to be lower, the higher the temperature at which the specimen was rested. In no case was the value of the damping so obtained comparable to the value obtained in the annealed condition. That is to say, the recovery was incomplete. In addition, a determination of the δ against T relation of these specimens indicated a shift of the temperature peak for relaxation across the grain boundaries to a lower temperature. This shift may be interpreted to mean a smaller grain size (caused by fatigue stressing) which is in conformity with the nature of the influence of cold-work, since it is a well established fact that cold-work results in crystal breakup. It is not quite clear why recovery should be inhibited by high temperatures if fatigue stressing was done at room temperature.

SUMMARY OF RESULTS

The following results were obtained from the investigation of recovery of internal friction during periods of rest in specimens of high-purity aluminum subjected to fatigue stresses in torsion:

1. It was found that the effective heat of activation for the process is approximately 10,000 calories per gram molecule.

2. Variation of recovery with stress level and history indicates that recovery becomes independent of stressing history sooner at higher stress levels. It appears that this recovery may be associated with

the establishment of a subgrain structure and the basic process then happening may be one of indefinite to-and-fro motion of some free dislocations within the subgrains.

3. It is suggested that the dislocations responsible for recovery are the same as those responsible for fine slip, which, according to one existing theory, is the mechanism responsible for fatigue failure.

California Institute of Technology,
Pasadena, Calif., August 1, 1956.

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TABLE I

SUMMARY OF RESULTS OF RECOVERY TESTS

75° F			150° F			200° F			250° F			300° F		
Cycles of stress	δ_1 at 1 min	γ	Cycles of stress	δ_1 at 1 min	γ	Cycles of stress	δ_1 at 1 min	γ	Cycles of stress	δ_1 at 1 min	γ	Cycles of stress	δ_1 at 1 min	γ
Stress level, 4,000 psi														
0	2.63×10^{-3}					0	2.72×10^{-3}		0	6.15×10^{-3}		0	9.51×10^{-3}	
$.142 \times 10^{5}$	25.8	1.55				2.83×10^{5}	39.2	1.71	1.42×10^{5}	52.3	1.55	$.142 \times 10^{5}$	46	1.15
1.42	35.3	1.78				5.66	45.9	1.52	2.83	54.4	1.36	2.83	64.5	1.21
2.83	38.5	1.94				7.496	49.8	1.41	5.66	54.4	1.18	5.66	56.3	1.34
5.66	39.9	2.07				22.49	45	1.49	7.496	53.6	1.29	7.496	62.7	1.09
14.99	31	2.45							22.49	61	1.22			
29.98	69	2.05												
44.97	70.5	1.94												
59.96	81	2.03												
89.94	77	2.03												
119.92	75.7	2.03												
Stress level, 5,000 psi														
0	5.35×10^{-3}					0	2.72×10^{-3}		0	7×10^{-3}		0	7.3×10^{-3}	
$.142 \times 10^{5}$	24	1.52				$.142 \times 10^{5}$	25.9	1.44	$.142 \times 10^{5}$	26.6	1.31	$.142 \times 10^{5}$	28.4	1.19
1.42	34.5	1.59				2.83	40	2.16	2.83	33	1.20	2.83	38.4	1.19
2.83	49.8	2.41				5.66	57.5	1.79	5.66	41.5	1.19	7.496	48.4	1.15
5.66	54	2.31				7.496	40	1.49						
7.496	48.8	2.09												
14.99	60.5	2.25												
Stress level, 6,000 psi														
0	2.7×10^{-3}					0	2.5×10^{-3}		0	4.1×10^{-3}		0	6.5×10^{-3}	
$.142 \times 10^{5}$	30.5	1.64				$.142 \times 10^{5}$	29.7	1.62	$.142 \times 10^{5}$	44.8	1.4	$.142 \times 10^{5}$	32.3	1.15
1.42	37.2	1.99				1.42	32.3	1.43	1.42	43.1	1.04	1.42	44.7	1.61
2.83	43	1.85				2.83	38	1.35	2.83	52.2	1.50	2.83	50.4	1.49
5.66	43	1.94												
7.496	42.2	2.04												
Stress level, 7,000 psi														
0	3.2×10^{-3}		0	2.2×10^{-3}		0	29×10^{-3}		0	4.8×10^{-3}		0	5.75×10^{-3}	
$.142 \times 10^{5}$	22.9	1.52	$.142 \times 10^{5}$	22.6	1.38	$.142 \times 10^{5}$	24.3	1.5	$.142 \times 10^{5}$	28.5	1.51	$.142 \times 10^{5}$	31.2	1.26
1.42	31	2.08	1.42	28.9	1.37	1.42	28	1.37	.708	34.3	1.48	.708	36.4	1.31
2.83	37.2	2.06	2.83	33.8	1.39	2.83	31.4	1.25	1.42	35.4	1.42	1.42	32.5	1.18
5.66	34.9	2.11							2.83	42.2	1.44	2.83	69.3	1.26

