THE INFLUENCE OF STRAIN RATE AND STRAIN AGEING ON THE FLOW STRESS OF COMMERCIALLY PURE ALUMINUM

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FACULTY OF MECHANICAL ENGINEERING
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The Influence of Strain Rate and Strain Ageing on the Flow Stress of Commercially Pure Aluminum.

by

A. Rosen[1] and S. R. Bodner[2]

Abstract

Tensile tests on annealed, commercially pure aluminum specimens were performed in a hard machine at different temperatures for a number of constant strain rates. The results show that a range of strain, strain rate, and temperature exists for which the flow stress decreases with strain rate. This material property, when coupled with suitable mechanical conditions, is believed to be cause of discontinuous, repeated yielding. Various experiments, especially relaxation and reloading tests, show that the decrease of flow stress with strain rate can be directly attributed to rapid strain ageing due to impurity diffusion. Other experiments indicate the same effect for different loading histories. An analytical representation of the strain rate and strain ageing of the material is developed which predicts results in very good agreement with the experiments.

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1 Senior Lecturer, Faculty of Mechanical Engineering, Technion - Israel Institute of Technology.

2 Professor of Mechanics, Faculty of Mechanical Engineering, Technion - Israel Institute of Technology.

Introduction

The phenomenon of discontinuous, repeated yielding of various metals is fairly well known, Portevin - Le Chatelier Effect (1923), but has recently received renewed interest and attention. This seems due to the general advance of knowledge of microscopic material behavior and the relevance of discontinuous yielding to basic questions in plasticity theory. A large amount of recent experimental work on this problem has been based on commercially pure and low alloy aluminums which exhibit pronounced "staircase" yielding under slowly applied dead weight loading at room temperature, McReynolds (1949), Krupnik and Ford (1953), Phillips (1953), Dillon (1963 a, b), Bell and Stein (1962). Although the presence of impurities has been shown to be essential to the occurrence of discontinuous yielding - it does not occur in the usual form for very pure aluminum - it does not seem that the basic mechanism is well understood. Most proposed explanations describe the macroscopic results entirely by microscopic mechanisms, e.g. the breaking away of dislocation lines from pinning points.

The primary aim of the present pair of papers is to show that the phenomenon of repeated yielding is due not only to certain metallurgical properties but is critically dependent upon the mechanics of the test procedure. The responsible metallurgical mechanism is considered to be a negative slope in the flow stress-strain rate relation. For commercially pure aluminum this appears to be due to strain ageing by impurity diffusion to dislocations. This paper is concerned with the metallurgical aspects of the problem. The accompanying paper describes the mechanics of the test procedures that lead to discontinuous, repeated yielding.

An experimental program was performed to investigate the effects of strain rate and strain ageing on the flow stress of commercially pure aluminum. Experiments were conducted in a hard testing machine over a wide temperature range and included tension tests at constant straining rates, tests with changing strain rates, and relaxation and reloading tests. The results showed that a decrease of flow stress with strain rate (for the same strain level) occurs over a range of temperature and strain rate and could be directly attributed to the strain ageing effect of impurities. That is, at relatively low strain rates and strain levels, impurities can diffuse to dislocations and impede their motion thereby resulting in a time dependent increase of flow stress. Due to the time hardening factor, slower strain rates would require higher flow stresses at the same strain level.

An analytical model for the flow stress is developed which incorporates the effects of strain rate and strain ageing. Predictions based on the model are found to be in very good agreement with the experimental results. The analysis indicates that a decrease of flow stress with strain rate should occur at low strain levels and in a range of relatively low strain rates.

Experimental Procedure and Results

The main objective of the test program was to determine the strain rate dependence of the flow stress of commercially pure aluminum at various strain levels and temperatures. The procedure was to subject specimens to constant tensile straining rates in a relatively hard machine, a 5 ton capacity Instron. The maximum load on the specimens was only 200 Kg. The specimen dimensions are shown in Fig. 1a. The initial effective gage length, L_0, for the calculation of strains from the crosshead displacement was taken to be 54 mm.

The composition of the specimens is given in Table 1. It is noted that the impurity content (0.6%) is somewhat lower than that of materials used in recent investigations on discontinuous yielding (0.8%). The specimens were all prepared from a single 2mm. thick cold rolled sheet. They were all cut with the tensile axis perpendicular to the rolling direction in order to minimize the effect of rolling. The specimens were sealed together in aluminum foil and annealed for 2 hours at 600°C and subsequently furnace cooled. This heat treatment is similar to that used in the related investigations. Microscopic examination of the specimens after annealing showed an almost equiaxial grain structure with slightly remaining traces of the cold rolling direction. The average grain diameter was found to be 0.06 mm.

In order to maintain a uniform temperature bath around the specimen, the apparatus was modified as shown in Fig. 1b. The load bar and cage were very rigid (compared to the specimen) and did not significantly diminish the rigidity of the loading system. The Instron machine records load against time (crosshead displacement) by means of a load cell in series with the specimen. A 5 ton capacity load cell was used for greatest rigidity of the system. The load could be measured to an accuracy of $\pm$ 0.1 Kg., but a sensitivity of $\pm$ 0.5 Kg. was used in most of the tests. The temperature could be kept constant and uniform to within $\pm 1^{\circ}C$. A number of specimens were instrumented with strain gages in order that the actual strains of the specimens could be directly observed. The strain time history for these tests was recorded on a rapid response potentiometer recorder.

Since the material does not exhibit a sharp yield point, the proportional limit was taken as the datum for the measurement of plastic strains. The proportional limit varied somewhat with strain rate but was about 1.4 Kg./sq.mm. for $23^{\circ}C$ which is about the same as that observed for the specimen materials used in the related experiments. Since strains up to 20% were measured, it was considered desirable to account for the change in cross section when computing stress values from the force, (i.e. $\sigma = F(1+\varepsilon)/A_0$). In addition, the natural strain expression ($\log_e L/L_0$) was used for the strain representation.

The controlled straining tests were intended to establish flow stress-strain rate relationships. These can be obtained from stress strain curves obtained at constant strain rates by replotting the curves as stress against strain rate at constant strain levels. This procedure is commonly used and was used in the present paper. It is noted, however, that these results should not be interpreted as completely general relationships but rather the consequence of a particular test method. The replotting and representation as a general relationship presupposes the existence of an equation of state of the form

$$\sigma = \sigma\ (\varepsilon, \dot{\varepsilon}, T) \tag{1}$$

for the stress in the inelastic region. It is well established that such an equation does not generally apply since the flow stress depends upon the complete loading history. For not too dissimilar loading histories, however, Eq. (1) may be a reasonable approximation. A dead weight loading test that exhibits discontinuous yielding has a varying strain rate history, so part of the test program was directed toward examining strain rate dependence under conditions of changing strain rate. Some preliminary work on this aspect has been previously reported, Rosen and Bodner (1966).

Controlled straining tests were performed at strain rates varying from 1.54×10^{-6} sec.$^{-1}$ to 1.54×10^{-2} sec.$^{-1}$ at temperatures of -10^{o}, 0^{o}, 23^{o} and 40^{o}C. Some experiments were also carried out at -185^{o}C (liquid air), but since the results were very similar to those at -10^{o}C, they are not reported in detail. Most of the experiments were repeated at least once in order to ensure reproducibility and to reduce the scatter of the test data.

The load time records from the Instron were generally smooth. Some slight irregularities appeared in the records at 0°, 23° and 40°C at the lower strain rates. The strain gage records at 0°C and 23°C also showed some irregularities at low strain levels ($< 1\%$), but very much less then that observed in slowly applied dead weight loading tests. The sharp "steps" commonly observed in those tests were not present in the controlled straining tests. Smooth stress strain could therefore be plotted from the present test results.

Stress strain curves obtained from constant strain rate tests at various temperatures are shown on logarithmic scales in Figs. 2 - 5. These scales were used in order to emphasize the low strain region. The strain value is the plastic strain based on the proportional limit strain as the datum. The curves are the average of a number of experiments. The experimental points are not shown on the diagram due to the large number involved. The most interesting result is the overlapping of some of the stress strain curves. The curves for 0°, 23° and 40°C show that in the low strain region the flow stress for the same strain increases with decreasing strain rate. The experiments at -10° and -185°C, however, exhibited an increase of flow stress with strain rate over the complete testing range.

As discussed previously, the stress strain curves can be replotted as flow stress against strain rate for different strain levels, Figs. 6 - 9. The experimental points can be well fitted by straight lines. The lines for the -10° and $-185^\circ C$ tests have positive slopes. The 0°, 23° and $40^\circ C$ test results can be represented by two straight lines of opposite slope, i.e. at low strain rates the flow stress decreases with strain rate. The intersection (lowest) point of the lines will be referred to as the "critical strain rate". Since the lines are almost horizontal, a certain amount of error exists in the location of the critical strain rate. However, the experimental results clearly show the existence of such a point.

Since the results show that a region of strain and strain rate exists for which the flow stress decreases with increasing strain rate, it is of interest to examine the corresponding temperature dependence of the flow stress. Such a plot, for a constant strain and strain rate, is shown in Fig. 10. The flow stress is seen to increase with temperature contrary to normal behavior. The relation of this effect to the occurrence of repeated yield points has been noted by Friedel (1964), p. 414.

The negative slope of the flow stress-strain rate relation was believed to be due to strain ageing of the material due to impurity diffusion to dislocations. This would lead to a time dependent increase of flow stress so that tests over longer times to reach the same strain level (i.e. slower strain rates) would show higher stresses than shorter time tests (higher strain rates). To explore this interpretation, a series of relaxation and re-straining experiments were performed. The specimens were strained at a certain rate (1.54×10^{-4} sec.$^{-1}$) to some deformation value, allowed to relax at constant strain for a time interval, after which straining was resumed. The procedure was repeated at various strain levels. Experiments were performed for relaxation times of 12, 30, 60, 120 and 300 seconds at temperatures of -185^{o}, 23^{o}, 57^{o} and 83^{o}C.

Fig. 11 is a schematic representation of a relaxation experiment at 23^{o}C for a strain rate of 1.54×10^{-4} sec.$^{-1}$. When straining is resumed after a relaxation interval (at constant strain), the material exhibits a sharp yield point. This result has been discussed by Cottrell (1963) to be a consequence of impurity hardening. For this test, both the upper and lower yield points are higher than the flow stress before relaxation which can be attributed to strain ageing. Fig. 12 is a plot of $\Delta\sigma$, the difference in flow stress before and after relaxation, as a function of strain for different relaxation times. The stress difference $\Delta\sigma$ is based on the lower yield stress (the flow stress) upon restraining (shown in Fig. 11). From Fig. 12, ageing appears to be limited to low strain levels and is most pronounced at about 1% strain. The time dependence of hardening at 1% strain is plotted in Fig. 13 to a logarithmic scale. Ageing was also observed in the relaxation tests performed at 57^{o} and 83^{o}C, Fig. 14. The amount of age hardening tends to decrease at the higher temperatures. A few ageing tests were performed at 23^{o}C in which the load rather than the strain was kept constant for some time interval. A sharp yield point and an increase of flow stress was again observed when straining was resumed. The increase of flow stress appeared for higher strain levels than in the constant strain tests.

In order to study the role of impurity diffusion on the ageing process, relaxation tests on commercially pure aluminum were conducted at $-185^{\circ}C$, where the diffusion effect would be very small. In addition, relaxation tests were performed on very pure (99.99%) aluminum at $23^{\circ}C$. A schematic representation of the results for both tests is shown in Fig. 15. The sharp yield upon restraining was not present nor was any increase of flow stress observed after long relaxation intervals (300 seconds). The relaxation in both of these tests is probably due to the re-arrangement of dislocations. These tests provide further support to impurity diffusion as the principal strain ageing mechanism.

