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LUCAS HEIGHTS

THE CREEP PROPERTIES AT AMBIENT TEMPERATURE OF SOME HIGH STRENGTH STEELS

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J.T.A. POLLOCK
S. G. BARTON

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ABSTRACT

Ambient temperature stress relaxation and creep data are reported at various stress levels for two grades of MAR300 steel, a MAR350 steel and a type 301 stainless steel. Analysis has shown that there is a linear relationship between the logarithm of time and relaxed stress, or creep strain. On the basis of this linear relationship, two parameters are calculated that are characteristic of the relaxation and creep behaviour at a given stress level. Using a conjugate relationship suggested by Feltham [1961 - *Phil. Mag.*, 6:259; *J. Inst. Met.*, 89:210], these characteristic parameters are compared to allow an assessment to be made of the possibility of deriving constant load creep characteristics from short-term stress relaxation data. Because of the difficulties in measuring some of the factors in this relationship, a large error band is associated with the derivation at stresses where the work hardening

(continued)

rate is very high. Nevertheless, in this difficult stress range, a working relationship is demonstrated which is sufficient to warrant the application of the approach if design information is required. At stresses in the range 1000-1500 N mm^{-2}, creep increments per year of about 10^{-4} mm mm^{-1} were calculated from the experimental data for the maraging steels. In terms of creep increment per year, the type 301 stainless steel is roughly an order of magnitude inferior to the maraging steels over the same range of stress. At higher stresses in the range 1600-2150 N mm^{-2}, which is beyond the applicability of the MAR300 steels or the type 301 stainless steel, an unchanged yearly creep increment of $\sim 10^{-4}$ mm mm^{-1} was obtained with the MAR350 steel.

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CREEP; STRESS RELAXATION; STAINLESS STEEL-301; MARAGING STEEL-300 (proposed descriptor); MARAGING STEEL-350 (proposed descriptor); AUSTENITE; MARTENSITE; MEDIUM TEMPERATURE; GAS CENTRIFUGES

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1. INTRODUCTION

Little information is available on the long-term mechanical properties of high strength steels at ambient temperatures. In particular, creep characteristics are virtually unknown and difficult to estimate from simple tensile tests in which very little deformation may be exhibited before fracture. Most room temperature applications of high strength steels do not require prior knowledge of creep properties. It is assumed that up to 80 per cent ultimate tensile stress (UTS) no creep occurs. However, there are applications in which dimensional stability during operation is of such importance that any creep occurring under applied load could be critical. Therefore, the magnitude of this creep must be known and taken into account during design. Gas centrifuges for the enrichment of uranium represent such an application. The current work has as a primary goal, the generation of relevant design information. A secondary aim is to make pertinent suggestions for improvements in desired properties.

Among the materials suitable for centrifuge construction, maraging steels have evoked interest among several groups. There are no reports in the literature dealing with the ambient (or sub-ambient) creep properties of maraging steels. However, two reports on the resistance to creep at temperatures around 500°C have appeared [Berns & Brühl 1970; Gurewitz *et al*. 1977].

Maraging steels combine strength and considerable toughness with easy fabrication. The major factor in the production of these attractive characteristics is the hysteresis which exists between the metastable martensite, formed by air cooling the solution treated alloy, and its reversion to the stable, high temperature, austenite structure after heating. This hysteresis allows the virtually carbon free and, therefore, soft martensite to be formed into the required component at room temperature, before the development of high strength by ageing for a few hours at about 485°C. The ageing treatment results in a fine intermetallic precipitation throughout the martensite. In this final form, commercially available steels having ultimate tensile strengths up to 2400 N mm^{-2} retain some ductility ($\sim$ 1.5 per cent elongation) and fracture toughness (K_{1c} > 1500 N $mm^{-3/2}$). The fracture toughness of maraging steel is superior to that exhibited by the competitively strong, quenched and tempered carbon alloy steels.

The direct measurement of creep properties can be time-consuming, particularly in the case of strong materials such as maraging steels which also have very high work hardening rates. The original goal was to develop and apply the stress relaxation technique in an effort to establish a short-term method for characterising the long-term deformation properties of any steel (or alloy). A report was published on the preliminary stages of this work [Pollock & Barton 1976]. However, it was realised that since the analysis used to convert the short-term data [Feltham 1961a, 1961b) had been verified with brass and copper, materials having strength and work-hardening characteristics that are substantially different from those of high strength steels, it was necessary to compare the derived values with directly measured creep data. Two high sensitivity differential displacement creep rigs were constructed to meet this need.

Analysis has been made of experimental data from two brands of MAR300 steels, a MAR350 steel and a type 301 stainless steel. The latter was included to provide a comparison with the maraging steels. Experimental methods and material specifications are presented in Section 2. In Section 3, a brief description of the analysis is given. Experimental data and analysis are presented in Section 4. In Section 5, the usefulness of the approach and the creep characteristics of the steels are discussed. Finally, Section 6 contains some brief conclusions.

2. EXPERIMENTAL PROCEDURES

The chemical compositions and thermomechanical histories of the materials are listed in Table 1. The commonly used KSI (KSI = lb in^2 x 10^3) designations for maraging steels are included in Table 1 and will be used in the text. Variations in concentration of the hardening elements cobalt, molybdenum, aluminium and titanium, together with the ageing response of individual alloys, cause the age-hardened UTS of maraging steels to fall within a scatter band, the upper UTS of which is the grade designation expressed in KSI units. Round-shouldered tensile samples having gauge dimensions 20 mm long x 4 mm wide were stamped from thin sheets of the steels and used to determine the stress/strain relationship and stress relaxation characteristics. Round-shouldered tensile samples having gauge dimensions 100 mm long x 4 mm wide were machined from the same or similarly heat-treated sheets and used to determine creep properties.

2.1 Stress Relaxation Measurements

Stress relaxation measurements were made using a Model 1122 Instron. Figure 1 is a schematic diagram showing the tensile loading and relaxation sequence. The sample is held in a constant temperature oil bath at 35° ± 0.25°C and deformed at a constant strain rate of $8 \times 10^{-5}\ s^{-1}$. At a preselected loading point, L_o, the cross head is stopped. If the sample continues to deform, then the plastic deformation replaces the elastic deformation of the system (*i.e.* sample plus machine) and a relaxation of the applied load is measured as a function of time. Load sensitivity of ± 0.2 N is achieved by the use of load suppression units. Stress, strain (during the loading sequence), stress rate and relaxed time may be measured from the chart record and used to determine a number of parameters which characterise the dislocation dynamics of materials under load. A number of relaxation tests may be measured with the same sample by incremental loading.

2.2 Creep Measurements

Creep data were measured directly on high sensitivity differential displacement creep rigs based on the design of Raj *et al.* [1974]. Accurate long-term temperature control limits the short- to medium-term measurement of small strain rates. This problem is largely overcome by the design of the creep rig. A schematic diagram of the rig is shown in Figure 2. The temperature of the annular tank is held at 35° ± 0.25°C. One sample carries the test load and the other carries a small weight sufficient to hold it taut, but well within its elastic limit.

Creep is measured with linear voltage displacement transducers (LVDT). The LVDT cores rest on platforms connected firmly to the weight trains of the samples, and the transformers are mounted rigidly to the support frame. Small variations in the test space temperature cause each sample to expand or contract equally and simultaneously. Subtraction of the signals from the LVDTs ensures that only creep is measured. Normal experimental procedure requires the measurement of several creep curves by incrementally loading the same sample.

3. DERIVATION OF CONSTANT LOAD CREEP DATA FROM STRESS RELAXATION DATA

3.1 Basic Relationship

Feltham [1961a, 1961b] outlined a method for extracting creep parameters from stress relaxation data. The approach may be applied when

(a) the total strains involved during relaxation and creep are small, and

(b) there is a linear relationship between the logarithm of time and relaxed stress, or strain.

Very high strength steels satisfy condition (a). The experimental data provide a check on the satisfaction of condition (b).

To establish the relationship between stress relaxation and creep, it is necessary to define a number of terms. The applied stress, σ, is the sum of the effective stress, σ^*, which is temperature sensitive, and σ_i, the long range internal athermal stress.

$$\sigma = \sigma^* + \sigma_i \quad . \tag{1}$$

A work-hardening coefficient, Θ, defined as $d\sigma_i/d\varepsilon$, may be measured from the slope of the stress/strain curve at a given stress without introducing too much error. Thus,

$$\Theta = d\sigma_i/d\varepsilon \quad (=d\sigma/d\varepsilon) \quad , \tag{2}$$

where ε is the plastic strain. The strain rate sensitivity of the flow stress may be expressed as

$$\lambda = d\sigma^*/d\ell n\dot{\varepsilon} \quad . \tag{3}$$

During stress relaxation, elastic contraction of the testing system is replaced by continuing plastic deformation of the sample. Thus, the sample plastic strain rate ($\dot{\varepsilon}_p$) and system elastic strain rate ($\dot{\varepsilon}_e$) are equal but opposite, and may be related to the desired stress change by Hooke's law:

$$\dot{\varepsilon}_p = -\dot{\varepsilon}_e = -\dot{\sigma}/E \; , \tag{4}$$

where E is the combined modulus of the sample and test rig and may be measured from the elastic loading section of the chart record. During relaxation, σ_i is assumed constant as a consequence of the small strains involved ($\sim 5 \times 10^{-5}$). Combining Equations (1), (3) and (4), the strain rate sensitivity may be expressed in terms of the relaxation rate:

$$\lambda = \frac{d\sigma^*}{d\ell n\dot{\varepsilon}_p} = \frac{d\sigma}{d\ell n(-\dot{\sigma})} \quad . \tag{5}$$