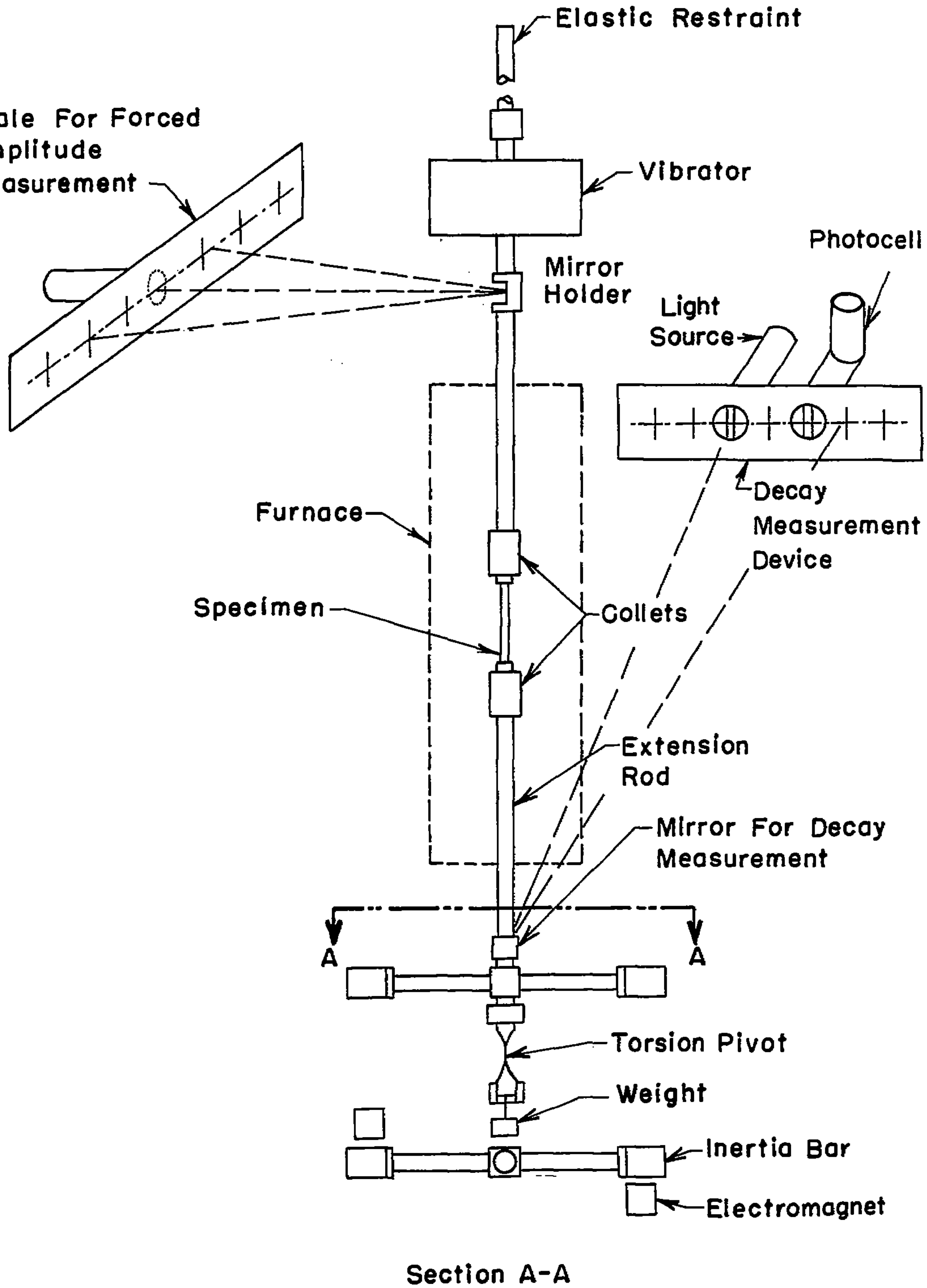


Figure 1.- Schematic diagram of test setup.

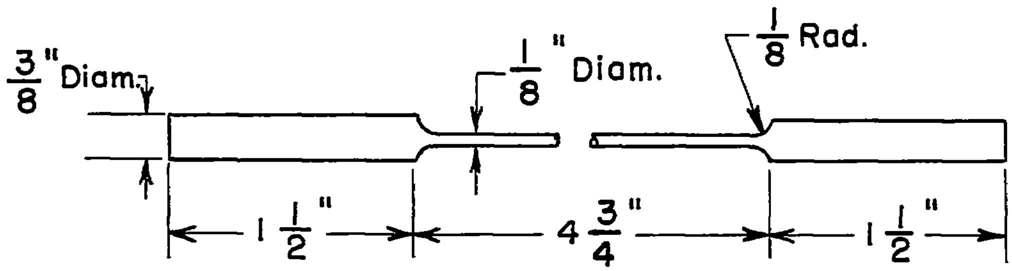
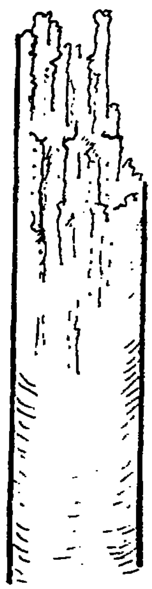
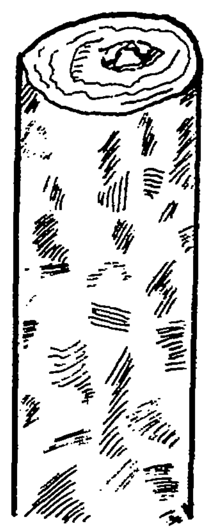


Figure 2.- Dimensions of test specimen.

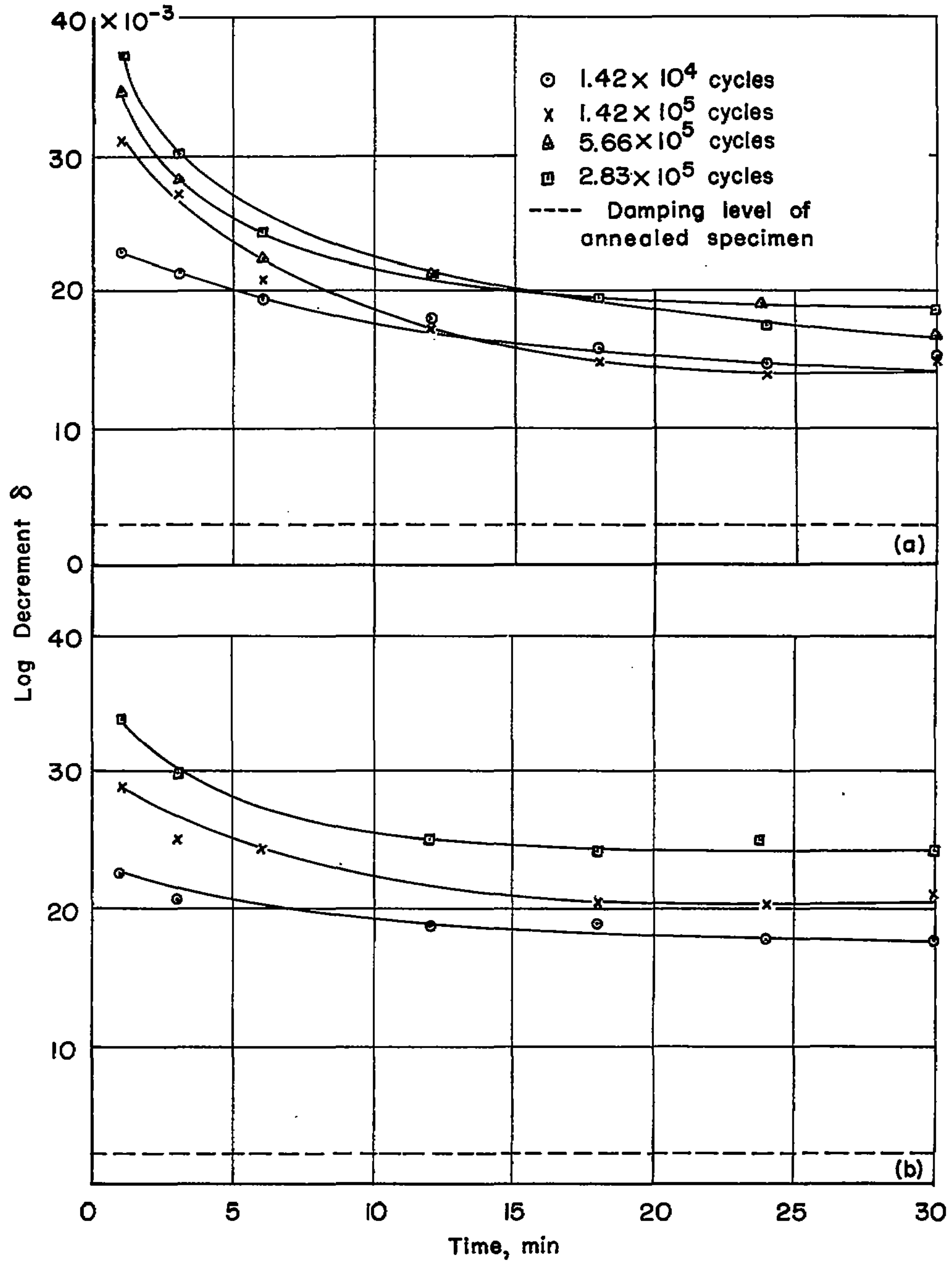


(a) Short-time, low-temperature anneal.



(b) Long-time, high-temperature anneal.