The other auxiliary experiments were directed toward examining the effect of loading history on the flow stress-strain rate relationship. In particular, it was desirable to determine if the decrease of flow stress with strain rate occurred for histories other than constant strain rate. The specimens in dead weight loading tests for which "steps" are observed experience large variations in strain rate.

One set of tests was performed in which the rate of loading was rapidly changed by factors of 10 or 100 by pressing the appropriate crosshead speed control button or by switching gears of the Instron. These test results were difficult to interprete since the machine response was not entirely instantaneous and smooth and thereby interferred with the load recording. Nevertheless a definite decrease of flow stress with strain rate was observed at low strain levels, Rosen and Bodner (1966). More accurate data was obtained by unloading the specimen and reloading at another strain rate. Here again similar results were obtained. The stress differences due to reloading at a lower strain rate are shown in Fig. 16 compared to those obtained from constant strain rate tests. It appears that the flow stress-strain rate relationships shown in Figs. 6 - 9 would be reasonably representative for arbitrary loading histories that allow time for the ageing effect. It would seem reasonable, therefore, that these relationships could be used as a basis for the study of the dead weight loading of commercially pure aluminum.

Analytical Representation and Discussion

The principal features of the behavior of commercially pure aluminum in the region of interest are amenable to analytical representation based upon the theory of dislocations. The two main characteristics of the material would be the increase of flow stress with strain rate, which is controlled by the thermally activated dislocation intersection mechanism, and the time dependent increase of flow stress due to impurity diffusion and interaction with dislocation motion. The applicability of the intersection mechanism to aluminum has been shown by Mitra, Osborne and Dorn (1961), Mitra and Dorn (1963), and Lindholm (1964).

For the thermally activated mechanism of dislocation intersection, the strain rate $\dot{\varepsilon}$ is related to the flow stress, σ, [as shown by Seeger (1957)],

$$\dot{\varepsilon} = NAb\nu \exp\left[-U(\sigma)/kT\right] \qquad (2)$$

In this expression N is the number of intersections of the dislocation per unit volume, A is the area swept out by the dislocation line after intersection, b is the Burgers vector, ν is the dislocation frequency, $U(\sigma)$ is the energy of thermal activation which is necessary for a dislocation to cut through a forest or overcome certain obstacles, k is the Boltzman constant and T is the absolute temperature.

An expression for the stress dependence of the activation energy, proposed by Seeger (1957), is

$$U = U_0 - v(\sigma - \sigma_G) \tag{3}$$

where U_0 is the total energy required for intersection, v is the activation volume, σ is the applied stress and σ_G is the overall back stress opposing dislocation motion.

Substituting Eq. (3) into Eq. (2) and solving for σ leads to

$$\sigma = (1/v) \left[U_0 + kT \ln(\mathring{\varepsilon}/NAb\nu) \right] + \sigma_G \tag{4}$$

Eq. (4) shows that σ is an increasing function of $\mathring{\varepsilon}$ provided that the other factors of the equation are constant at constant temperatures. This is probably the case for pure metals, for which case σ_G is both athermal and independent of strain rate. However, in the case of commercially pure aluminum σ_G depends on the strain rate since the material age hardens during straining. The effect of strain ageing is greater as the rate of straining is slower or the testing time is longer. This was demonstrated by the relaxation experiments. The back stress σ_G may therefore be written as

$$\sigma_G = \sigma'_G + \sigma_A \tag{5}$$

where σ'_G is the time independent part of the overall back stress and σ_A is the back stress due to strain ageing. The stress required to force a dislocation through the matrix between particles of precipitate a mean distance λ apart on the slip plane can be expressed as, [Mitchell, Mitra and Dorn (1963)],

$$\sigma_p = (Gb/\lambda) + \sigma'_p \tag{6}$$

where σ'_p is the long range back stress and G is the shear modulus. The term σ'_p is time independent and is part of σ'_G, so that the stress due to strain ageing is

$$\sigma_A = Gb/\lambda \tag{6a}$$

Friedel (1964), p. 408, relates the mean distance between precipitates, i.e. the average dislocation length, to the concentration of impurity atoms having arrived on the dislocation, $c - c_0$,

$$\frac{1}{\lambda} = \frac{1}{\lambda_0} + \frac{c - c_0}{b} \tag{7}$$

The time dependence of this concentration of impurity atoms is given by Friedel (1964), Eq. (16.2):

$$c - c_0 = \frac{\pi c_0}{b^2} \left[\frac{n(n+2)\ D\ |W_M|\ b^n t}{kT} \right]^{\frac{2}{n+2}} \tag{8}$$

where c_0 is the concentration of the solute atoms, D is the diffusion coefficient of the impurities, W_M is the binding energy (since W_M is negative, its absolute value is taken in the above equation) and n is a constant. Friedel states that $n=1$ for size effects and $n=2$ for differences in the elastic constants, i.e. in the case of substitutional solid solution, $n= 2$. For $n=2$ the flow stress due to ageing would increase with the square root of time. The initial slope of the line in Fig. 13 ($\log \Delta\sigma$ vs. $\log t$) is 0.43, which is in good agreement with the above assumption. Substituting Eqs. (8) and (7) into Eq. (6a) with $n=2$, the stress σ_A becomes

$$\sigma_A = Gb \left[\frac{1}{\lambda_0} + \frac{\pi\, c_0}{b^2} \left(\frac{8D\,|W_M|\,t}{kT} \right)^{1/2} \right] \tag{9}$$

Substituting Eqs. (9) and (5) into Eq. (4), the expression for the stress due to both strain rate and strain ageing becomes

$$\sigma = (1/v)\left[U_0 + kT \ln(\dot{\varepsilon}/NAb\nu) \right] + \sigma'_G + Gb \left[\frac{1}{\lambda_0} + \frac{\pi c_0}{b^2} \left(\frac{8D|W_M|t}{kT} \right)^{1/2} \right] \tag{10}$$

For constant strain rate conditions, the time can be expressed as $t = \varepsilon/\dot{\varepsilon}$, so that Eq. (10) can be written as a function of $\dot{\varepsilon}$,

$$\sigma = C + A \ln \dot{\varepsilon} + B\dot{\varepsilon}^{-1/2} \tag{11}$$

where

$$C = (1/v)\left[U_0 - kT \ln(NAb\nu) \right] + \sigma'_G + (Gb/\lambda_0) \tag{12}$$

$$A = kT/v \tag{13}$$

$$B = \frac{G\pi\, c_0}{b} \left(\frac{8D|W_M|\,\varepsilon}{kT} \right)^{1/2} \tag{14}$$

Eq. (11) represents the strain rate dependence of the flow stress of a metal which undergoes age hardening during straining. It indicates that the flow stress is not a monotonically increasing function of strain rate, but at low strain rates the flow stress decreases with increasing strain rate. This is what was found experimentally for commercially pure aluminum as shown in Figs. 6-9.

Eq. (11) also shows that for very high strain rates the flow stress should increase with increasing strain rate since $\dot{\varepsilon}^{-1/2}$ is small compered to $\ln \dot{\varepsilon}$. The reason for this behavior is that at high strain rates the impurity diffusion is too slow to interact with dislocation motion. In other words, strain ageing would have very little hardening effect at high rates of strain. Although according to Eq. (11) the flow stress should continually increase with $\ln \dot{\varepsilon}$, this relation does not seem to hold for very fast rates of strain apparently due to the operation of other mechanisms. High rate of strain tests on commercially pure aluminum indicate an asymptotic stress strain curve, i.e. the flow stresses become essentially independent of strain rate at very high rates of strain, Bell (1960).

On the other hand, when strain rates are extremely slow, the stress should decrease with increasing strain rate according to Eq. (11). This would be the case if the $\dot{\varepsilon}^{-1/2}$ law would hold for any time. Fig. 13 shows, however, that for long times the ageing effect tends to stabilize toward a constant $\Delta\sigma$. A physical interpretation is that the impurities would move along with the dislocations (microcreep). In this case the dominant factor of Eq. (11) is again $\ln \dot{\varepsilon}$, i.e. the flow stress would increase with strain rate at extremely slow rates of strain.

Eq. (11) should, therefore, be expected to be a reasonable representation only within a range of moderately low strain rates.

Although Eq. (11) includes the temperature, it does not show that strain ageing would occur only within a limited temperature range. At low temperatures the diffusion rate is too slow to interfere with dislocation motion, whereas at high temperatures the migration of the impurities with the dislocations becomes easier so that the interference drag becomes small.

Since Eq. (11) exhibits both negative and positive slopes for low and high strain rates respectively, it should demonstrate a minimum flow stress at the "critical strain rate" for a given strain. The condition for a minimum is obtained by setting

$$\frac{d\sigma}{d\dot{\varepsilon}} = \frac{A}{\dot{\varepsilon}} - \frac{1}{2} B\dot{\varepsilon}^{-3/2} = 0 \tag{15}$$

which leads to

$$\dot{\varepsilon}_{cr} = \frac{B^2}{4A^2} = \frac{2(G\pi\ c_0 v)^2\ D\ |W_M|\ \varepsilon}{k^3\ T^3\ b^2} \tag{16}$$

Computation of the second derivative shows that Eq. (16) is, indeed, the minimum point. Most all of the terms in Eq. (16) are known constants or can be reliably estimated. For the specimen material (99.4% aluminum) at 23°C and for 1% strain, the terms have the following values:

$G = 2.37 \times 10^{11}$ dynes/ cm^2

$c_0 = 6 \times 10^{-3}$ (based on percentage of impurities)

$D_0 = 0.9\ cm^2/sec.$, Smithells (1955))The values of D_0 and

)Q are for dilute solution

$Q = 30.55 \times 10^3$ cal/gram mole, Smithells (1955))of Si in Al. The con-

)stants for Fe or Cu in

)Al are similar.

$D = D_0 \exp(-Q/RT) = 0.35 \times 10^{-22} cm^2/sec.$

$|W_M| = 0.48 \times 10^{-12}$ erg, Friedel (1964), p. 413

$b = 2.86 \times 10^{-8}$ cm., Friedel (1964), p. 454

$k = 1.38 \times 10^{-16}$ erg/^oK

$T = 296^o$ K

$\varepsilon = 10^{-2}$

The only term whose value is somewhat indefinite is the activation volume. The activation volume can be expressed in the form $v = ab^3$ where "a" is a material constant. A value of "a" of about 100 for polycrystalline, commercially pure aluminum at low strain levels is suggested from the work of Nolder and Dorn (1962) and Nunes, Rosen and Dorn (1965).

Substitution of the numerical values into Eq. (16) leads to $\dot{\varepsilon}_{cr} \cong 7 \times 10^{-4}$ sec.$^{-1}$. This is in very good agreement with the value of 1.54×10^{-4} for $\dot{\varepsilon}_{cr}$ shown in Fig. 8. This good agreement may be somewhat fortuitous, but the use of the maximum permissible variations of the numerical values would not alter $\dot{\varepsilon}_{cr}$ by more than an order of magnitude.

The preceding analysis shows that the inclusion of strain ageing in the flow stress-strain rate relation leads to a negative slope over part of the strain rate range. However, the analysis is too simplified to permit detailed studies of the role of strain and temperature on the limits of the negative slope and on the value of the "critical strain rate". Such analytical studies should also be coupled with a more extensive experimental investigation.