Guiu & Pratt [1964] have shown, on annealed molybdenum, that the same strain rate sensitivity parameter may be measured from the slope of relaxed stress *versus* the logarithm of relaxed time:

$$\lambda = \frac{-d\sigma}{d\ln(t+c)} \quad . \tag{6}$$

At times long enough to ignore the time constant c,

$$\lambda = \frac{-d\sigma}{d\ln t} \quad . \tag{7}$$

A similar approach may be taken with constant load creep, where $d\sigma = 0$. Under these circumstances, it follows from Equations (1) and (2) that the creep strain may be expressed as

$$d\varepsilon = -d\sigma^*/\Theta \quad . \tag{8}$$

Feltham [1961a] demonstrated experimentally that creep tensile strain data could be fitted to a logarithmic expression of the form:

$$\varepsilon - \varepsilon_o = \delta\ln(t+g) \ , \tag{9}$$

where δ and g are time-independent constants. This expression has found general acceptance for most materials tested at temperatures below 0.3 T_m, where T_m is the absolute melting point [Garofalo 1965]. From Equation (9),

$$\frac{d\varepsilon}{d\ln(t+g)} = \delta \quad . \tag{10}$$

At long times g may be ignored; Equations (8) and (10) may then be combined and lead to

$$\frac{d\sigma^*}{d\ln t} = -\Theta\delta \quad . \tag{11}$$

Comparison shows that Equations (7) and (11) are conjugates; from this it follows that

$$\lambda = \Theta\delta \quad . \tag{12}$$

Equation (12) states that the strain rate sensitivity measured from a relaxation test is equal to that measured from a creep test. Feltham [1961a, 1961b] validated Equation (12) in a series of tests with well annealed brass and copper.

3.2 The Effect of Work Hardening During Relaxation on the Derivation of the Creep Parameter, δ, from the Relaxation Parameter, λ

When experimentally verifying Equation (12), Feltham [1961a, 1961b] ignored work hardening during relaxation. The high work-hardening rates measured with the present materials preclude this simplification.

Following Sargent [1965], a factor to account for work hardening during relaxation was introduced. The simple equality $d\sigma^* = d\sigma$ (from Equation (1) with σ_i = constant) does not hold for relaxation if work hardening occurs; it is replaced by the expression

$$d\sigma^* = d\sigma \left(1 + \frac{\Theta a_o}{M\ell_o}\right) \quad ,$$

where M = elastic loading slope measured from the Instron chart record (N mm^{-1}),

ℓ_o and a_o = initial gauge length (mm) and mass section (mm^2) respectively, and

Θ = work hardening coefficient at the stress of interest (σ_o).

The strain rate sensitivity measured from the stress relaxation data (Equation (7)), is consequently modified to the form

$$\frac{d\sigma^*}{d\ell nt} = \frac{d\sigma}{d\ell nt}\left(1 + \frac{\Theta a_o}{M\ell_o}\right) = \left(1 + \frac{\Theta a_o}{M\ell_o}\right)\lambda \quad , \tag{13}$$

where $\lambda = \frac{d\sigma}{d\ell nt}$ is measured from the experimental data. Therefore, to account for work-hardening during relaxation, Equation (12), which relates the creep and stress relaxation parameters, takes the form

$$\lambda\left(1 + \frac{\Theta a_o}{M\ell_o}\right) = \delta\Theta \quad . \tag{14}$$

Both the factor $\left(1 + \frac{\Theta a_o}{M\ell_o}\right)$ and the work-hardening rate, Θ, may be obtained from the experimental chart record. Equation (14) then takes the form,

$$\lambda \times \frac{M}{M-W'} = \delta \times \frac{MW}{M-W} \times \frac{\dot{c}}{\dot{s}} \times \frac{\ell_o}{a_o} \quad , \tag{15}$$

where W = work-hardening slope (N mm^{-1}) measured at the load of interest, L_o, from a tensile load/extension curve

W' = work-hardening slope (N mm^{-1}) measured immediately before L_o during a relaxation test, and

$\dot{c}$ and $\dot{s}$ = chart and cross head speeds, respectively (mm m^{-1})

Normally W and W' are equal and Equation (15) reduces to the form

$$\lambda = \delta \times W \times \frac{\dot{c}}{\dot{s}} \times \frac{\ell_o}{a_o} \quad . \tag{16}$$

4. EXPERIMENTAL RESULTS AND ANALYSIS

4.1 Tensile Stress/Strain Behaviour

Tensile stress/strain curves for the maraging steels in both the solution-treated and age-hardened conditions are presented in Figure 3. Data for the type 301 stainless steel, tested in the as-received cold-worked condition, are also included in this figure. The solution-treated MAR300 steels were only slightly weaker than the solution-treated MAR350 steel which had a UTS of ~ 1150 N mm^{-2}. However, in the age-hardened condition, the MAR350 developed a UTS of nearly 2500 N mm^{-2} which is almost 25 per cent higher than the MAR300 steels. At 2500 N mm^{-2}, the UTS of the fully hardened MAR350 is very close to the grade specification. The MAR300 steels in the same condition had a UTS value of ~ 1900 N mm^{-2} which is lower than the grade specification. However, this low value is within the range normally encountered for MAR300 steels. The relative increase in strength of the MAR350 steel over the MAR300 steels on ageing is the result of the increase in Co content to ~ 12 per cent, combined with the increase in the supplementary hardening element Ti to about 1.5 per cent (see Table 1). Cobalt is not believed to be directly responsible for raising the hardened strength, but operates by driving molybdenum out of solution to form the precipitate Ni_3Mo, which is the principal hardener in MAR300 and MAR350 steels [Floreen 1968]. The tensile stress/strain data for the type 301 stainless steel were similar to those for the age-hardened MAR300 steels.

4.2 Creep and Stress Relaxation

4.2.1 Maraging steels

Creep and stress relaxation test measurements have been used to calculate the strain rate sensitivity gradients, $d\varepsilon/d\ell nt$ and $d\sigma/d\ell nt$, as a function of stress. These gradients are required to test the usefulness of Equation (16) which, for a given stress, relates the long-term creep gradient, $d\varepsilon/d\ell nt$, to that measured in the short-term stress relaxation test, $d\sigma/d\ell nt$. The latter is obtained by dividing the gradients measured from load drop *versus* logarithm of time data by the initial sample cross section a_o. Examples of the use of the measured data are presented in Figures 4 to 7 for the Vascomax MAR300 steel in the solution-treated and aged conditions. Two or more sets of data were measured for each material. Each group was obtained by incrementally loading a single sample. Stress increments of ~ 30-300 N mm^{-2} were used. Although the loading sequences were not identical, groups of data measured with the same maraging steel fell within an error band of ± 30 per cent. The scales chosen for these figures are convenient for condensed presentation and show the variation in behaviour.

Actual measurements of the strain rate sensitivity gradients were made from individual plots for each load using scales to suit the data. The creep data are plotted as the accumulated strain *versus* the logarithm of time. Thus the ordinate also contains the instantaneous extension which accompanied the addition of a new load increment. Reasonable agreement was noted between the accumulated total strain during creep testing and that measured during an uninterrupted tensile test.

For the maraging steels, a satisfactory linear relationship was obtained in all cases between load drop or creep strain and the logarithm of time. Significiant relaxation or creep occurred at stresses below the 0.1 per cent proof stress. As expected, the relatively soft, solution-treated, maraging steels produced a larger variation in gradient than the age-hardened steels. Thus, measurement of the strain sensitivity slopes from the softer materials is more accurate than that measured with the age-hardened steels. A check on the applicability of using the relaxation gradient $dL/d\ell nt$ as a measure of the strain rate sensitivity was carried out [Guiu & Pratt 1964]. The relaxed load was plotted against the logarithm of related load rate (dL/dt) for a number of relaxations. A linear relationship, having the same slope, within ± 5 per cent, as that measured from load drop *versus* the logarithm of time was obtained, confirming the use of this simpler analysis.

The maraging steels continued to extend through the period of each creep test within the stress range 50-90 per cent UTS. With the aged steels, the overall extension at moderate creep loads was very small ($\sim 10^{-5}$ strain over 2000-3000 min) and became progressively more difficult to measure.

Using Equation (16), the stress relaxation gradients, $\lambda = d\sigma/d\ell nt$, have been converted to an estimate of the creep parameter, δ_R. In Figures 8 to 10 the logarithm of these derived values, represented by crosses (x), are plotted as a function of the stress at which the relaxation started. The solid line is a visual fit to the directly measured parameters, $\delta = d\varepsilon/d\ell nt$, which are represented by dots (•). These figures show that the derived data for the solution treated steels agree reasonably well with the directly measured data. In the age-hardened steels the agreement is not as good, although the overall trends are similar. There are at least two possible reasons for this less satisfactory agreement. First, the size and range of strain rate sensitivity gradients measured are small and graphical analysis involves a large error. Second, the work-hardening rates of these materials are very high, *e.g.* ~ 500 000 N mm^{-2} compared with 20 000 N mm^{-2} for 70/30 brasses [Feltham 1961a].