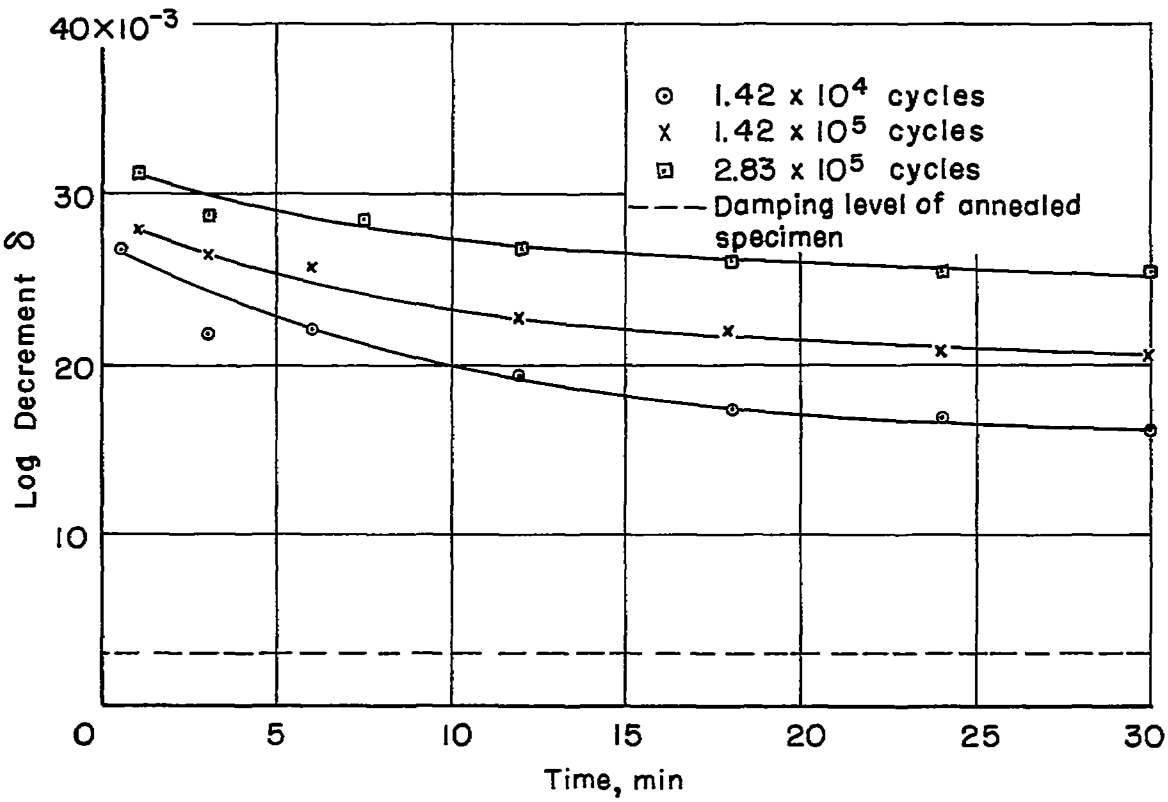
Figure 3.- Typical failures of test specimens.



(a) Tested at 75° F.

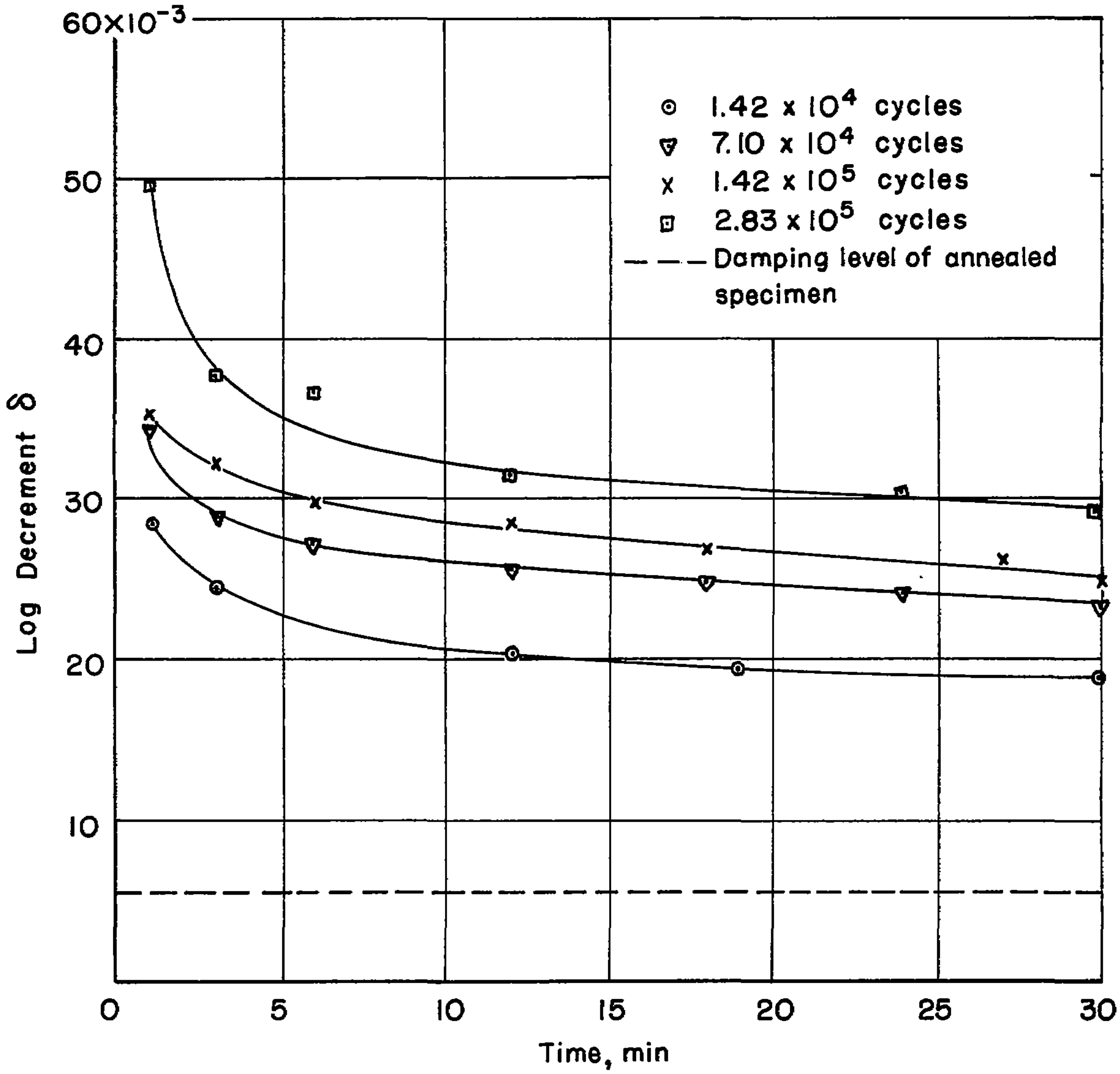
(b) Tested at 150° F.

Figure 4.- Recovery of internal friction during rest period. Samples tested under 7,000 psi.