Conclusions

Constant strain rate tensile tests on specimens of annealed, commercially pure aluminum show that the flow stress decreases with strain rate at low strain rates ($< 10^{-3}$ sec.$^{-1}$) and strains ($< 5\%$), and at temperatures about room temperature. This is shown to be a consequence of rapid strain ageing due to the diffusion of impurities to dislocations. Relaxation and reloading experiments at $23^{o}C$ show that hardening due to strain ageing depends upon time to the 0.43 power and is limited to strains less than 5%. An analytical representation for the strain rate and strain ageing dependence of the flow stress is developed. The derived equation shows the flow stress would decrease with strain rate at low strain rates. The predicted strain rate for the minimum flow stress is found to be in good agreement with the experimental results.

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Table 1

Impurity Content of Commercially

Pure Aluminum

Cu	-	0.002%
Si	-	0.15%
Fe	-	0.35%
Mg	-	0.05%
Ni	less then	0.02%
Mn	" "	0.02%
Total Impurity Content		≈ 0.60%

List of Captions for Illustrations

Fig.1a Specimen dimensions.

Fig.1b Temperature controlled, constant strain rate, tensile testing machine.

Fig. 2 Stress strain curves for annealed commercially pure aluminum at $-10^{\circ}C$ for various strain rates.

Fig. 3 Stress strain curves for annealed commercially pure aluminum at $0^{\circ}C$ for various strain rates.

Fig. 4 Stress strain curves for annealed commercially pure aluminum at $23^{\circ}C$ for various strain rates.

Fig. 5 Stress strain curves for annealed commercially pure aluminum at $40^{\circ}C$ for various strain rates.

Fig. 6 Flow stress at various strains versus strain rate for annealed commercially pure aluminum at $-10^{\circ}C$.

Fig. 7 Flow stress at various strains versus strain rate for annealed commercially pure aluminum at $0^{\circ}C$.

Fig. 8 Flow stress at various strains versus strain rate for annealed commercially pure aluminum at $23^{\circ}C$.

Fig. 9 Flow stress at various strains versus strain rate for annealed commercially pure aluminum at $40^{\circ}C$.

Fig.10 Dependence of flow stress on temperature for fixed strain and strain rate.

Fig.11 Schematic representation of stress time relation for relaxation tests (constant strain) on commercially pure aluminum at $23^{\circ}C$.

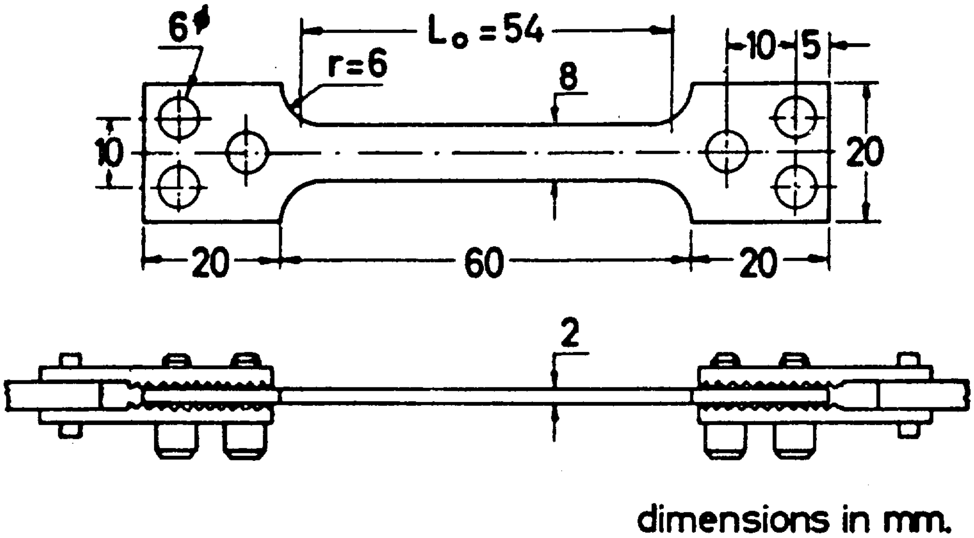
Fig.12 Time and strain dependence of flow stress increment due to strain ageing of commercially pure aluminum at $23^{\circ}C$.

Fig.13 Logarithmic plot of relaxation time dependence of flow stress increment due to strain ageing at 1% strain.

Fig.14 Strain and temperature dependence of flow stress increment due to strain ageing of commercially pure aluminum.

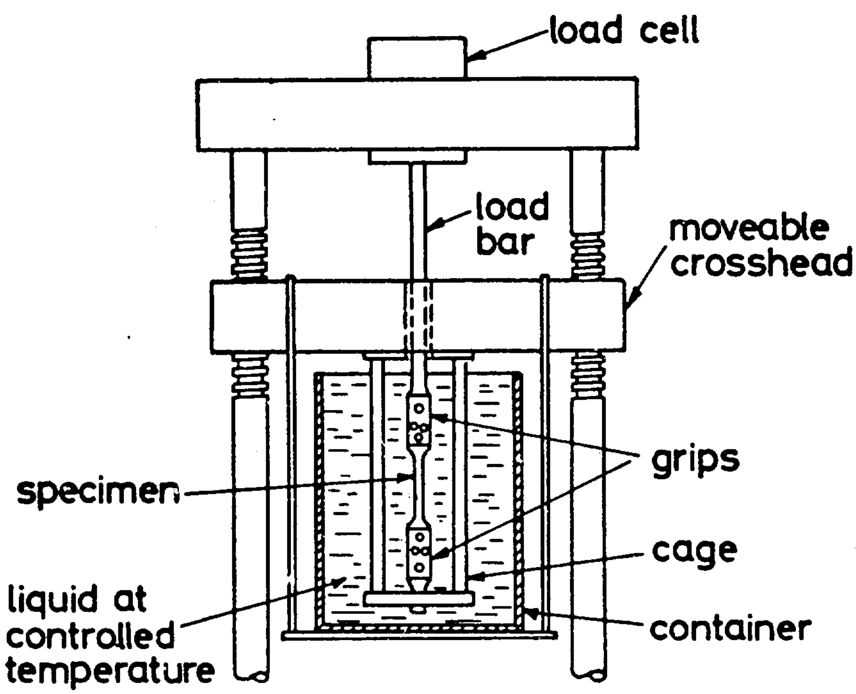
Fig.15 Schematic representation of stress time relation for relaxation tests (constant strain) on commercially pure aluminum at $-185^{\circ}C$, and on very pure aluminum at $23^{\circ}C$.

Fig.16 Flow stress differences due to strain rate for both constant and variable strain rate tests.



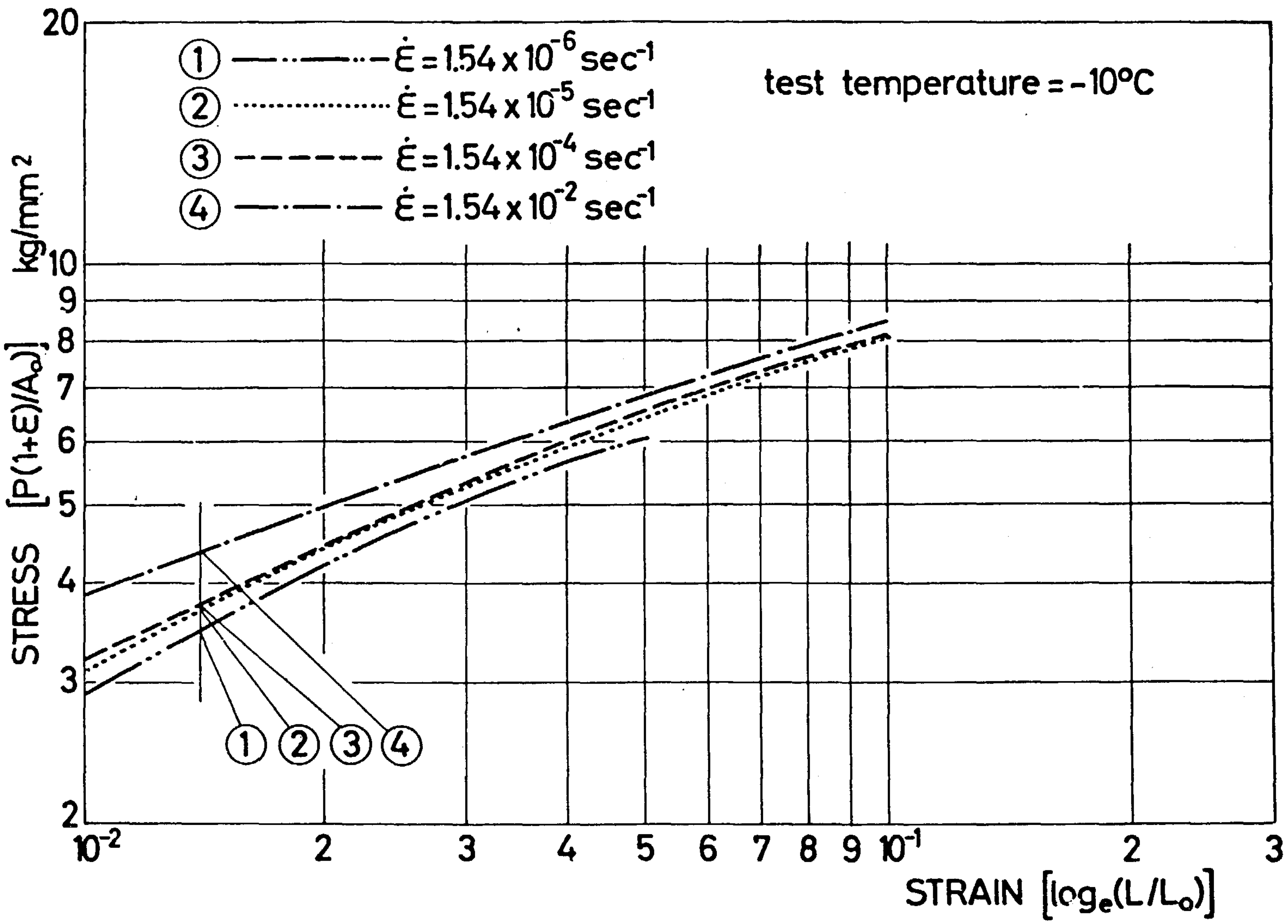
6ϕ
$L_o = 54$
r=6
8
10
5
10
20
20
60
20
2
dimensions in mm.