It is important, during incremental testing, that each relaxation should occur from a stress sufficiently larger than the previous stress to ensure that normal stress/strain behaviour was taken before the cross-head was stopped. The very small total strain to fracture in these materials combined with the high work-hardening rates means that in some instances this condition may not be realised.

Up to about 80 per cent UTS, where significant deformation occurs during a tensile test, the creep parameters measured with the age-hardened steels, both direct and derived, show little variation. The range of values measured with each steel is given in Table 2. Although the direct and derived ranges of values for the Sandvik steels agree within a factor of two, the range of the derived values for the Vascomax 300 steel is further removed from the essentially constant, low value of 3.5×10^{-6} directly measured. The near constancy of the directly measured creep parameter of the Vascomax 300 below 1500 N mm^{-2} was measured in two separate tests. Above 1500 N mm^{-2}, the direct parameter increased rapidly and became larger than the derived parameter. Only the Vascomax 300 showed this behaviour. (The solution-treated Vascomax 300 shows a similar crossover trend for the derived and directly measured parameters.)

4.2.2 Type 301 stainless steel

Graphical analysis of the data obtained when incrementally loading type 301 stainless steel creep specimens required the introduction of large time constants (~ 500 min) to make the creep strain fit a linear relationship with regard to the logarithm of time. This large time constant is probably the result of the load increments being too small for the early establishment of the new creep conditions in this material. A sample loaded in one stage from ~ 750 to 1400 N mm^{-2} produced a linear relationship without the use of a time constant. More importantly, the gradient measured in this test, $d\varepsilon/d\ell nt = 2.5 \times 10^{-5}$, was close to that measured with two samples loaded incrementally to the same stress, *i.e.* 3.2 and 3.7×10^{-5}, respectively.

As may be observed in Figure 11, very little variation in creep behaviour was measured over the 1200-1650 N mm^{-2} stress range. Over this stress range, values of $d\varepsilon/d\ell nt$ varied between 2.5×10^{-5} and 6.5×10^{-5}. Given the errors in determination of the creep parameter, arising from the small strains involved and the graphical analysis, a constant value of 5×10^{-5} may be assigned to the creep parameter. Similarly, although the time constants required for the stress relaxation data were not unusually large, the range of gradients measured, $\lambda = d\sigma/d\ell nt$, was small. Over the same stress range the parameter, δ_R, derived with Equation (16) and the relaxation gradients, ranged from 5.6×10^{-5} to 8.6×10^{-5}. The range of parameters, direct and derived, measured at stresses up to 80 per cent UTS are included in Table 2.

5. DISCUSSION

5.1 Usefulness of the Analysis for Deriving Creep Characteristics from Relaxation Data

The data presented in Figures 8 to 12 and the tensile properties of the materials allow the following comments to be made on the applicability of the method. In general, the agreement between the direct and derived creep parameters improves as the stress level at which they are measured increases. Exceptions to this are the very low creep data measured on the aged Sandvik 350, which showed very little deformation before fracture, and the high stress data measured on the Vascomax 300. This indicates that the analysis is more successful when the data have been obtained at stress levels where substantial deformation takes place. A more accurate determination of the strain rate sensitivities, $\lambda = d\sigma/d\ell nt$ and $\delta = d\varepsilon/d\ell nt$, which accompanies the larger stress drop, or strain, at these stress levels is then possible. Of course, relaxed load, $d\sigma$, and creep strain, $d\varepsilon$, may also be increased by an increase in the sample dimensions, but this action is limited by the experimental system.

The improvement in the agreement between the parameters with increasing stress also reflects the likelihood that the deformation condition of the sample during the loading part of the relaxation tests was identical with that over the same stress range during an uninterrupted tensile test. This means that the simplification of Equation (15) to the form of Equation (16), based on the work-hardening rates W (tensile test) and W' (relaxation test) being equal, is more likely to be justified.

At stresses close to the proof stress in these steels, the work-hardening rates are very high and it is difficult to check if the simplification is justified. Even so, the application of Equation (16) is not warranted owing to the error in measuring W' from the expanded load scale used in the relaxation test*

**Analysis of data measured with more easily deformed materials, namely molybdenum and a high strength wrought aluminium alloy type 7075 T6, confirms the validity of these comments. Good agreement was obtained for the molybdenum alloy at stresses above the yield point with Equation (16). Equation (15) was used with the aluminium alloy which exhibited strain ageing. However, the deformation characteristics of this alloy allowed W' to be measured accurately, and good agreement was obtained between the derived and directly measured parameters.*

The inferior creep behaviour of the type 301 stainless steel compared with that of aged maraging steels is evident from Figure 12. Table 3 indicates that at stress levels up to 85 per cent UTS, 5-10 times more creep may be expected with the type 301 steel. This confirms a conclusion drawn from relaxation data presented by Pollock & Barton [1976]. The relatively poor performance of the type 301 steel highlights the problem of assessing the likely ambient temperature creep behaviour of high strength steels on the basis of ultimate tensile strength. Since the MAR300 steels have similar UTS to the type 301 steel, the superior creep properties of the former at stresses up to 80 per cent UTS could not have been predicted.

Overall, the aged Sandvik 350 has superior creep properties. At stresses up to 90 per cent UTS, the directly measured creep parameter for this steel scarcely varies from about 4×10^{-6}. At stresses below 1500 N mm^{-2}, the aged Vascomax MAR300 has similar creep properties to the aged Sandvik 350. At similar stresses below 1500 N mm^{-2}, the parameter directly measured with the Sandvik MAR300 is about a factor of three larger. It is difficult to explain the constancy of the creep parameter for the aged Vascomax MAR300 below 1500 N mm^{-2} at about 4×10^{-6}. As noted earlier, this result was obtained with two samples.

At stresses up to 1500 N mm^{-2}, the creep increments calculated for the maraging steels with the direct creep parameters are about 10^{-4} mm mm^{-1} y^{-1}. If materials having creep increments per year of less than 10^{-3} mm mm^{-1} are suitable for centrifuge applications, then clearly the age-hardened type 300 and 350 maraging steels are suitable for use at this stress level. The type 301 stainless steel, although inferior to the maraging steels, would also meet this specification at stresses up to 1500 N mm^{-2}. If the spinning centrifuges generate stresses higher than 1500 N mm^{-2}, then the higher strength MAR350 would have to be used. However at lower stresses, where the creep characteristics are similar, the MAR300 steels would have a clear advantage owing to their better fracture toughness of $\sim$ 3000 N $mm^{-3/2}$ when compared with $\sim$ 1500 N $mm^{-3/2}$ for MAR350 steel [Magnee *et al*. 1973].

6. CONCLUSIONS

(1) The method of deriving creep properties from short-term stress relaxation data has application for design purposes. For a given type of steel, the agreement between the derived and directly measured creep characteristic parameters improves with increasing stress. This reflects the greater accuracy with which the data can be measured and the likelihood that the deformation state of the material is well defined. The method is conservative from the point of view of design. Predicted creep characteristics over the ranges of interest are higher than those measured directly.

(2) Each of the age-hardened maraging steels, MAR300 and 350, had superior creep properties to the type 301 stainless steel. At stresses up to 1500 N mm^{-2}, where the maximum allowable strain increment for centrifuge operation is 10^{-3} mm mm^{-1} y^{-1}, the maraging steels exhibited strain increments an order of magnitude less, whereas the type 301 steel approached this maximum value. Because of their better fracture toughness, the MAR300 steels would find preference over the MAR350 steel at stresses up to 1500 N mm^{-2}. However, at stresses in the range 1600-2150 N mm^{-2}, above the range of application of MAR300 steels, the MAR350 retained essentially the same creep properties.

7. ACKNOWLEDGEMENTS

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TABLE 1

COMPOSITION, CONDITION, 0.05% PROOF STRESS AND ULTIMATE TENSILE STRENGTH FOR THE MATERIALS INVESTIGATED

Material	Composition (wt %) Ni	Mo	Co	Ti	Al	C	P	S	Cr	Mn	Si	Condition	0.05% Proof Stress ($N\ mm^{-2}$)	UTS ($N\ mm^{-2}$)
Sandvik MAR300	*17.80	4.90	7.8	0.50	0.16	0.020	0.002	0.004	0.01	0.04	0.05	Soln. treated 1 h at 815°C	780	970
												Soln. treated then 3 h at 485°C	1825	1800
Sandvik MAR350	*18.15	4.40	10.8	1.46	0.17	0.010	0.004	0.007	< 0.01	0.03	0.04	Soln. treated 1 h at 815°C	800	1170
												Soln. treated then 3 h at 485°C	2100	2470
Vascomax MAR300	†18.5	4.8	9.0	0.60	0.10	0.03	0.01	0.01	n.a.	0.10	0.10	Soln. treated 1 h at 815°C	780	1000
												Soln. treated then 3 h at 485°C	1600	1800
Type 301 Stainless Steel	† 8.0	n.a.	n.a.	n.a.	n.a.	0.15	n.a.	n.a.	18.0	2.0	1.0	As received cold rolled sheet	1600	1850