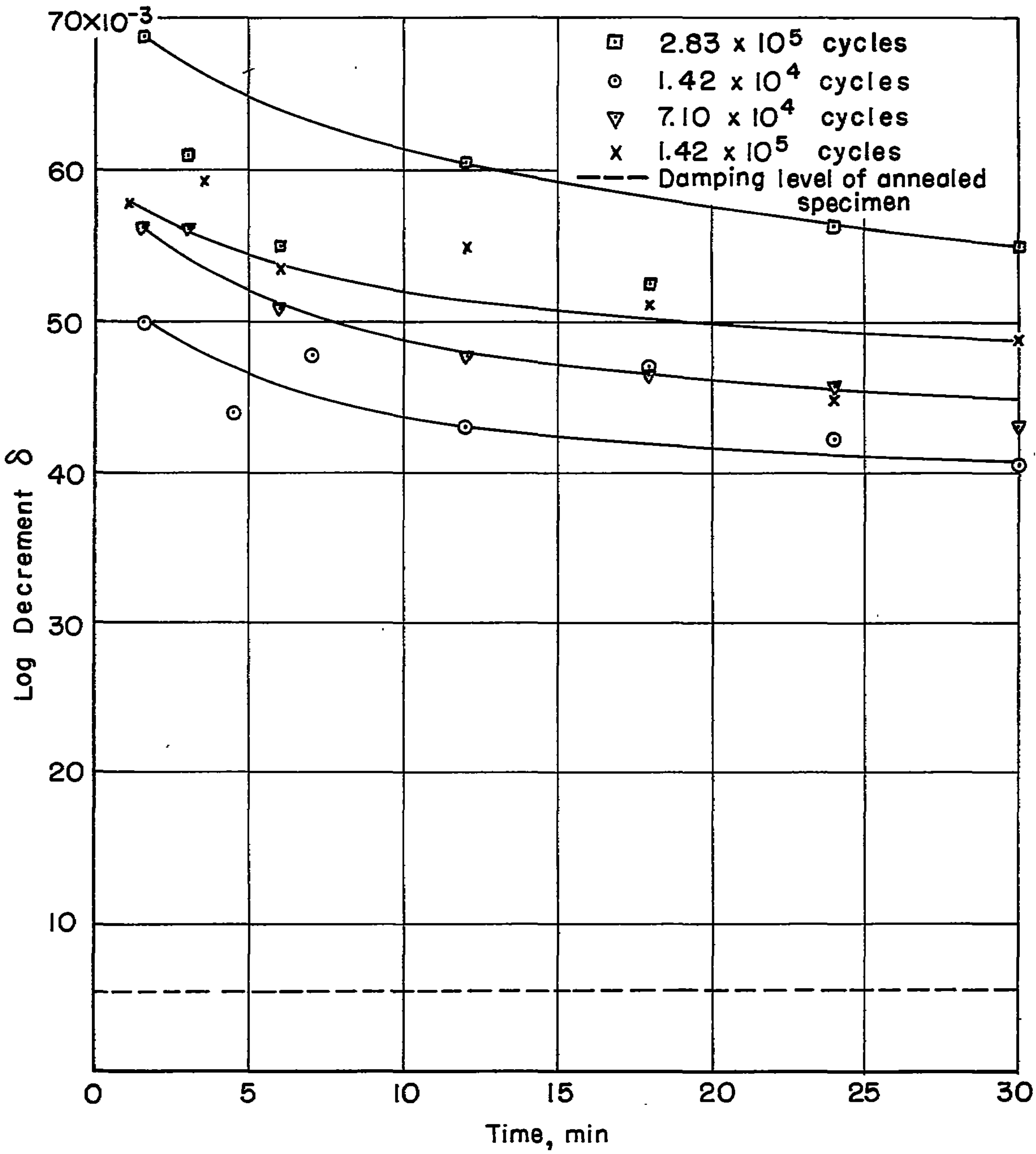
(c) Tested at 200° F.

Figure 4.- Continued.



(d) Tested at 250° F.

Figure 4.- Continued.



(e) Tested at 300° F.

Figure 4.- Concluded.

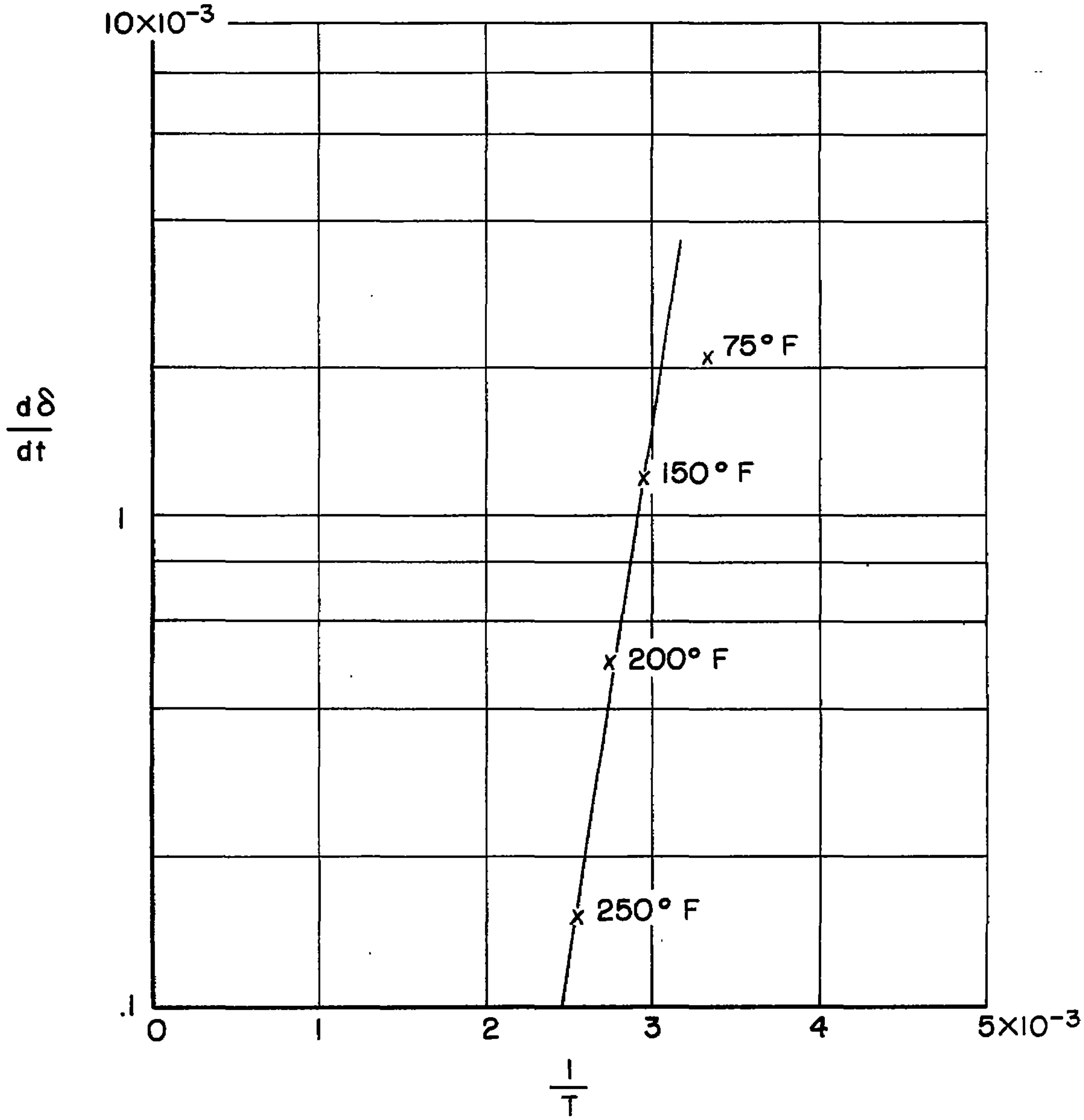


Figure 5.- Plot of $d\delta/dt$ against $1/T$ during period of rest after 2.83×10^5 cycles of fatigue stressing at 7,000 psi.