FIG. 1a



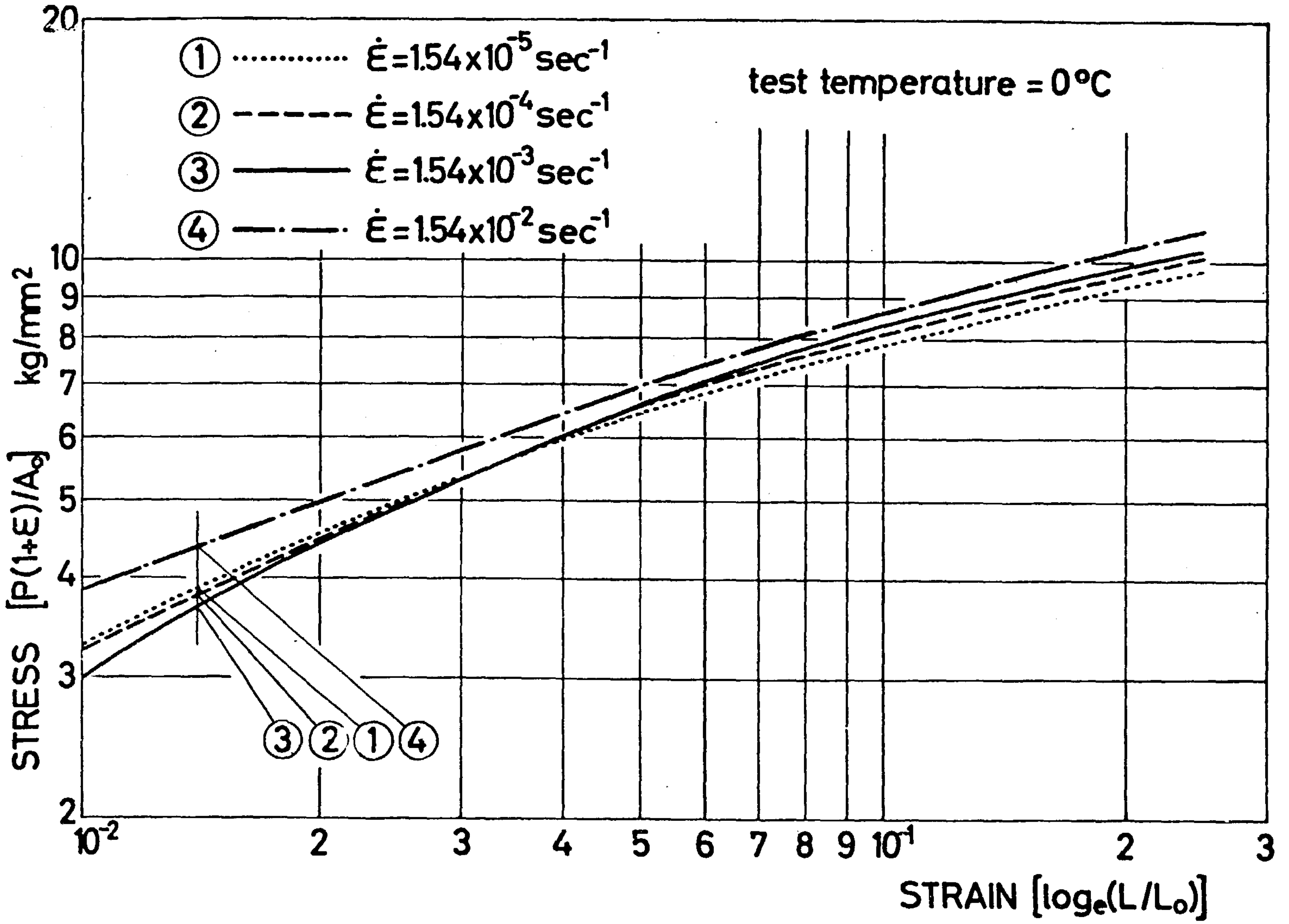
load cell
load bar
moveable crosshead
grips
specimen
cage
liquid at controlled temperature
container

FIG. 1b



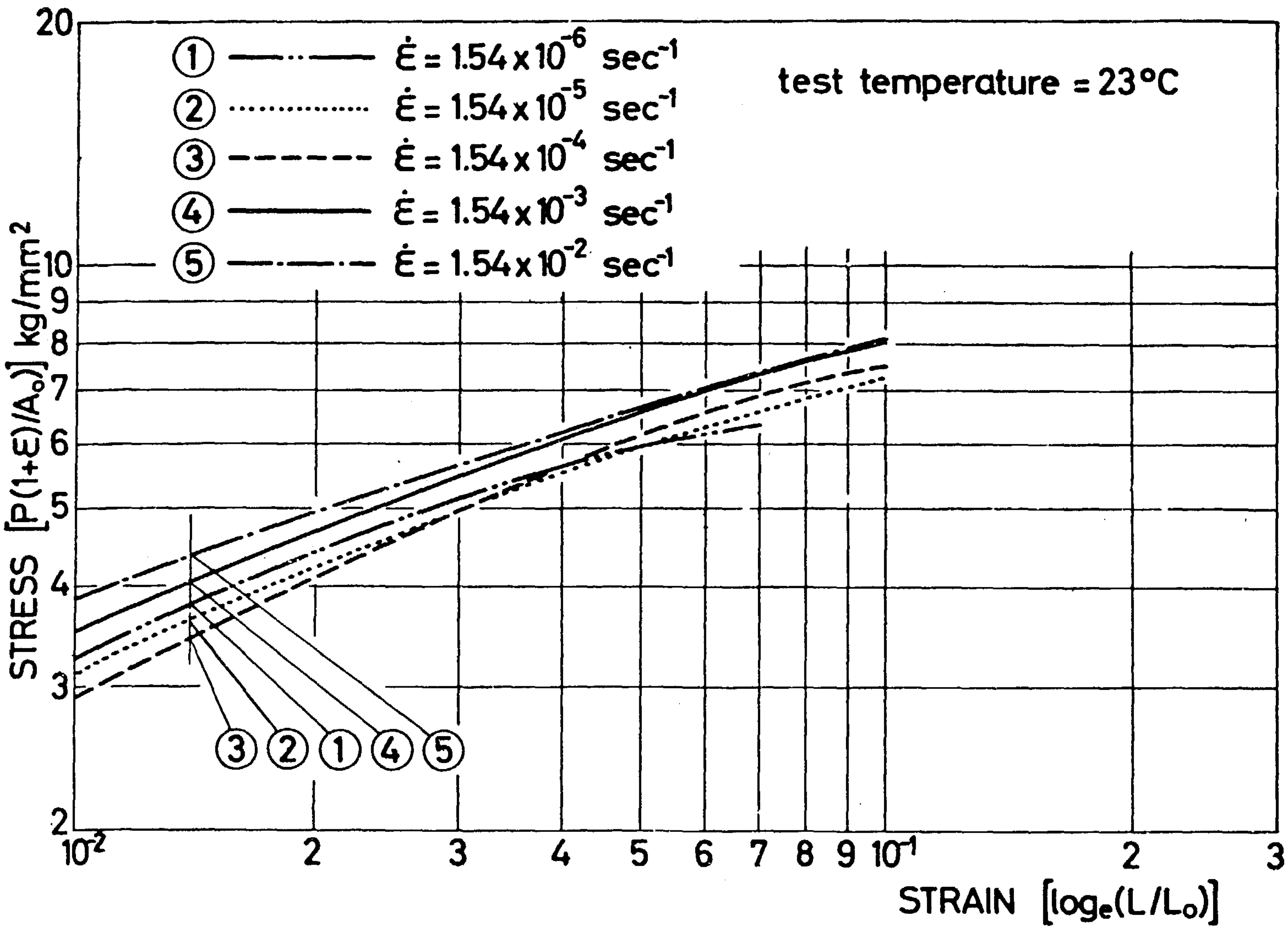
20
10
9
8
7
6
5
4
3
2
STRESS [P(1+E)/A$_o$] kg/mm^2
① $\dot{E}=1.54 \times 10^{-6}$ sec^{-1}
② $\dot{E}=1.54 \times 10^{-5}$ sec^{-1}
③ $\dot{E}=1.54 \times 10^{-4}$ sec^{-1}
④ $\dot{E}=1.54 \times 10^{-2}$ sec^{-1}
test temperature = -10°C
① ② ③ ④
10^{-2}
2
3
4
5
6
7
8
9
10^{-1}
2
3
STRAIN [log$_e$(L/L$_o$)]

FIG. 2



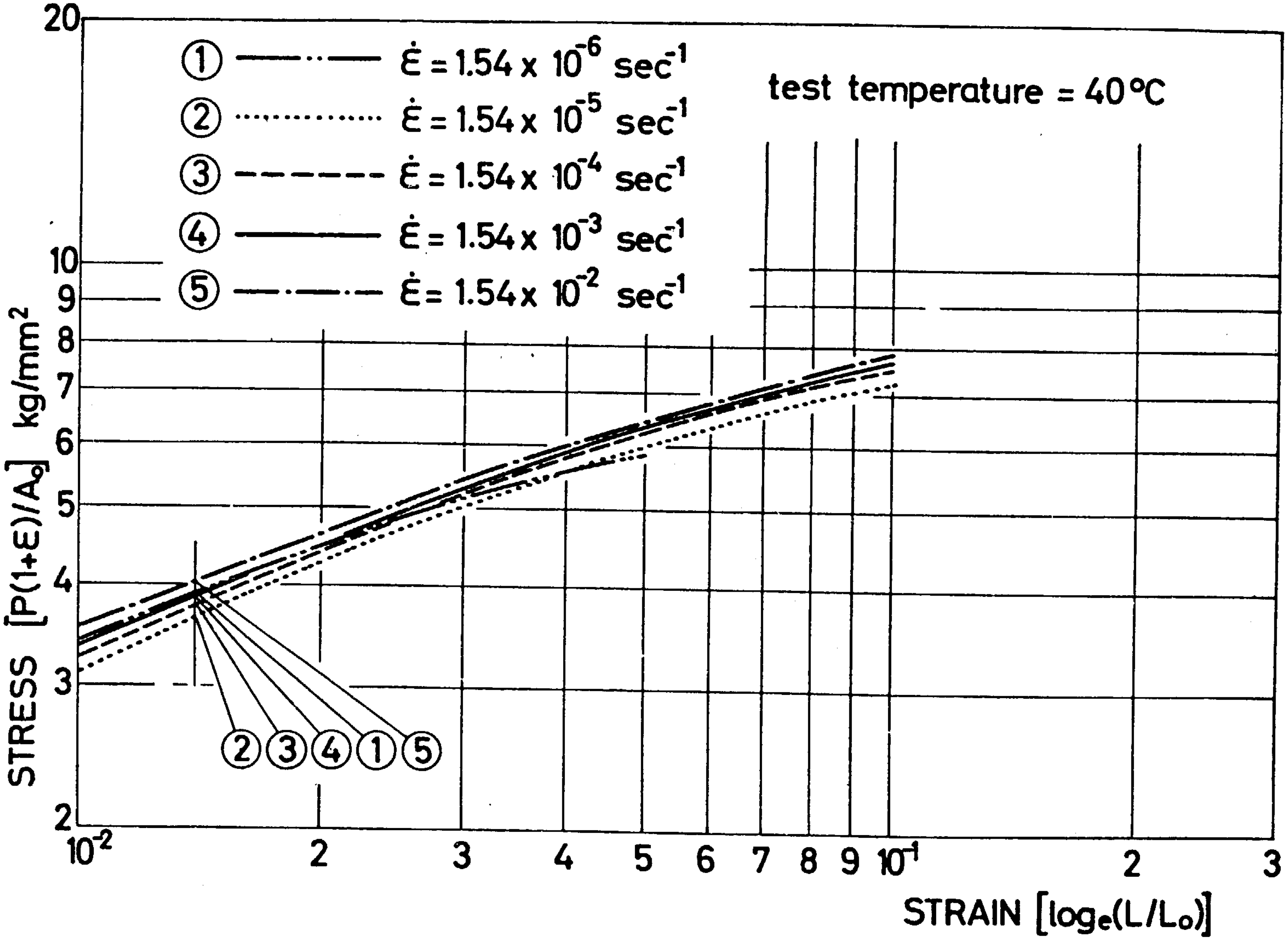
① $\dot{E}=1.54\times10^{-5}sec^{-1}$
② $\dot{E}=1.54\times10^{-4}sec^{-1}$
③ $\dot{E}=1.54\times10^{-3}sec^{-1}$
④ $\dot{E}=1.54\times10^{-2}sec^{-1}$
test temperature = 0°C
STRESS [P(1+E)/A₀] kg/mm²
STRAIN [logₑ(L/L₀)]
20
10
9
8
7
6
5
4
3
2
10^{-2}
2
3
4
5
6
7
8
9
10^{-1}
2
3
③②①④