*Analysis

†Manufacturer's specifications

n.a. = not analysed

TABLE 2

RANGES OF CREEP PARAMETERS MEASURED AT STRESSES UP TO 80% UTS

Material	~ 80% UTS ($N\ mm^{-2}$)	Creep Parameter δ_c, Direct Measurement	Creep Parameter δ_R, Derived from Relaxation
Sandvik MAR300, aged	1550	$5-12 \times 10^{-6}$	$9-30 \times 10^{-6}$
Sandvik MAR350, aged	1950	$4-5 \times 10^{-6}$	$9-11 \times 10^{-6}$
Vascomax MAR300, aged	1550	$\sim 3.5 \times 10^{-6}$	$8-30 \times 10^{-6}$
Type 301 stainless, as received	1550	$25-50 \times 10^{-6}$	$50-75 \times 10^{-6}$

TABLE 3

CREEP RATES EXPRESSED AS STRAIN INCREMENT CALCULATED USING EQUATION (9) AND EXPERIMENTALLY DETERMINED δ_c AND δ_R VALUES

Strain Increment/Year ($\times 10^5$)

Stress ($N\ mm^{-2}$)	Sandvik MAR300 Aged: Direct Measurement δ_c	Sandvik MAR300 Aged: Derived from Relaxation δ_R	Vascomax MAR300 Aged δ_c	Vascomax MAR300 Aged δ_R	Sandvik MAR350 Aged δ_c	Sandvik MAR350 Aged δ_R	Type 301 Stainless Steel δ_c	Type 301 Stainless Steel δ_R
2200	-	-	-	-	5	20	-	-
2000	-	-	-	-	5	13	-	-
1800	-	-	-	-	4	12	-	-
1600	13	26	52	52	4	12	120	120
1400	10	13	4	13	4	12	66	92
1200	8	12	4	12	-	-	39	79