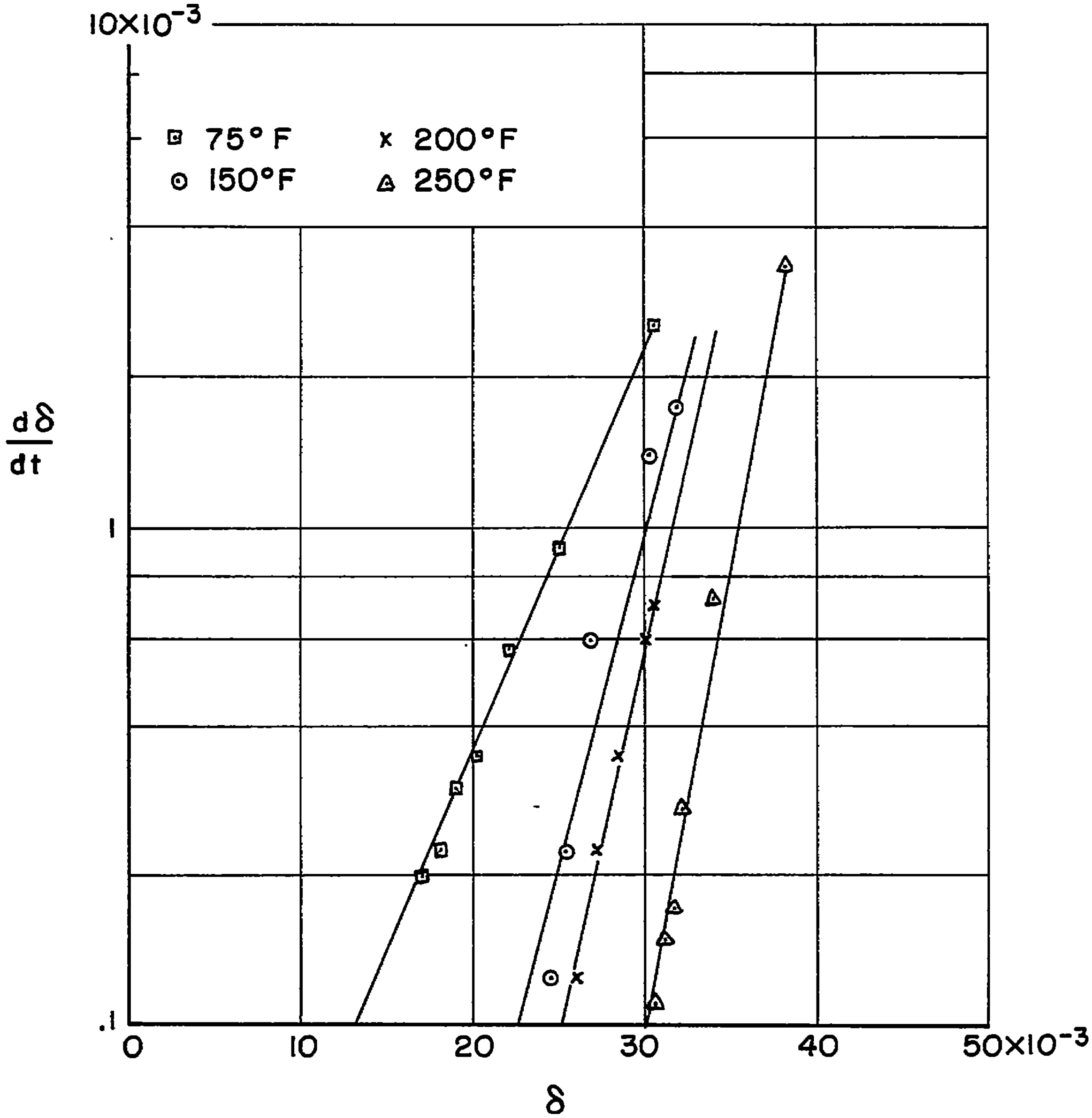


Figure 6.- Plot of dδ/dt against δ during period of rest after 2.83×10^5 cycles of fatigue stressing at 7,000 psi. Decay life, 20 minutes.