FIG. 3



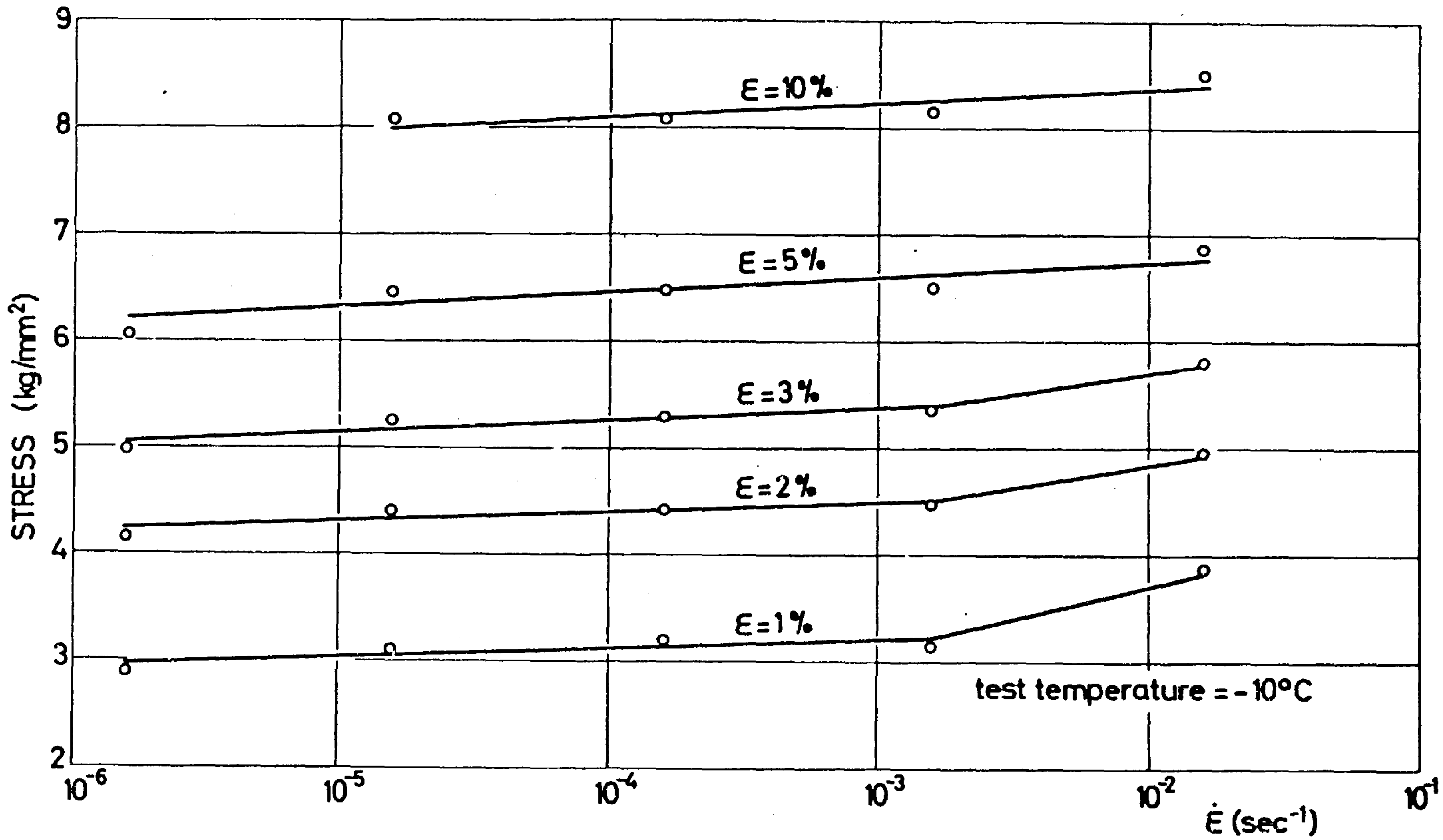
20
10
9
8
7
6
5
4
3
2
STRESS [P(1+E)/A₀)] kg/mm²
① Ė = 1.54 x 10⁻⁶ sec⁻¹
② Ė = 1.54 x 10⁻⁵ sec⁻¹
③ Ė = 1.54 x 10⁻⁴ sec⁻¹
④ Ė = 1.54 x 10⁻³ sec⁻¹
⑤ Ė = 1.54 x 10⁻² sec⁻¹
test temperature = 23°C
③ ② ① ④ ⑤
10⁻²
2
3
4
5
6
7
8
9
10⁻¹
2
3
STRAIN [logₑ(L/L₀)]

FIG. 4



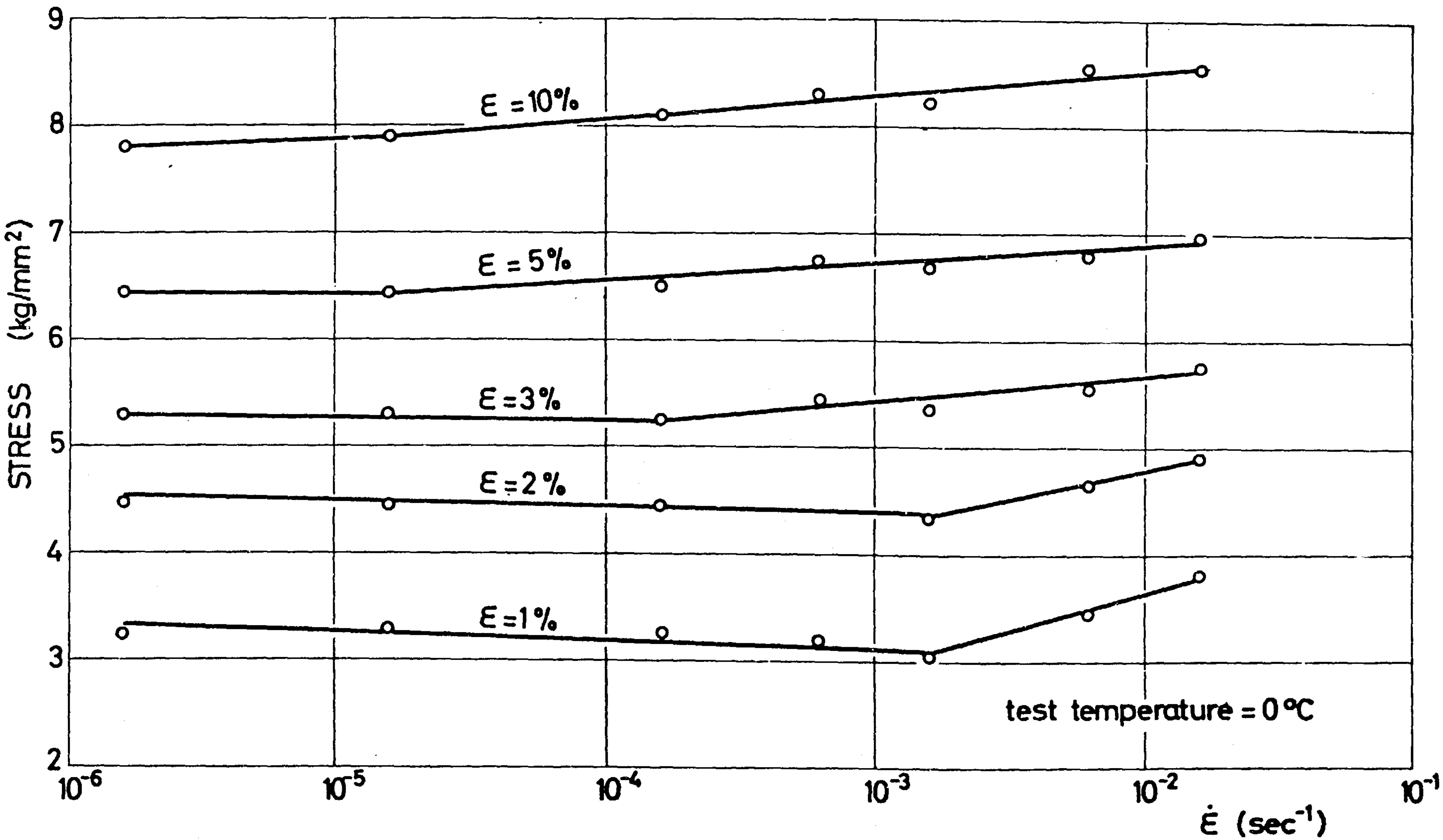
① Ė = 1.54 x 10⁻⁶ sec⁻¹
② Ė = 1.54 x 10⁻⁵ sec⁻¹
③ Ė = 1.54 x 10⁻⁴ sec⁻¹
④ Ė = 1.54 x 10⁻³ sec⁻¹
⑤ Ė = 1.54 x 10⁻² sec⁻¹
test temperature = 40°C
STRESS [P(1+E)/A₀] kg/mm²
STRAIN [logₑ(L/L₀)]
20
10
9
8
7
6
5
4
3
2
10⁻²
2
3
4
5
6
7
8
9
10⁻¹
2
3
② ③ ④ ① ⑤

FIG. 5



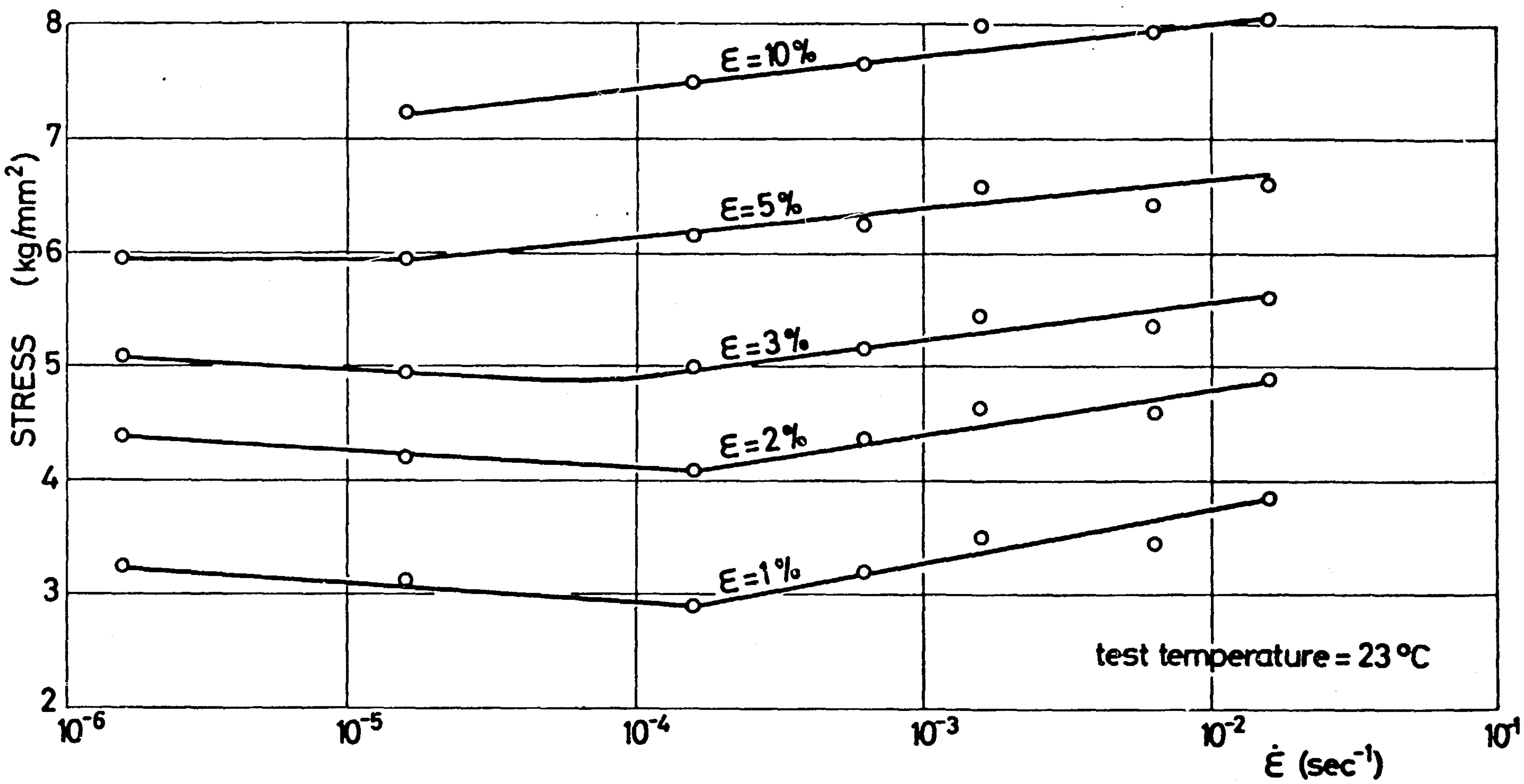
STRESS (kg/mm²)
9
8
7
6
5
4
3
2
ε = 10 %
ε = 5 %
ε = 3 %
ε = 2 %
ε = 1 %
test temperature = -10°C
10⁻⁶
10⁻⁵
10⁻⁴
10⁻³
10⁻²
10⁻¹
ε̇ (sec⁻¹)