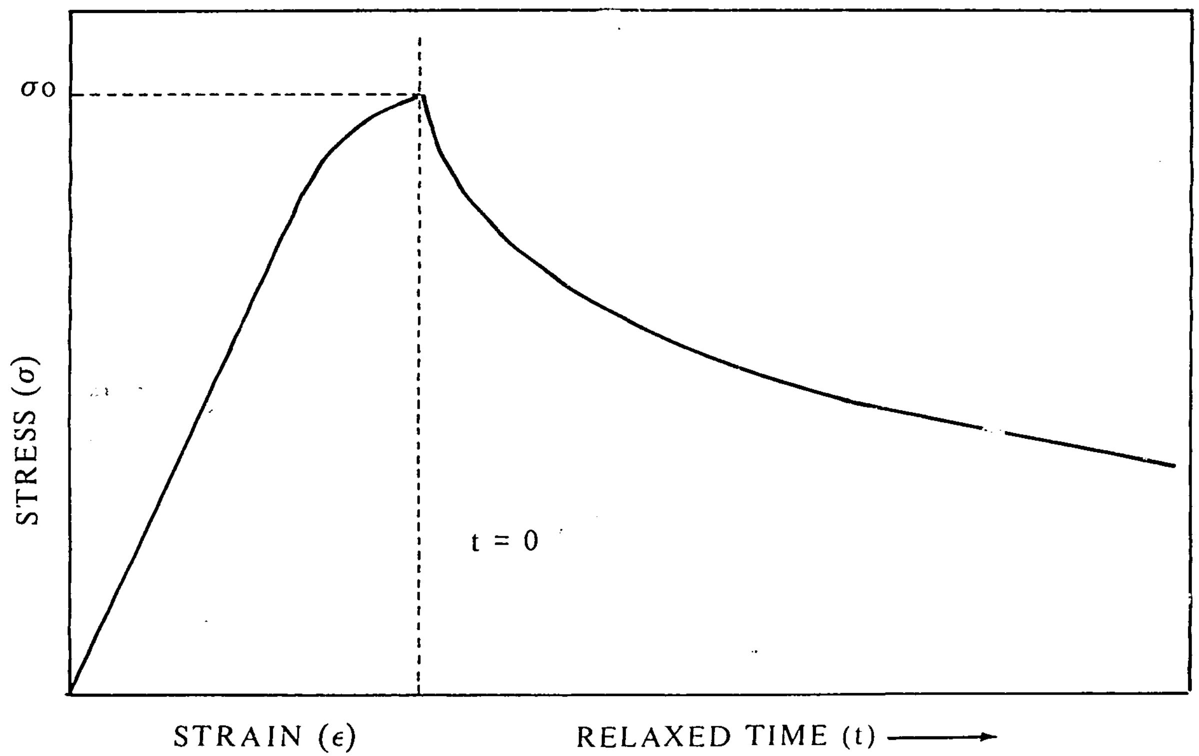

FIGURE 1. SCHEMATIC DIAGRAM SHOWING TENSILE LOADING PATH AND RELAXATION CURVE

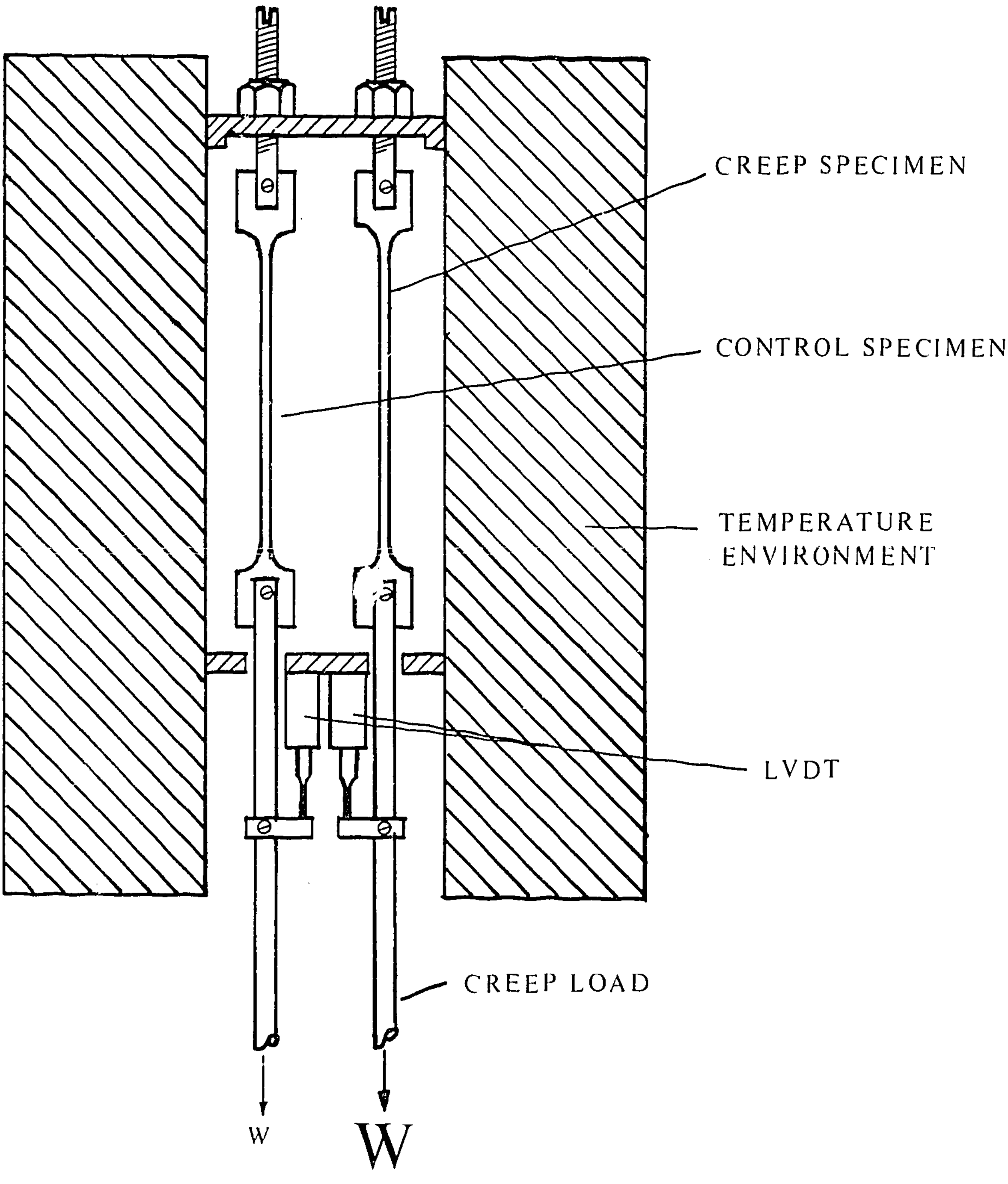

FIGURE 2. SCHEMATIC DIAGRAM OF DIFFERENTIAL CREEP DISPLACEMENT RIG

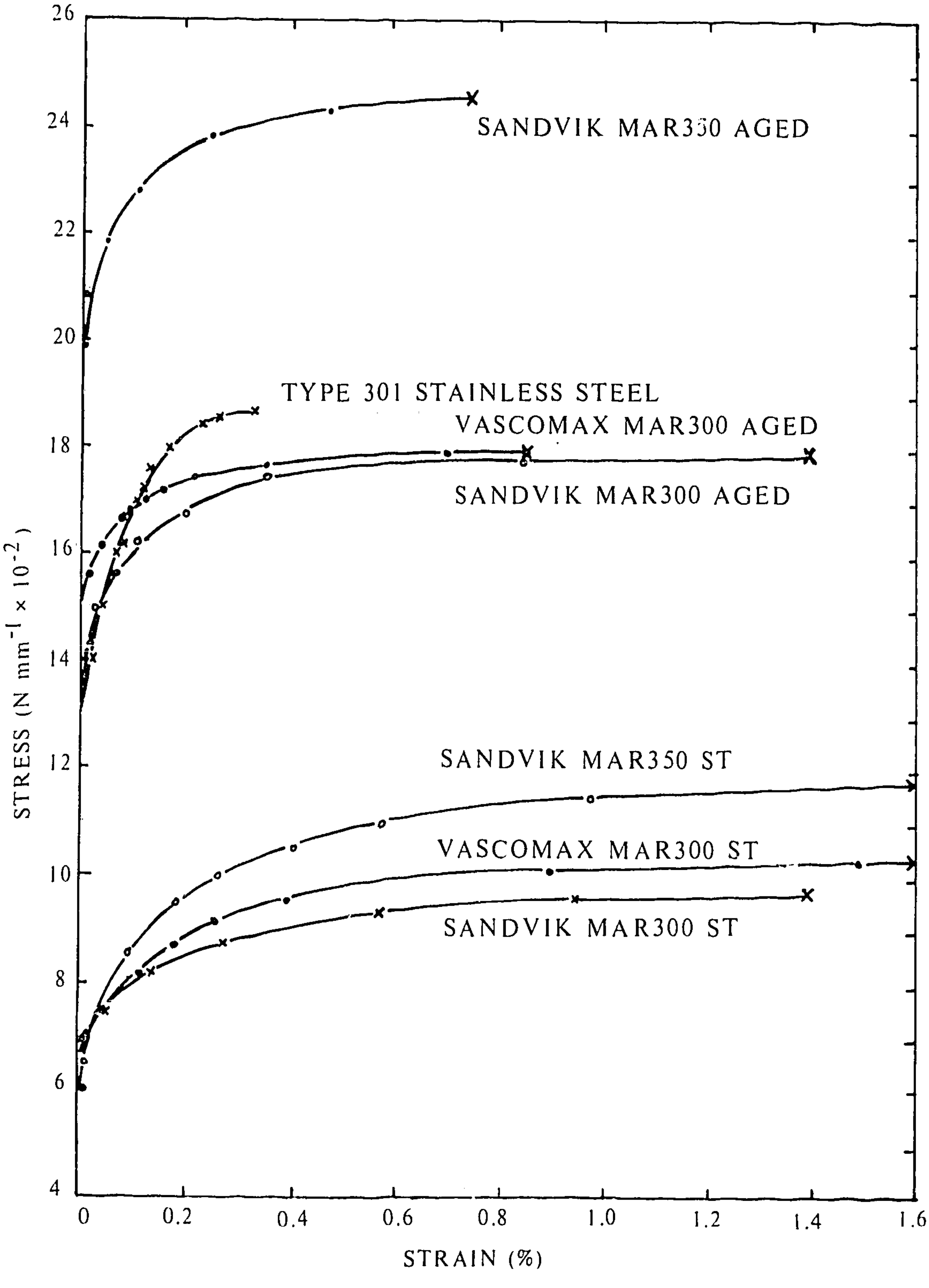

FIGURE 3. TENSILE STRESS/STRAIN DATA FOR MATERIALS INVESTIGATED