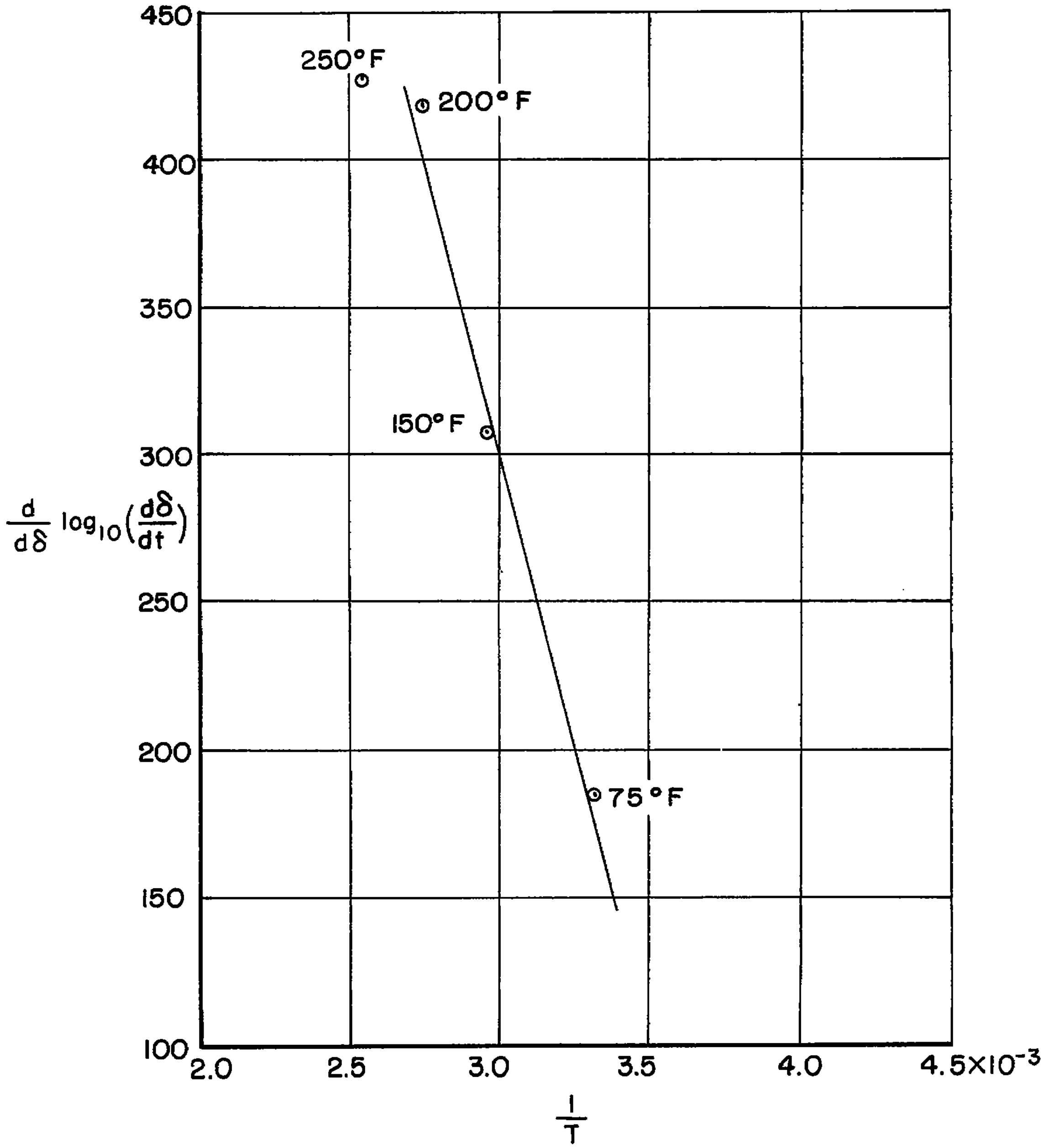
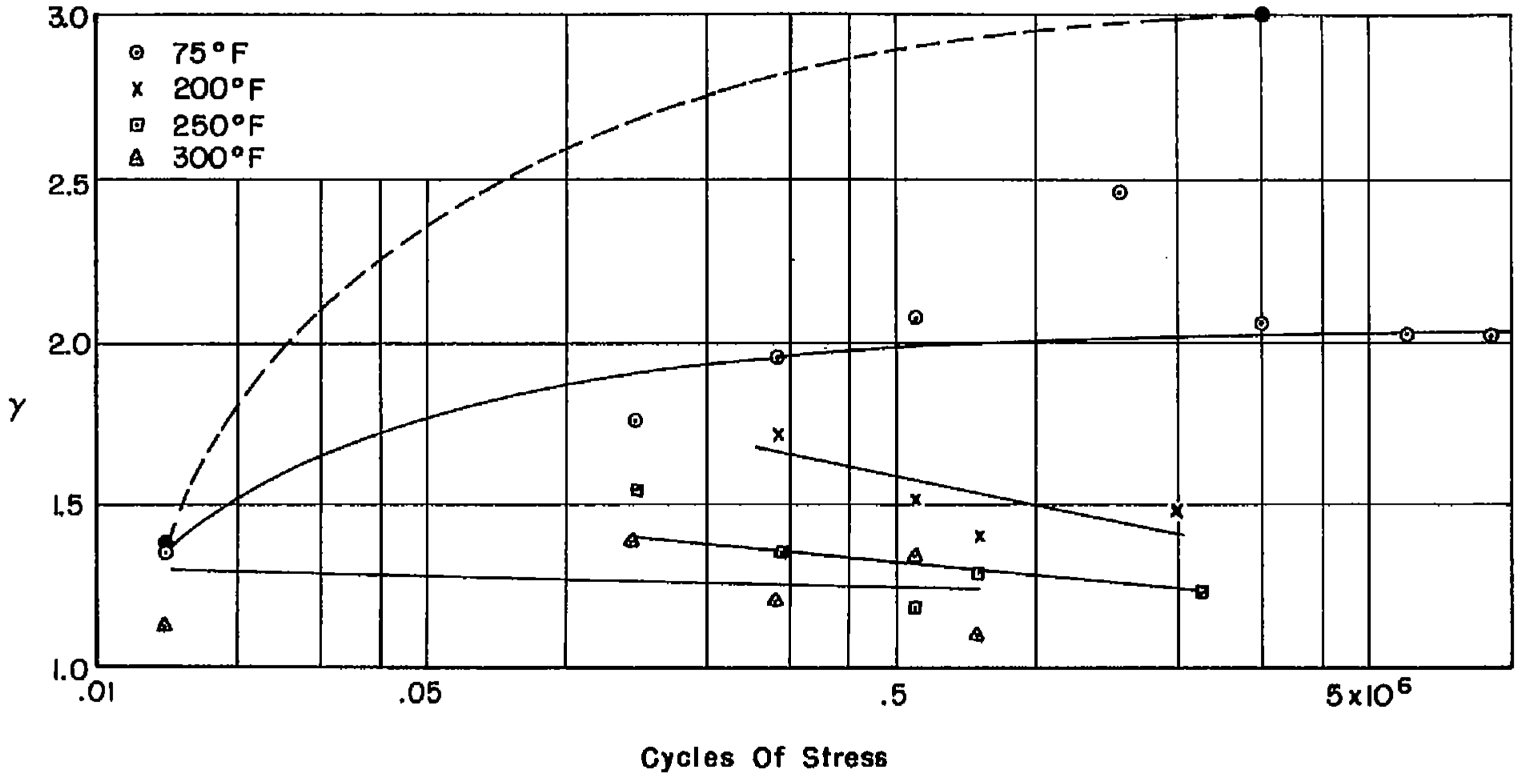
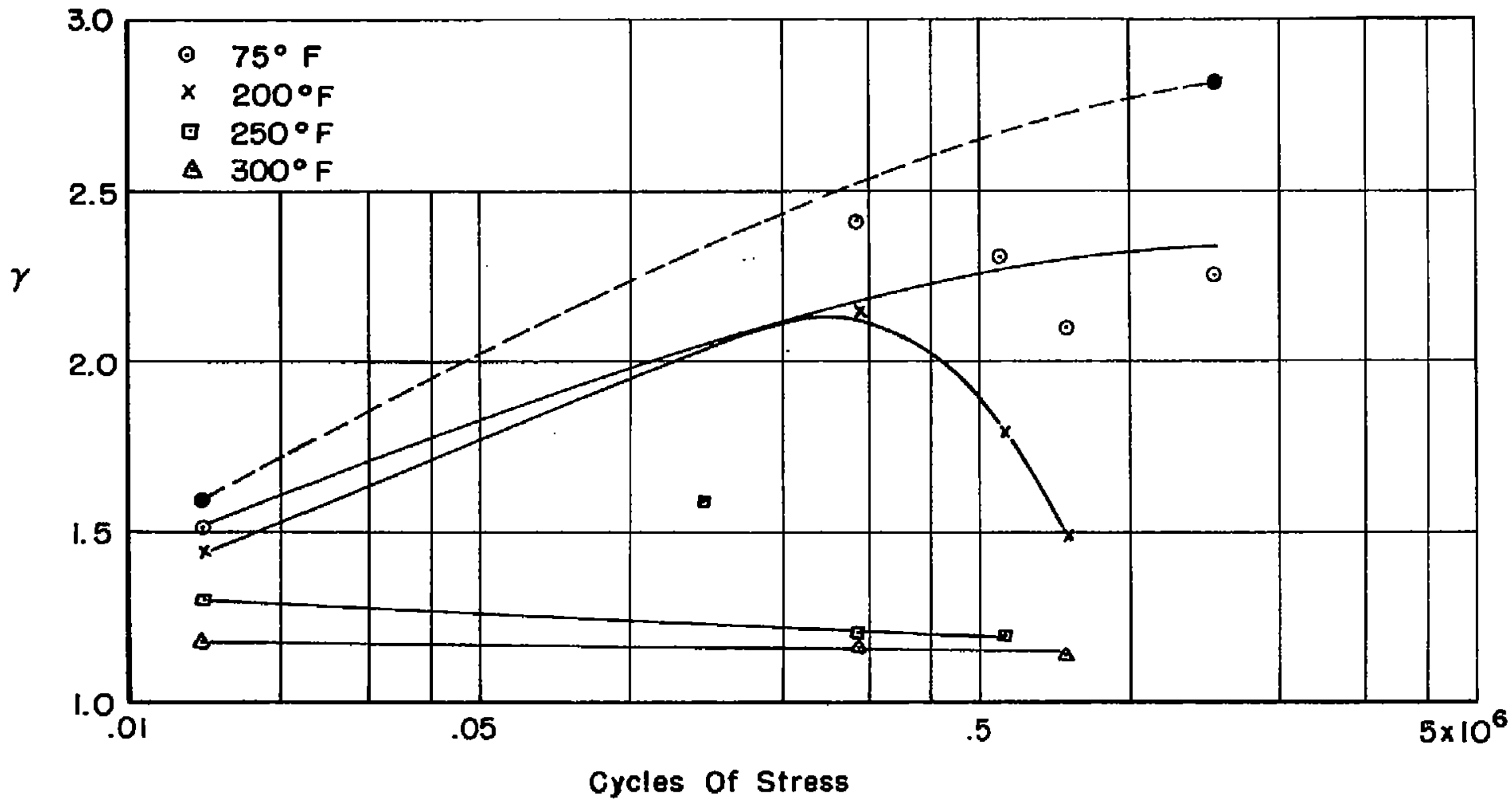


Figure 7.- Plot of $(d/d\delta)\log_{10}(d\delta/dt)$ obtained from figure 6 against $1/T$.