FIG. 6



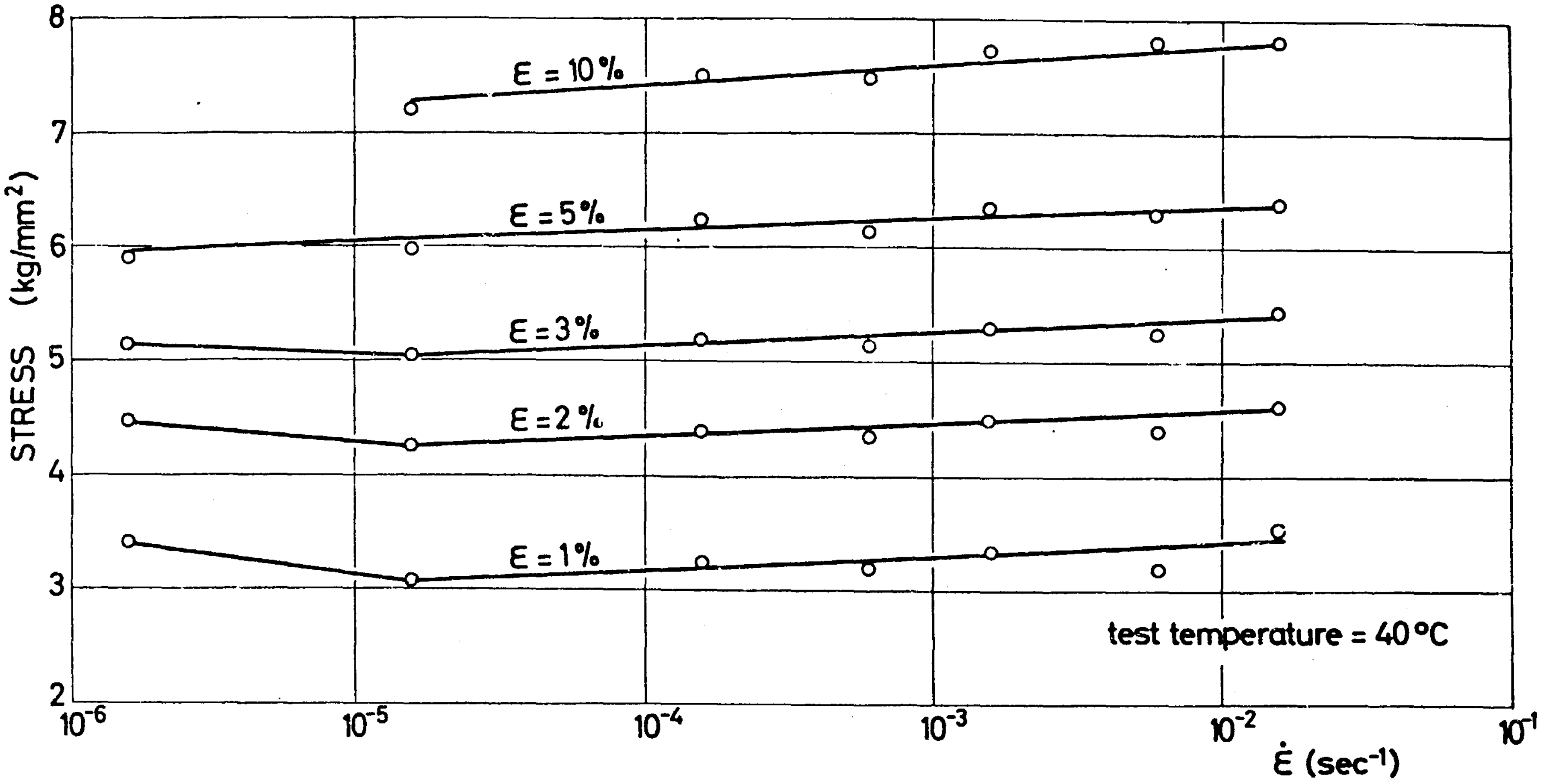
9
8
7
6
5
4
3
2
STRESS (kg/mm²)
ε = 10%
ε = 5%
ε = 3%
ε = 2%
ε = 1%
test temperature = 0 °C
10⁻⁶
10⁻⁵
10⁻⁴
10⁻³
10⁻²
10⁻¹
ε̇ (sec⁻¹)

FIG. 7



STRESS (kg/mm²)
8
7
6
5
4
3
2
ε=10%
ε=5%
ε=3%
ε=2%
ε=1%
test temperature = 23 °C
10⁻⁶
10⁻⁵
10⁻⁴
10⁻³
10⁻²
10⁻¹
ε̇ (sec⁻¹)

FIG. 8



8
7
6
5
4
3
2
STRESS (kg/mm²)
ε = 10%
ε = 5%
ε = 3%
ε = 2%
ε = 1%
test temperature = 40°C
10⁻⁶
10⁻⁵
10⁻⁴
10⁻³
10⁻²
10⁻¹
ε̇ (sec⁻¹)

FIG. 9

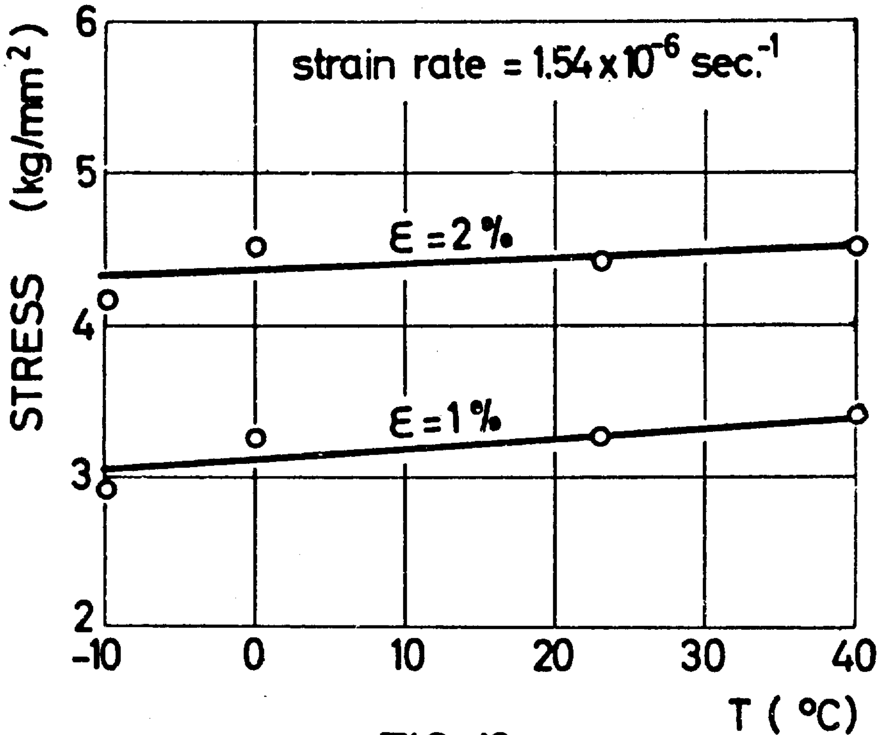
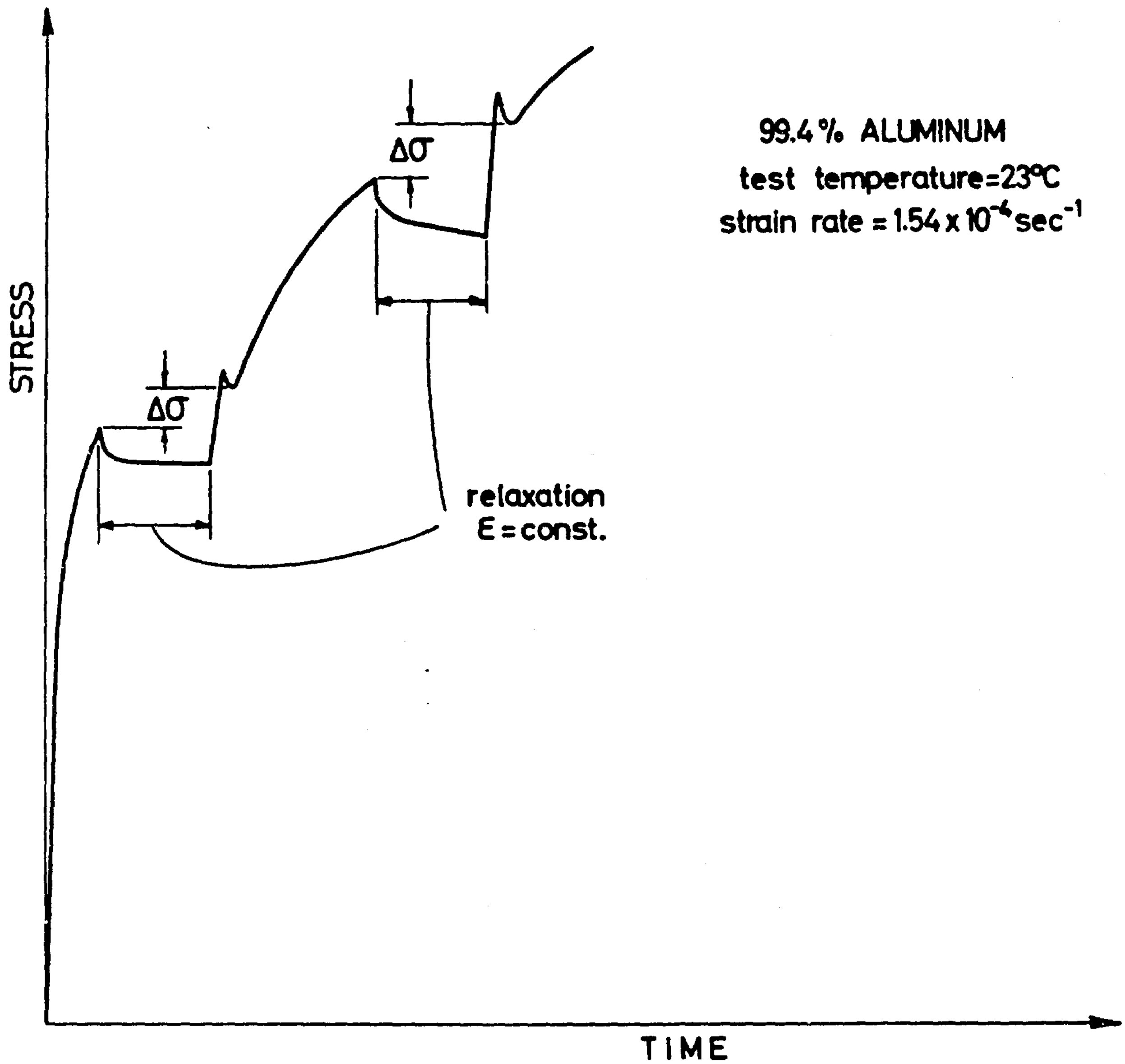

strain rate = 1.54 x 10^-6 sec.^-1
ε = 2 %
ε = 1 %
STRESS (kg/mm^2)
6
5
4
3
2
-10
0
10
20
30
40
T (°C)