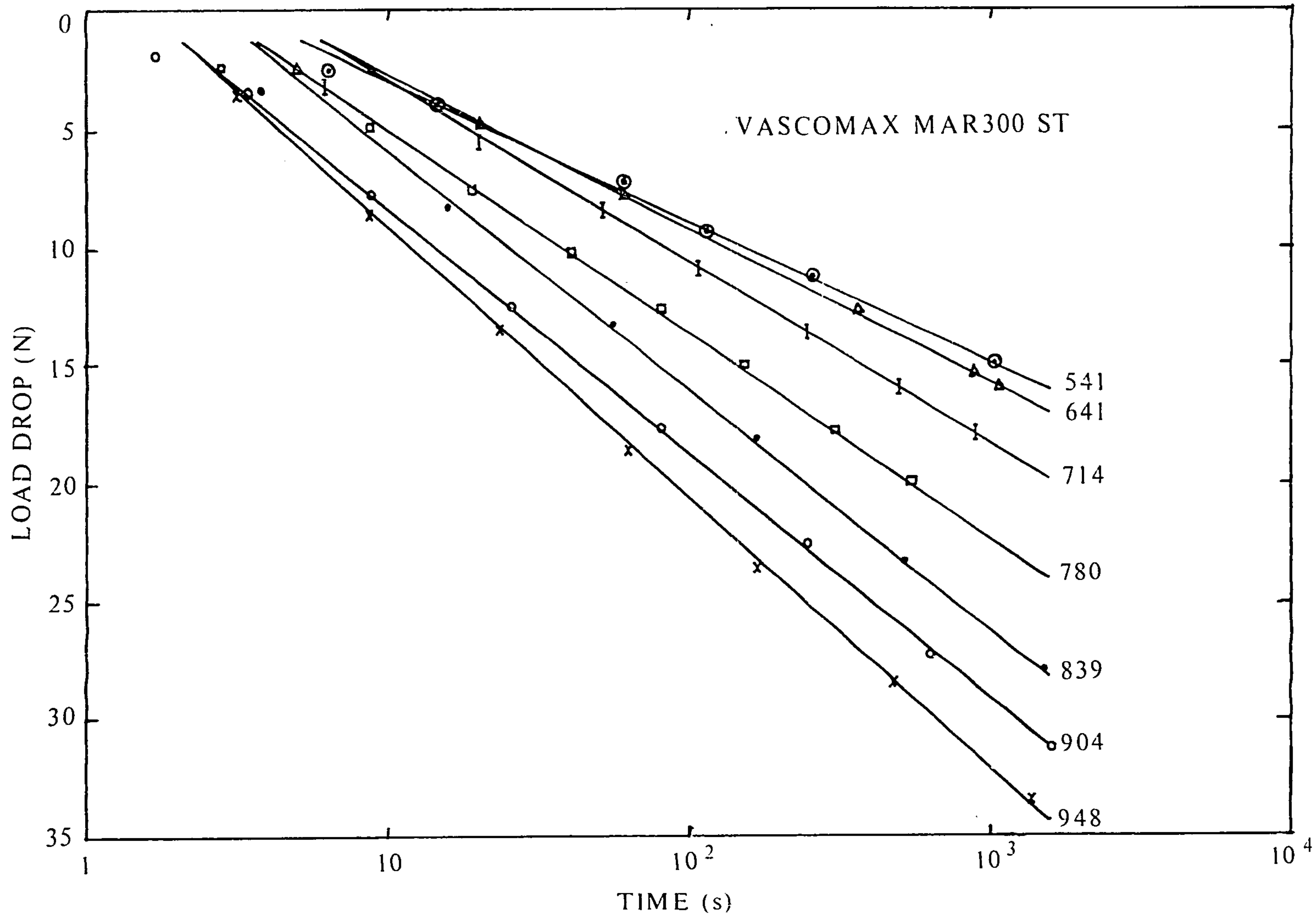

FIGURE 4. VASCOMAX MAR300 ST. RELAXED LOAD VERSUS LOG TIME

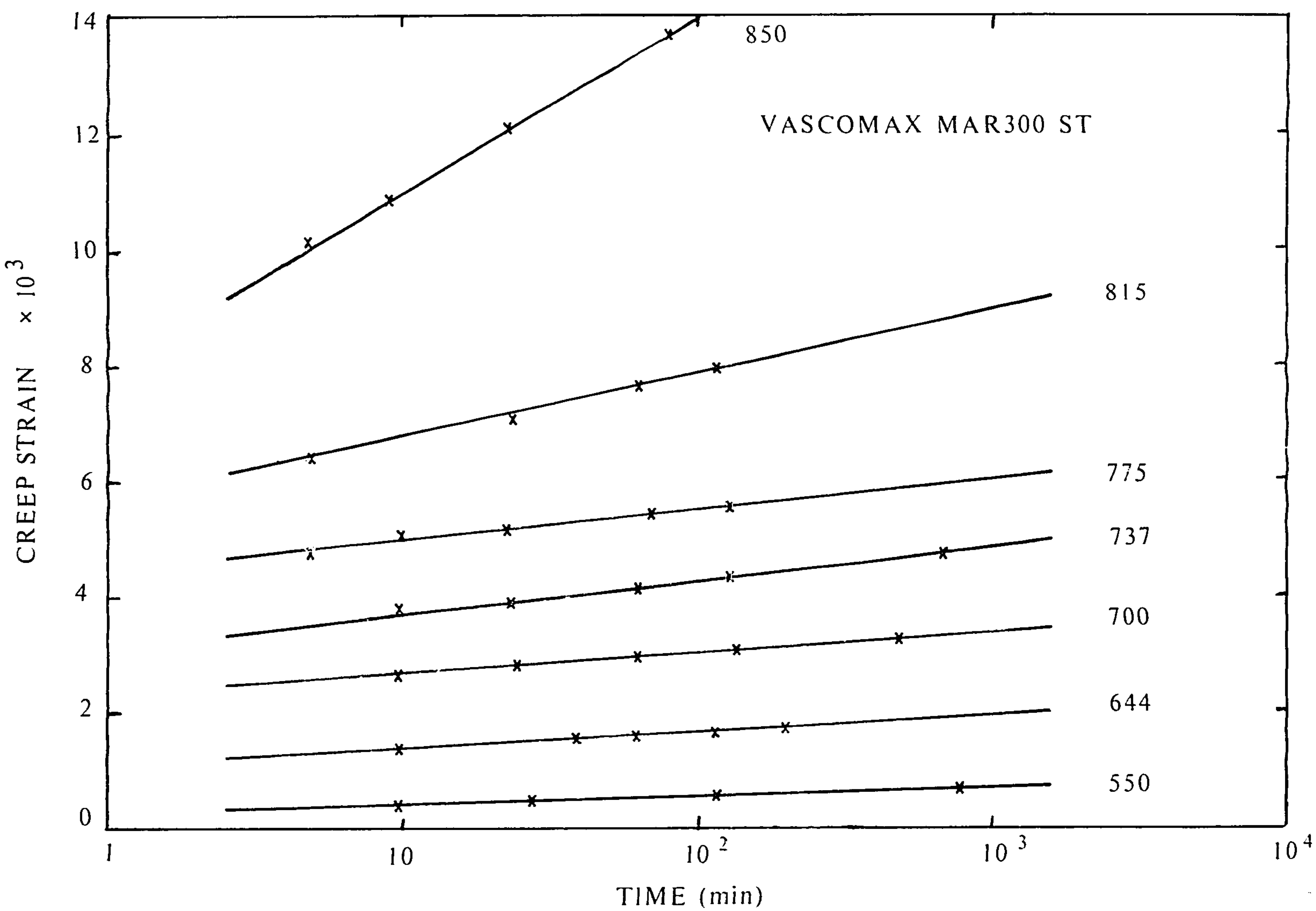

FIGURE 5. VASCOMAX MAR300 ST. CREEP STRAIN AT VARIOUS STRESS LEVELS

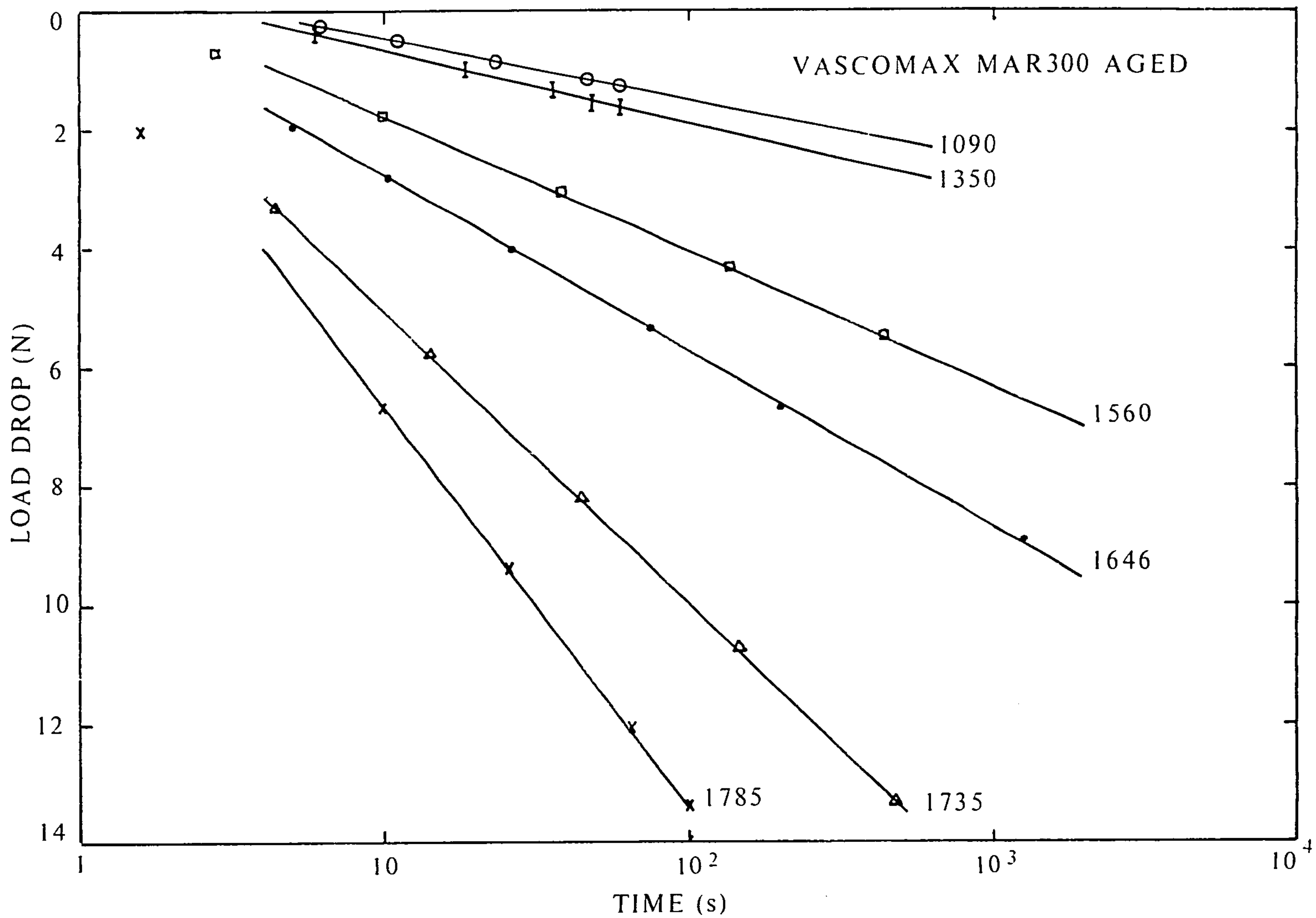

FIGURE 6. VASCOMAX MAR300 AGED. RELAXED LOAD VERSUS LOG TIME