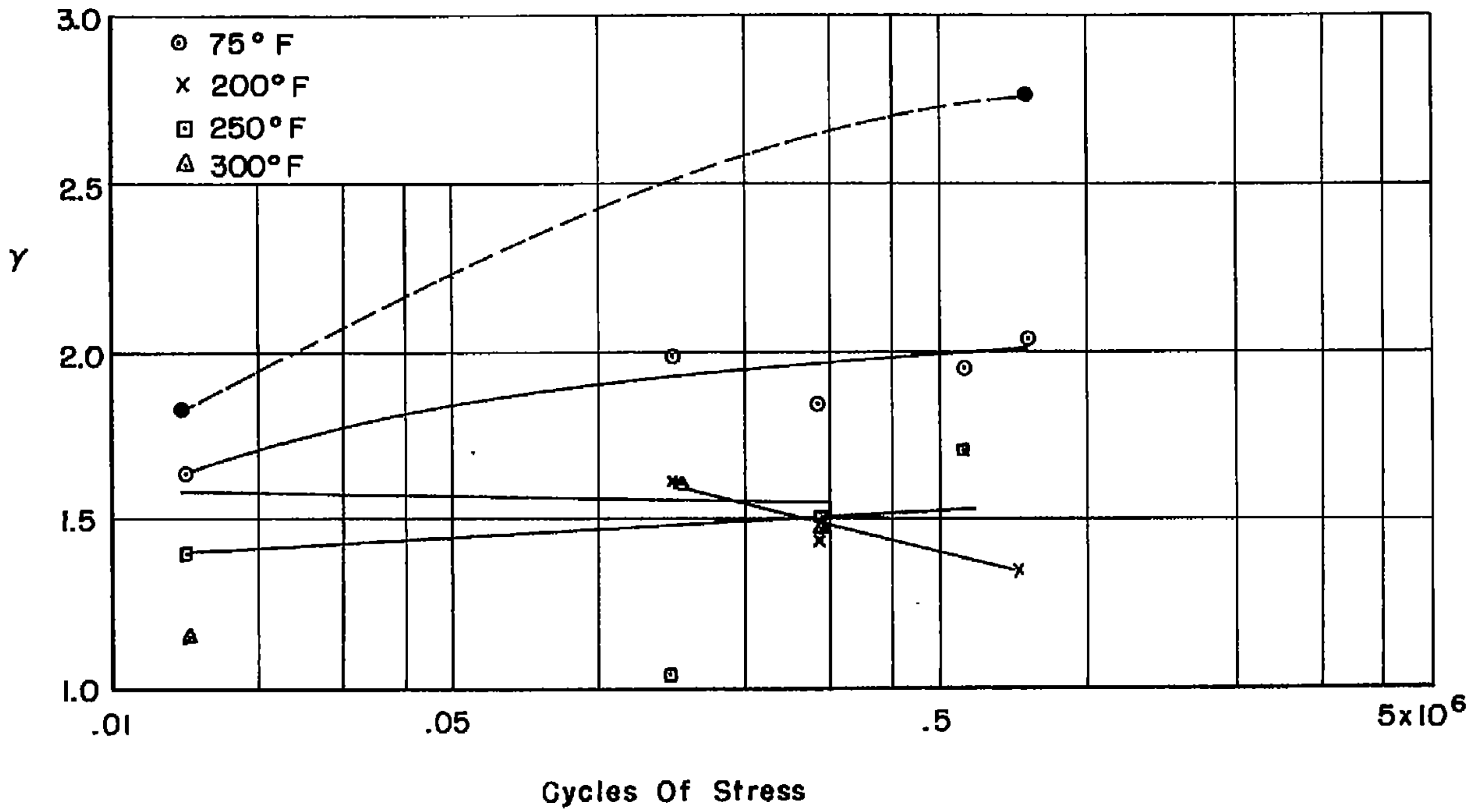
(a) Stress, 4,000 psi.

Figure 8.- Variation of recovery with stressing history. Broken curves join points obtained at 75° F for recovery factors for 20-hour rest periods.



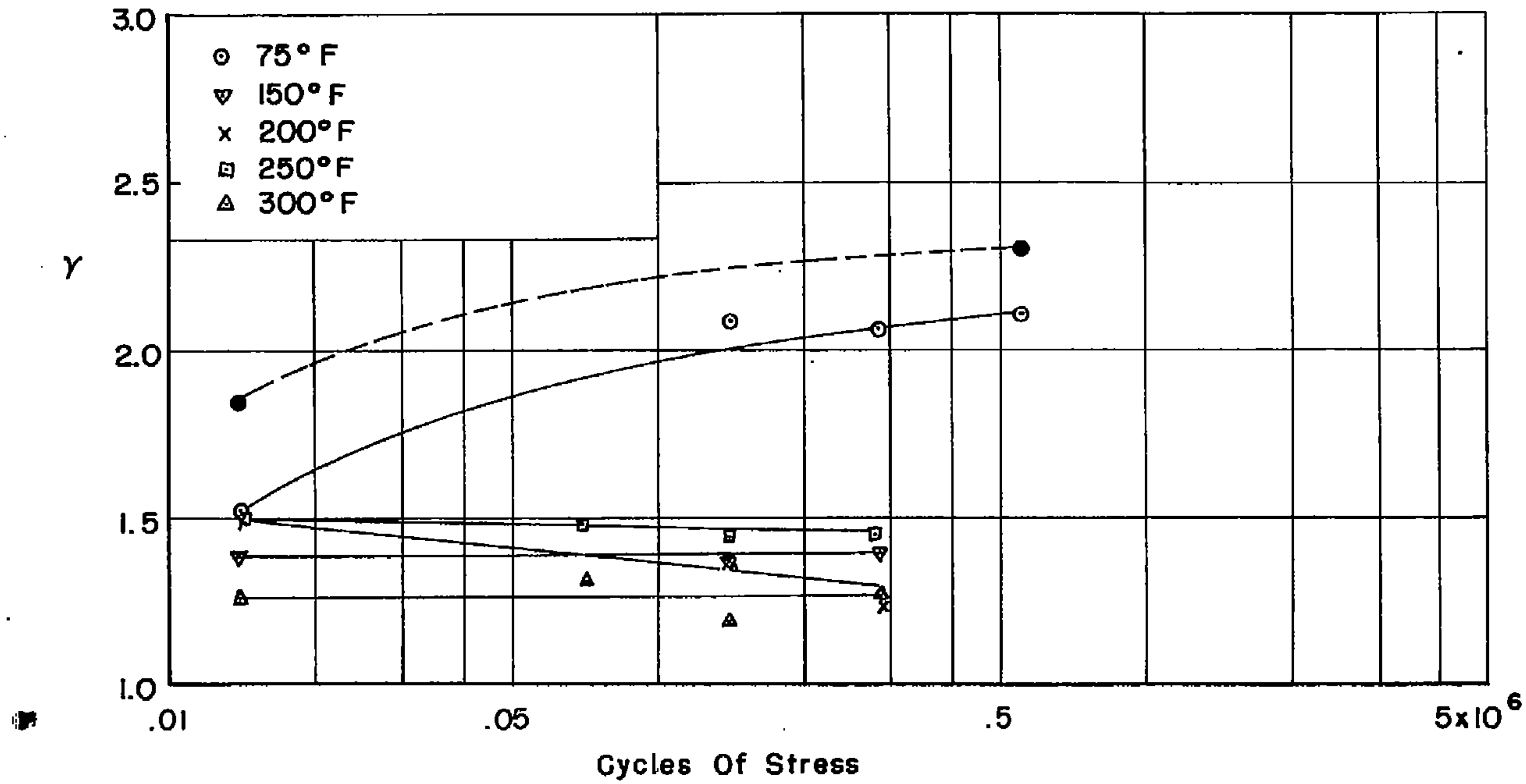
(b) Stress, 5,000 psi.

Figure 8.- Continued.



(c) Stress, 6,000 psi.

Figure 8.- Continued.



(d) Stress, 7,000 psi.

Figure 8.- Concluded.