FIG. 10



99.4 % ALUMINUM
test temperature=23°C
strain rate = 1.54 x 10^{-4} sec^{-1}
STRESS
TIME
Δσ
Δσ
relaxation
ε = const.

FIG. 11

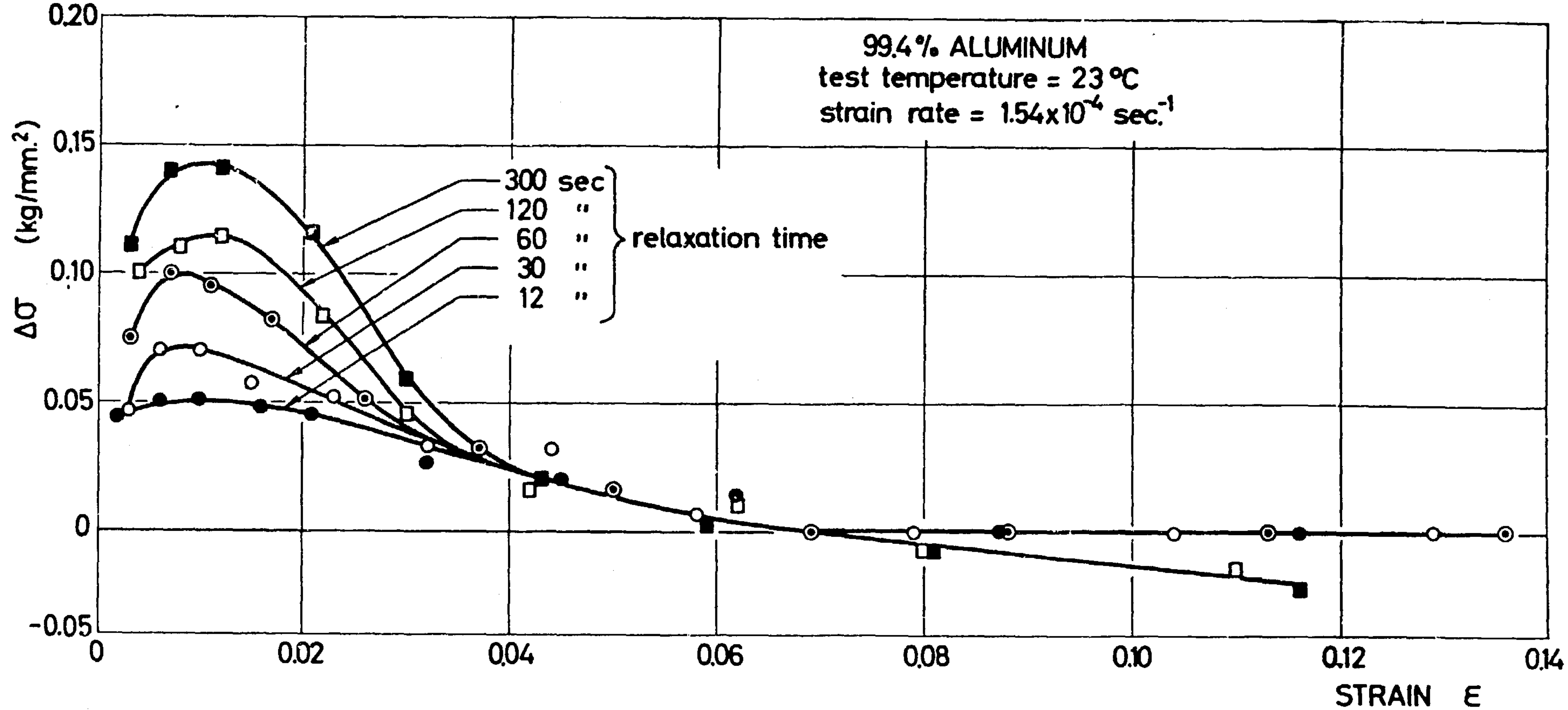
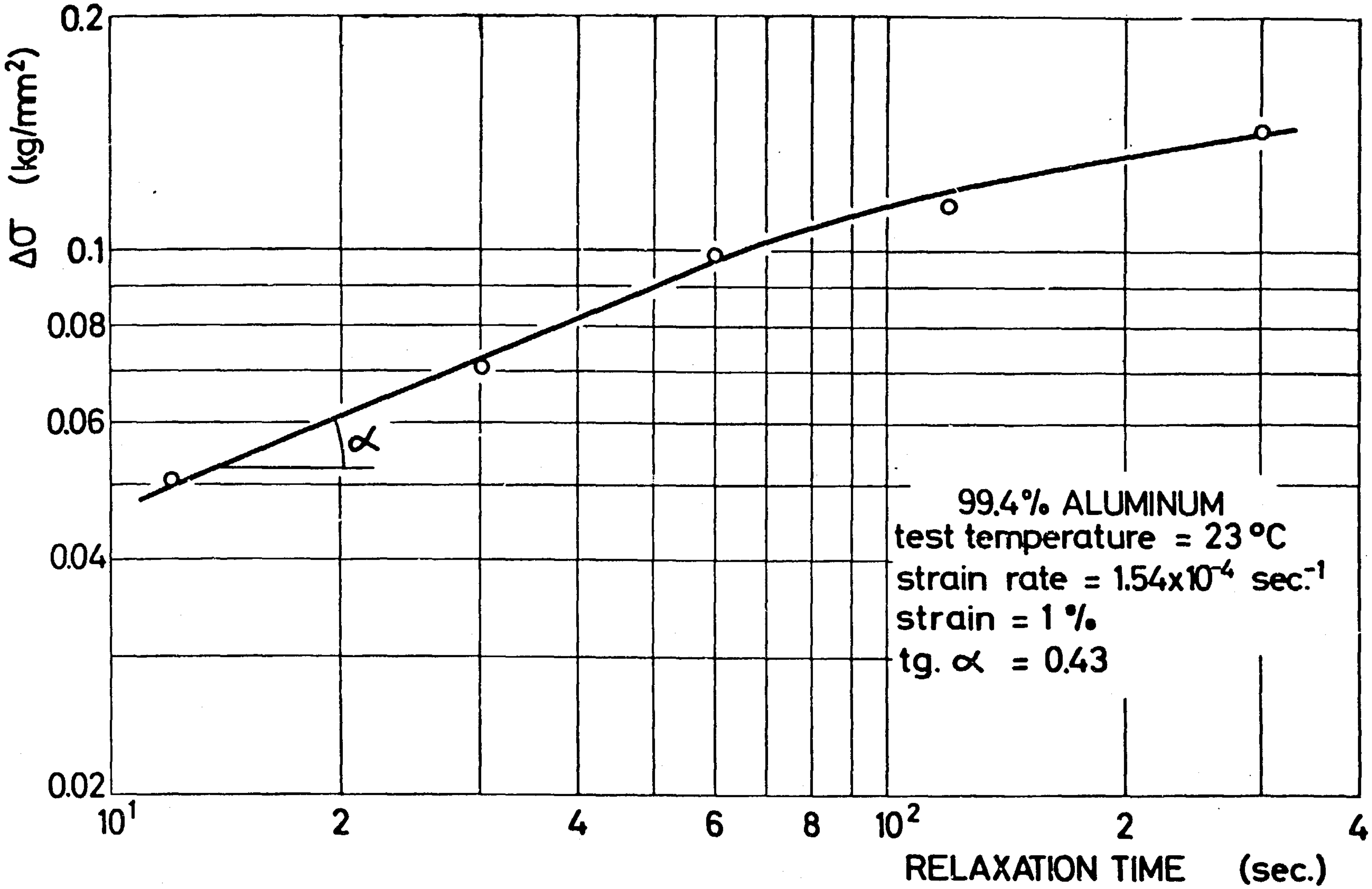
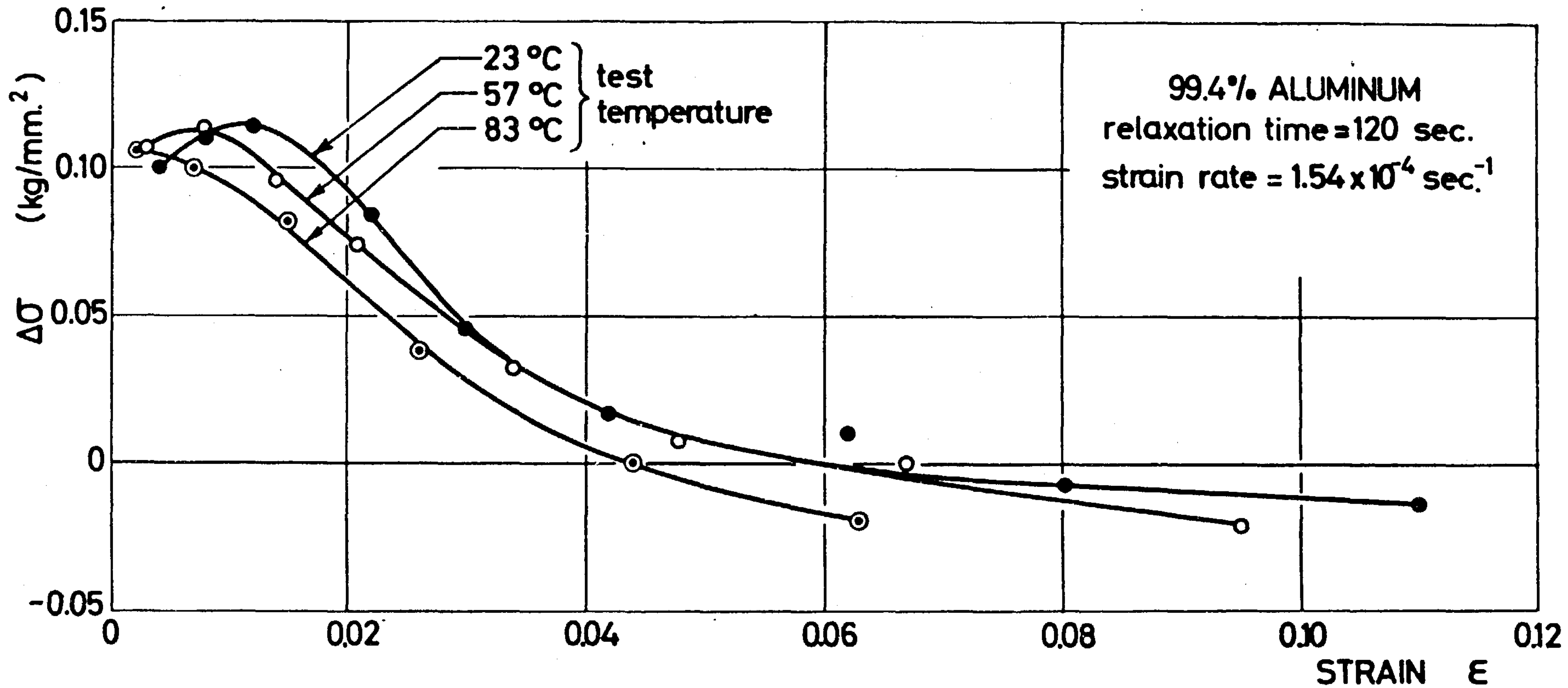