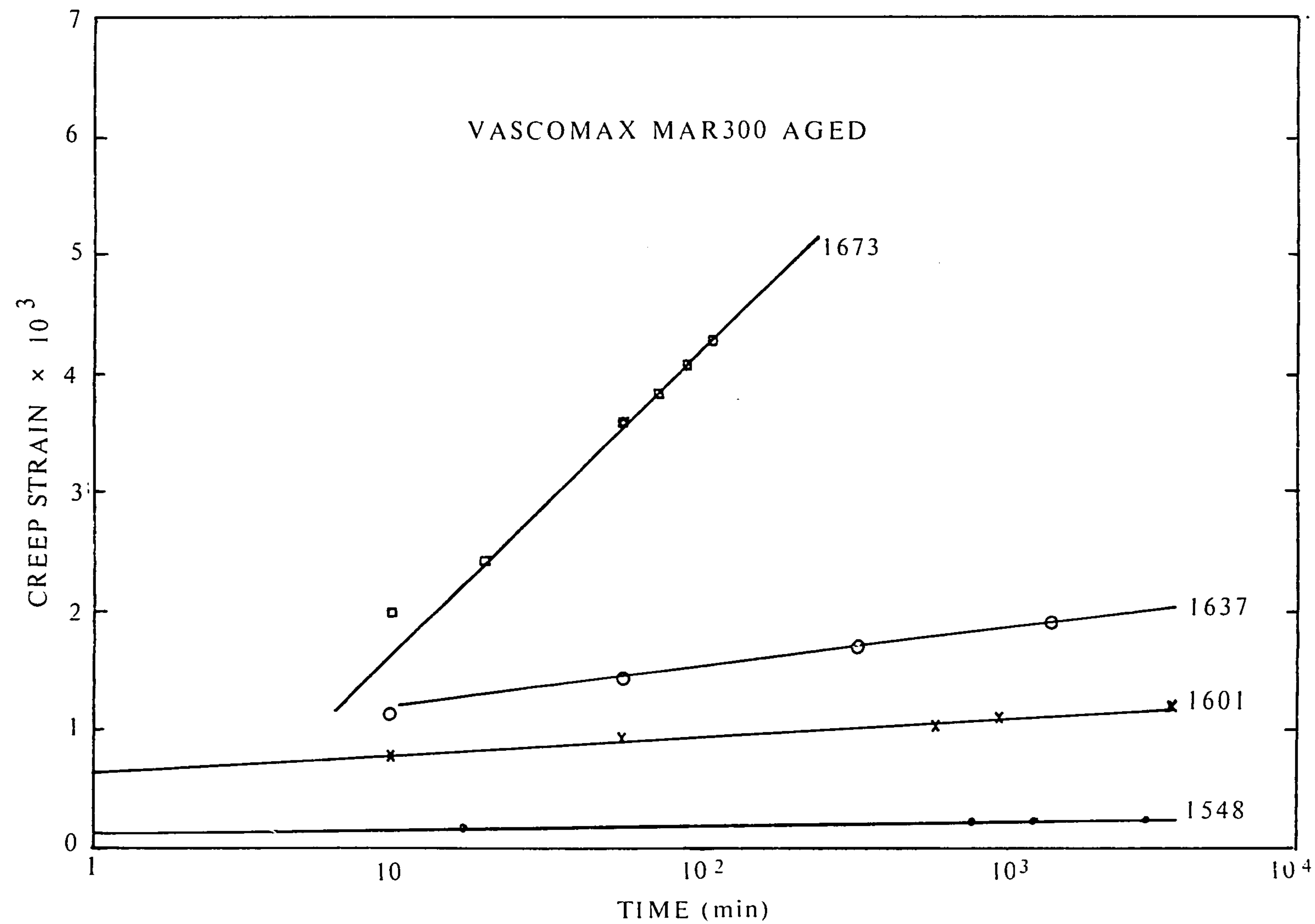

FIGURE 7. VASCOMAX MAR300 AGED. CREEP STRAIN AT VARIOUS STRESS LEVELS

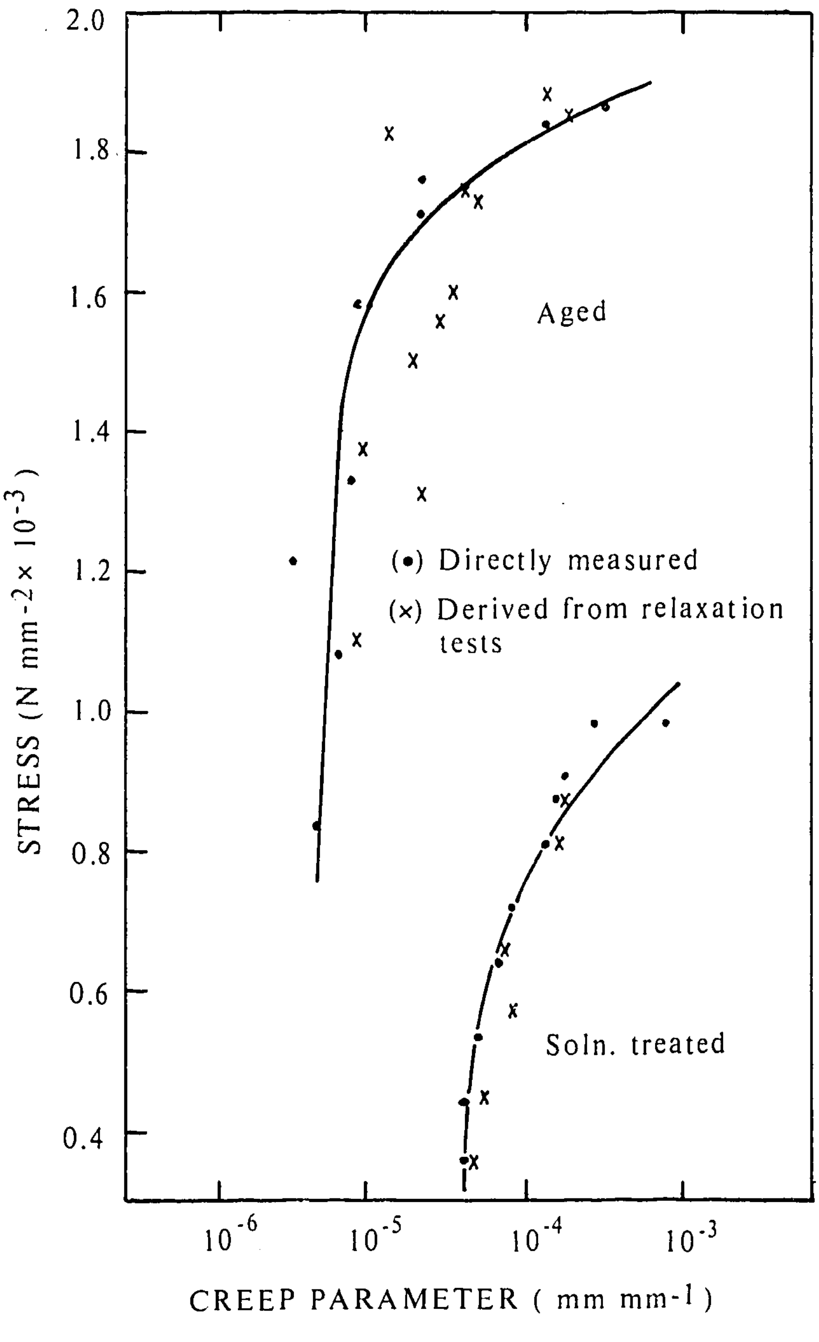

FIGURE 8. SANDVIK MAR300. CREEP PARAMETER VERSUS STRESS

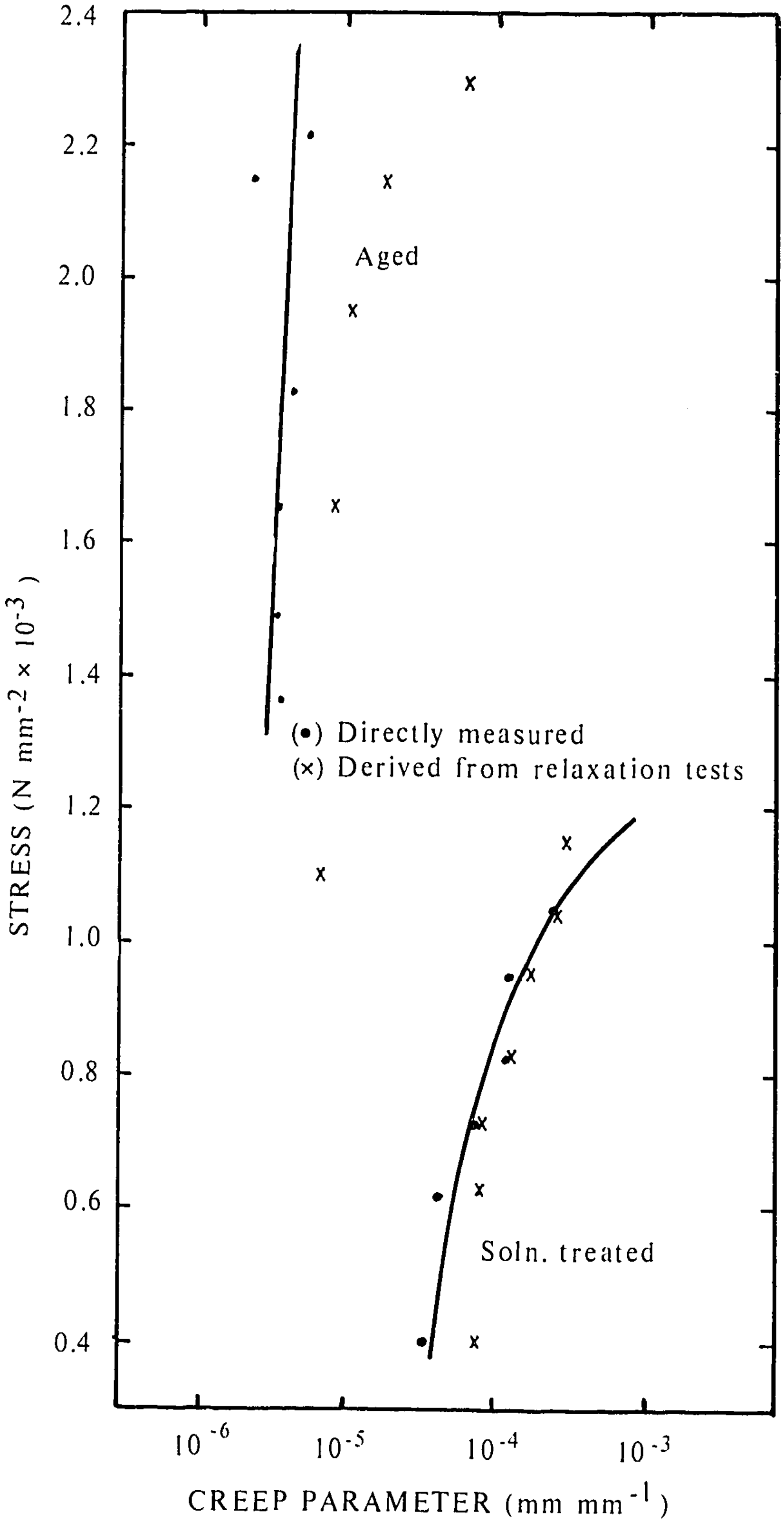

FIGURE 9. SANDVIK MAR350. CREEP PARAMETER VERSUS STRESS

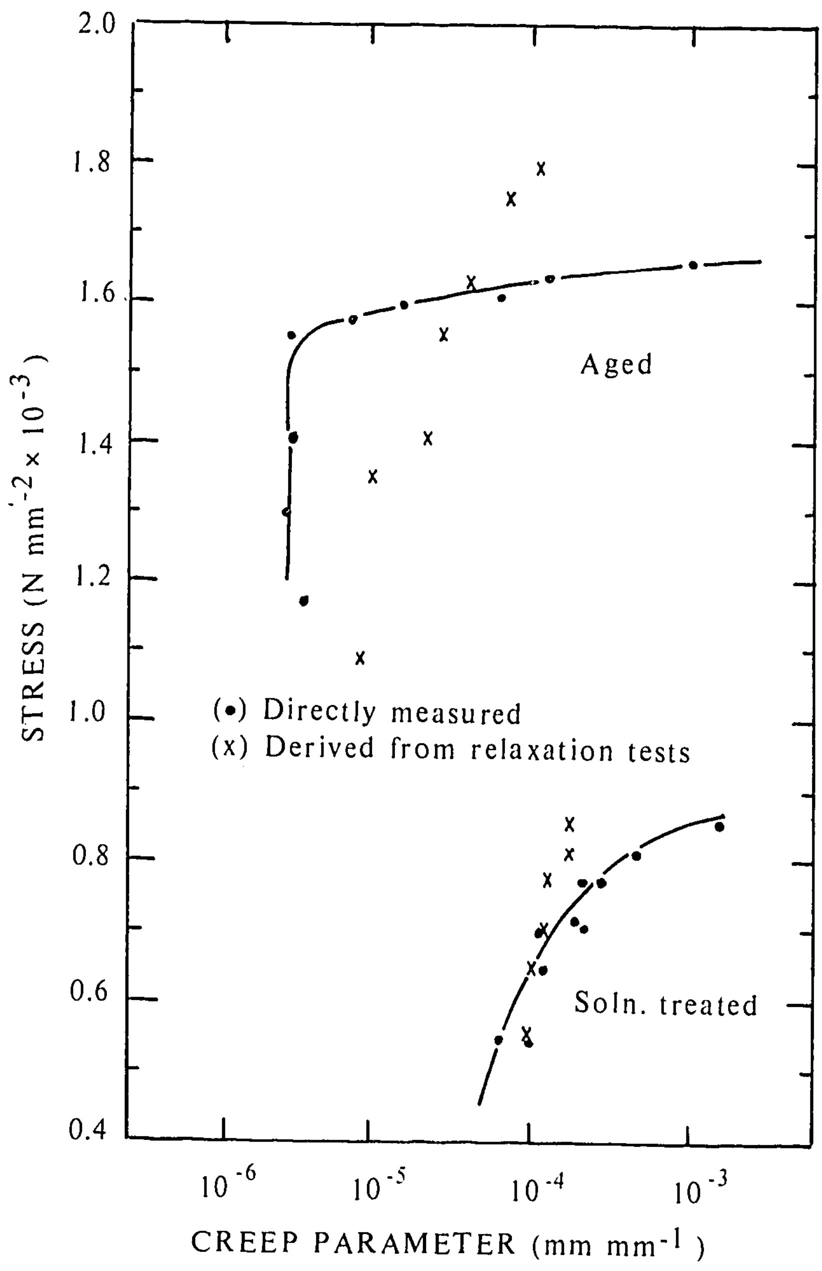