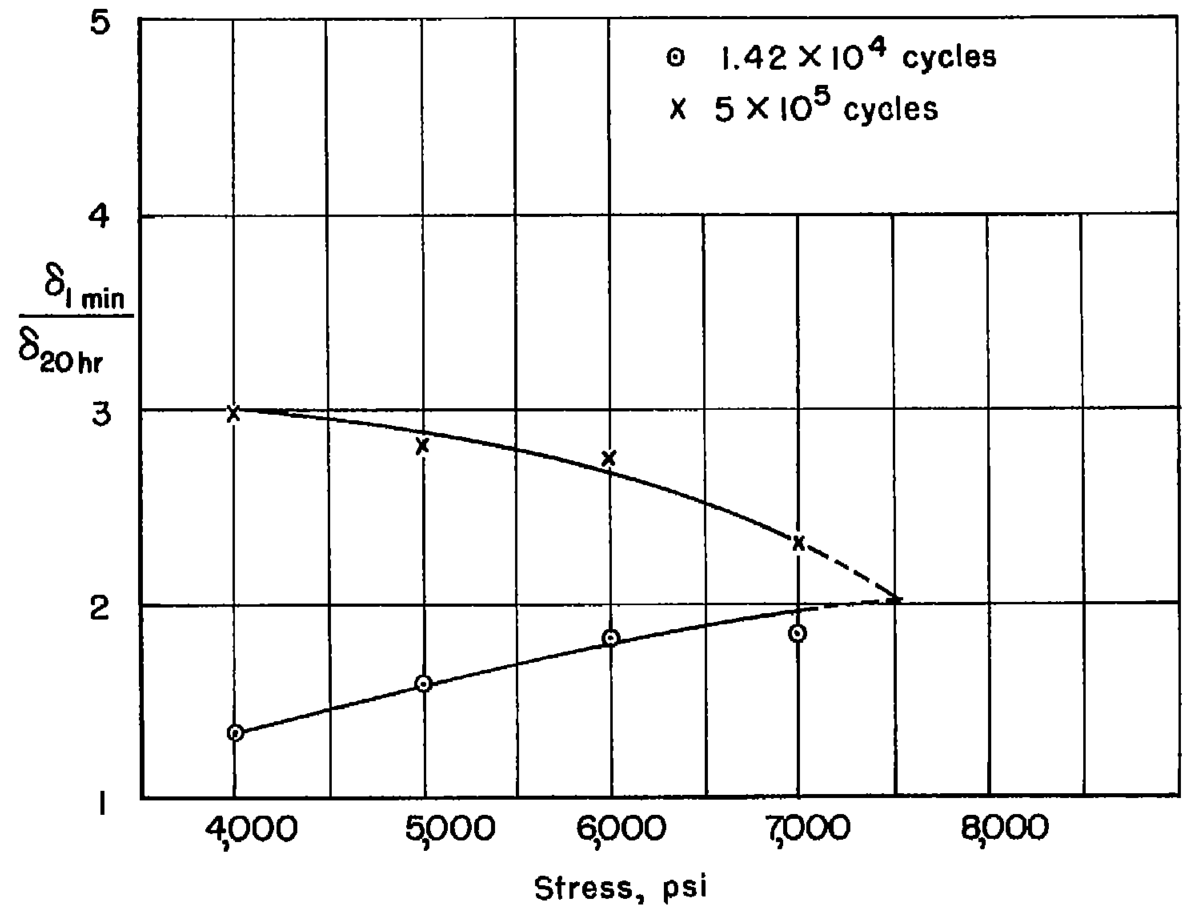


Figure 9.- Relation between recovery factor and short and long stressing histories.

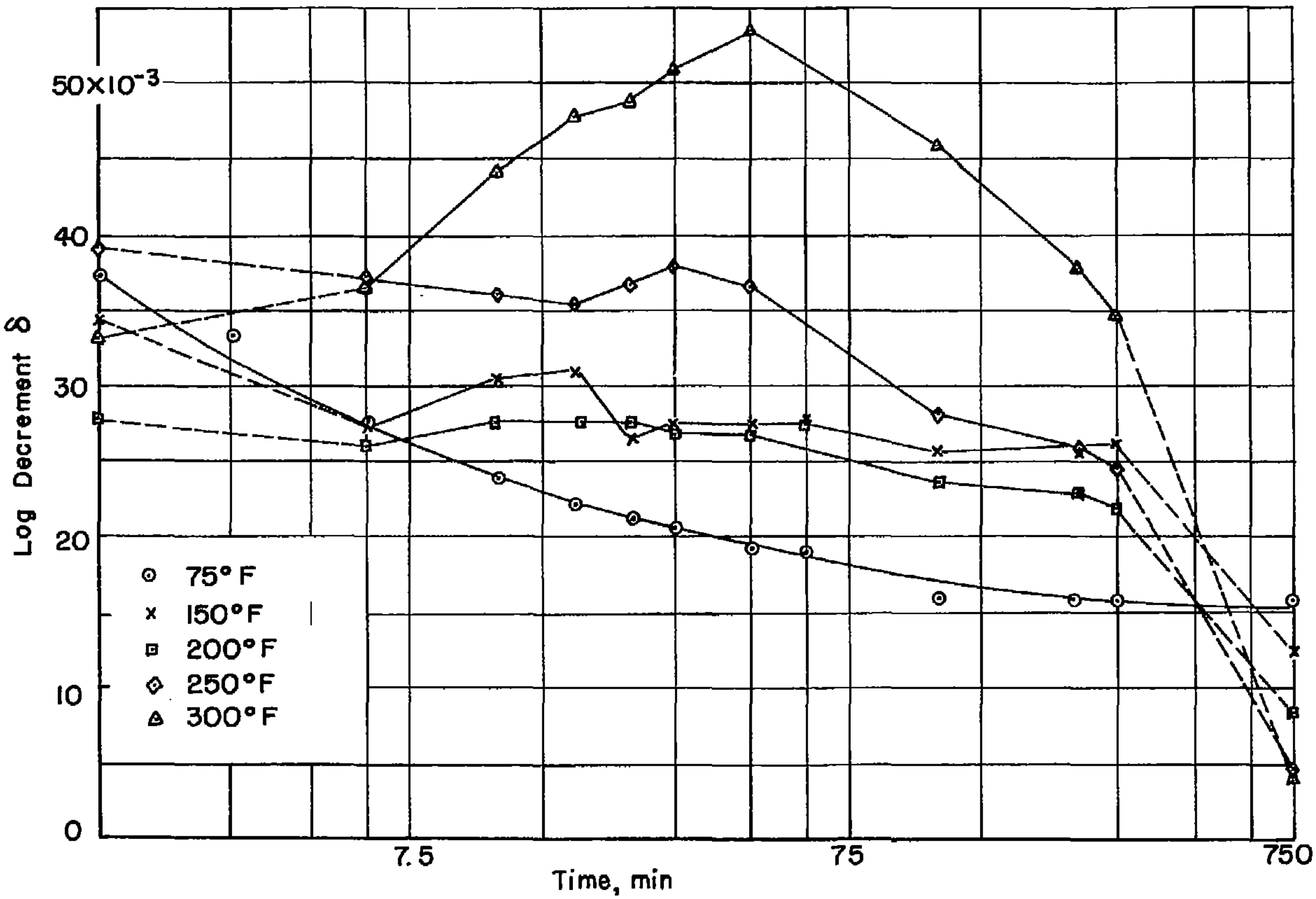


Figure 10.- Variation of recovery for testing at 75° F with rest periods at elevated temperatures. Broken lines on right indicate cooling processes and damping after cooling. End points on right indicate value of room-temperature damping.