99.4% ALUMINUM
test temperature = 23 °C
strain rate = 1.54x10⁻⁴ sec.⁻¹
300 sec
120 "
60 "
30 "
12 "
relaxation time
Δσ (kg/mm.²)
0.20
0.15
0.10
0.05
0
-0.05
0
0.02
0.04
0.06
0.08
0.10
0.12
0.14
STRAIN ε

FIG. 12



0.2
0.1
0.08
0.06
0.04
0.02
Δσ (kg/mm2)
10 1
2
4
6
8
10 2
2
4
RELAXATION TIME (sec.)
α
99.4% ALUMINUM
test temperature = 23 °C
strain rate = 1.54x10-4 sec.-1
strain = 1 %
tg. α = 0.43

FIG. 13



0.15
0.10
0.05
0
-0.05
Δσ (kg/mm.²)
0
0.02
0.04
0.06
0.08
0.10
0.12
STRAIN ε
23 °C
57 °C
83 °C
test temperature
99.4% ALUMINUM
relaxation time = 120 sec.
strain rate = 1.54 x 10⁻⁴ sec.⁻¹

FIG. 14

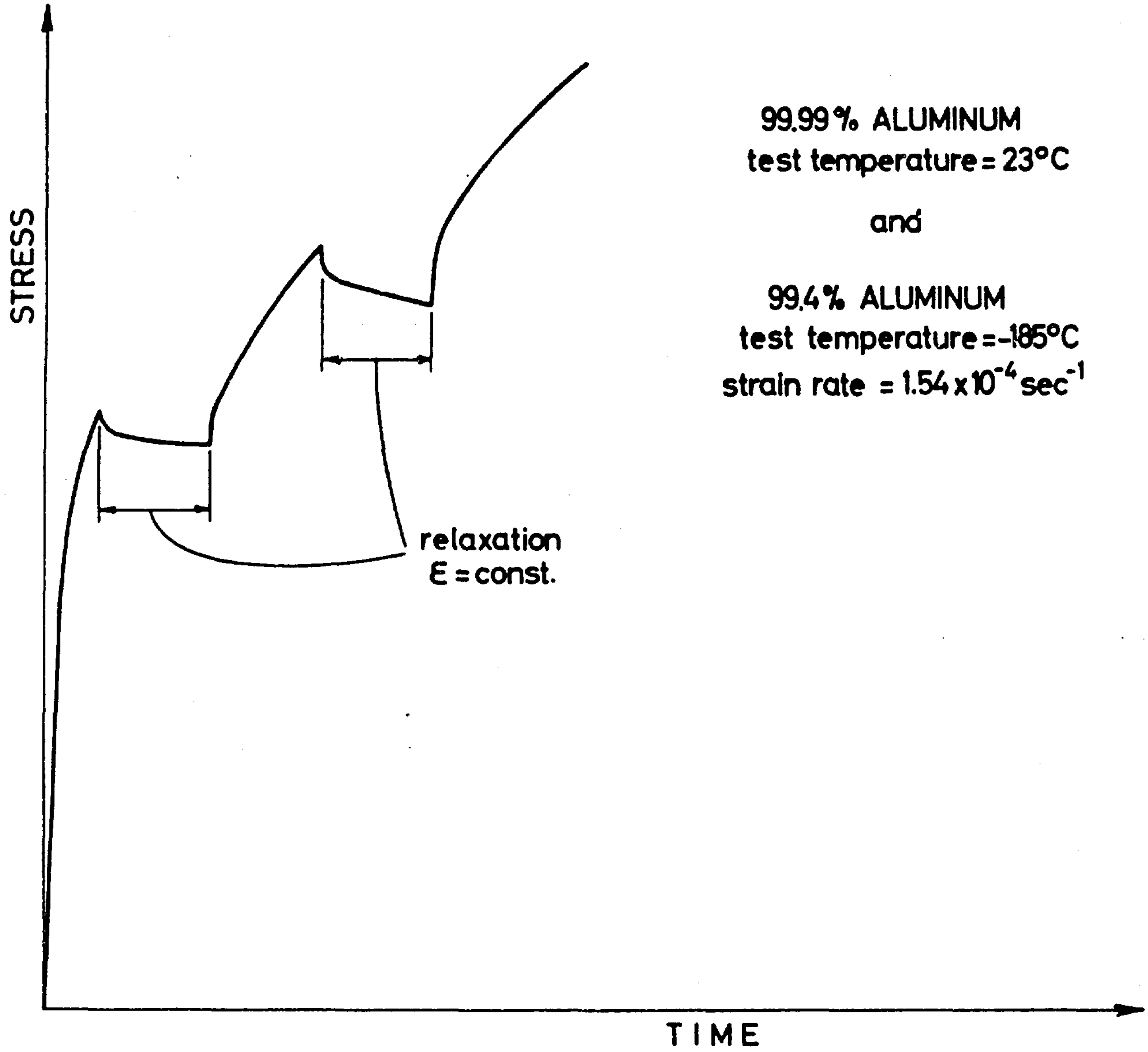
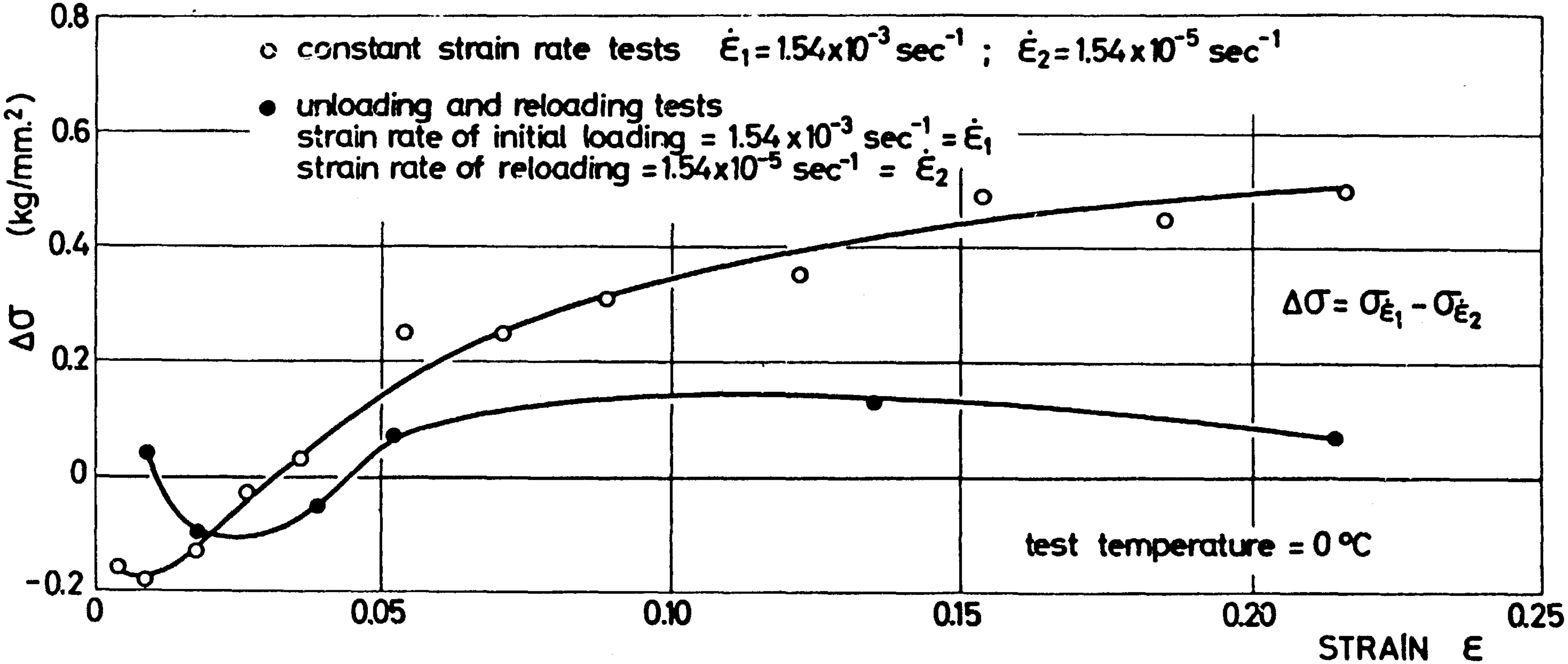

STRESS
99.99% ALUMINUM
test temperature = 23°C
and
99.4% ALUMINUM
test temperature = -185°C
strain rate = 1.54 x 10^-4 sec^-1
relaxation
ε = const.
TIME

FIG. 15



0.8
0.6
0.4
0.2
0
-0.2
Δσ (kg/mm.²)
o constant strain rate tests $\dot{\varepsilon}_1 = 1.54 \times 10^{-3}\ sec^{-1}$; $\dot{\varepsilon}_2 = 1.54 \times 10^{-5}\ sec^{-1}$
● unloading and reloading tests
strain rate of initial loading = $1.54 \times 10^{-3}\ sec^{-1} = \dot{\varepsilon}_1$
strain rate of reloading = $1.54 \times 10^{-5}\ sec^{-1} = \dot{\varepsilon}_2$
$\Delta\sigma = \sigma_{\dot{\varepsilon}_1} - \sigma_{\dot{\varepsilon}_2}$
test temperature = 0 °C
0
0.05
0.10
0.15
0.20
0.25
STRAIN ε

FIG. 16

Security Classification

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13. ABSTRACT

Tensile tests on annealed, commercially pure aluminum specimens were performed in a hard machine at different temperatures for a number of constant strain rates. The results show that a range of strain, strain rate, and temperature exists for which the flow stress decreases with strain rate. This material property, when coupled with suitable mechanical conditions, is believed to be cause of discontinuous, repeated yielding. Various experiments, especially relaxation and reloading tests, show that the decrease of flow stress with strain rate can be directly attributed to rapid strain ageing due to impurity diffusion. Other experiments indicate the same effect for different loading histories. An analytical representation of the strain rate and strain ageing of the material is developed which predicts results in very good agreement with the experiments.

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Security Classification

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14. KEY WORDS	LINK A		LINK B		LINK C	
	ROLE	WT	ROLE	WT	ROLE	WT
Strain Rate						
Strain Ageing						
Commercially Pure Aluminum						
Discontinuous Yielding						

Security Classification