FIGURE 10. VASCOMAX MAR300. CREEP PARAMETER VERSUS STRESS

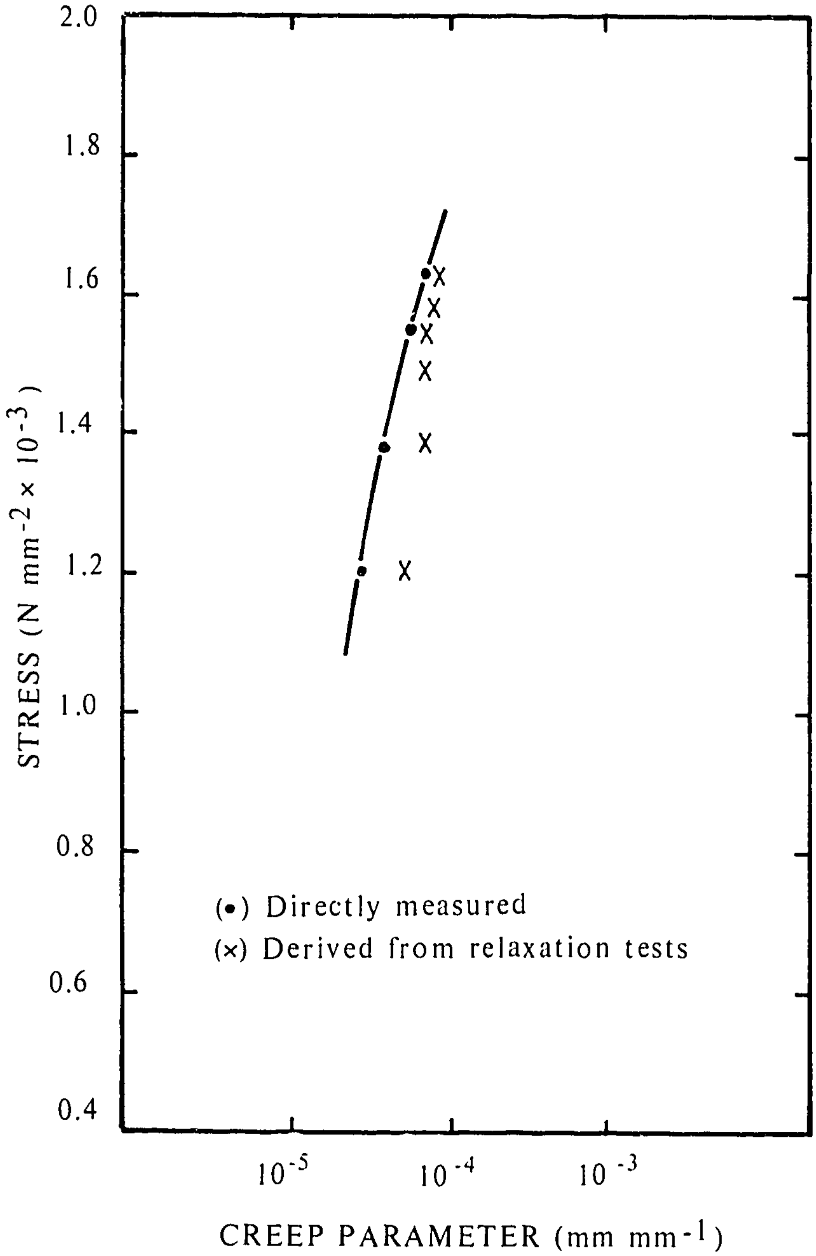

FIGURE 11. TYPE 301 STAINLESS STEEL. CREEP PARAMETER VERSUS STRESS

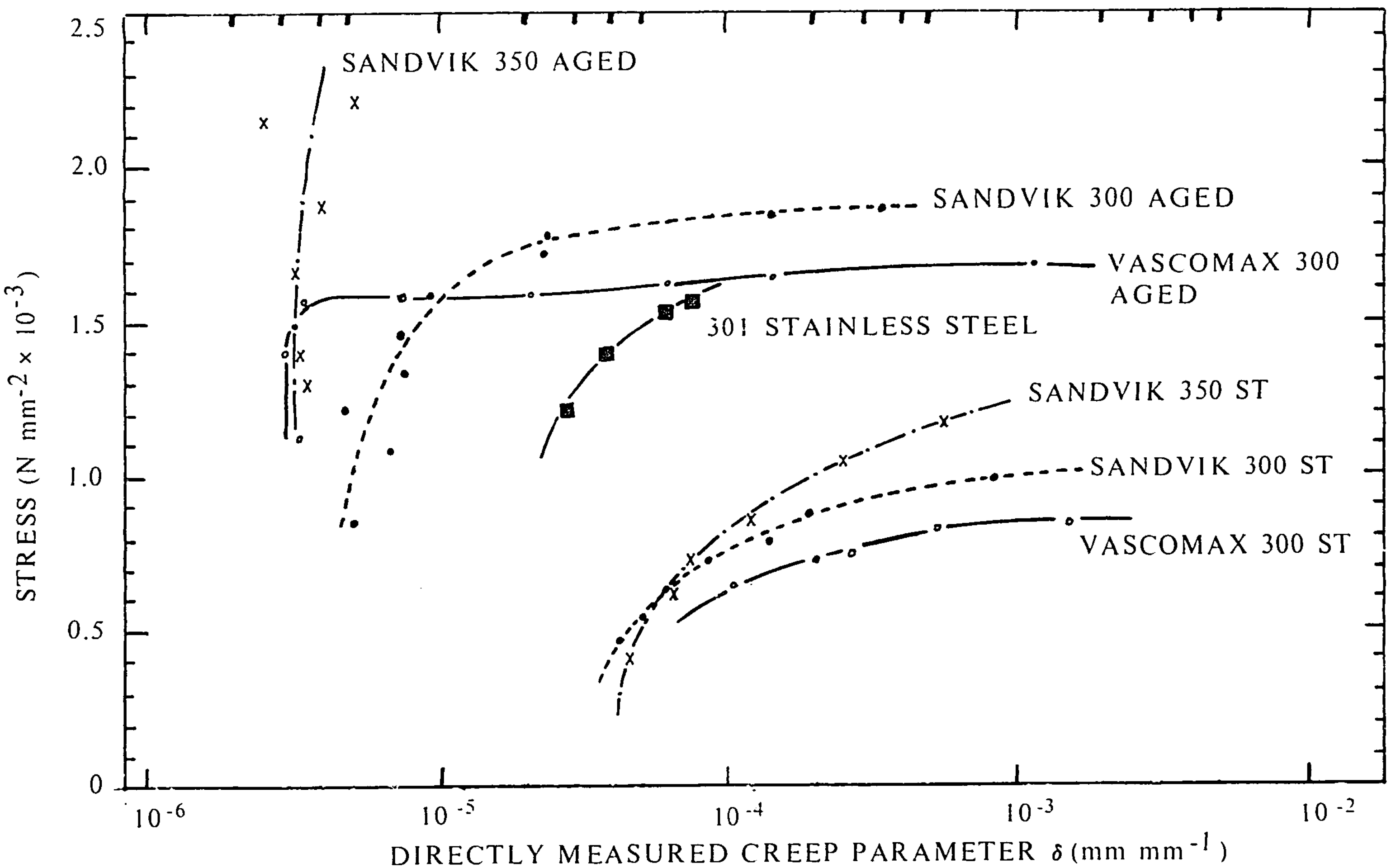

FIGURE 12. DIRECTLY MEASURED CREEP PARAMETERS AS FUNCTION OF STRESS

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