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**Modelling and Analysis of Creep Deformation and
Fracture in a 1 Cr ½ Mo Ferritic Steel**

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ABSTRACT

A quantitative model, based upon a proposed new mechanism of creep deformation in particle-hardened alloys, has been validated by analysis of creep data from a 13CrMo 4.4 (1Cr ½ Mo) material tested under a range of stresses and temperatures. The methodology that has been used to extract the model parameters quantifies, as a first approximation, only the main degradation (damage) processes - in the case of the 1Cr ½ Mo steel, these are considered to be the parallel operation of particle-coarsening and a progressively increasing stress due to a constant-load boundary condition. These "global" model parameters can then be modified (only slightly) as required to obtain a detailed description and "fit" to the rupture lifetime and strain/time trajectory of any individual test. Examples are given of the procedure used and the results of such detailed data-fitting, with and without the introduction of an additional damage mechanism - specifically in the present work, grain boundary cavitation. The global model parameter approach may be thought of as predicting average behaviour and the detailed fits as taking account of uncertainties (scatter) due to variability in the material. Using the global parameter dataset, predictions have also been made of behaviour under biaxial stressing; constant straining rate; constant total strain (stress relaxation) and the likely success or otherwise of metallographic and mechanical remanent lifetime procedures.

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1 INTRODUCTION

Much of the high temperature power plant that was commissioned in the 1960's is now approaching its design life and considerable effort has been expended in recent years to assess the remanent life of critical components in this plant. Techniques that have been used include

- metallographic inspection (1) and
- mechanical testing of material removed from the plant (2)

Various analysis methods (3-6) have been used to interpret these data but confidence in their predictive ability is restricted due to the lack of validated constitutive equations describing the creep behaviour of the component materials.

In a collaborative exercise within the framework of COST501, creep tests were performed in a number of laboratories on three important ferritic steels (13CrMo44, 14MoV63, 13CrMo910) over a range of temperatures and stresses. The purpose of this exercise was to produce a database for materials relevant to the power generation industry that could be used to assess a) various remanent lifetime techniques and b) predictive models of creep behaviour.

This report presents a new physics-based model of creep deformation in particle-hardened alloys which predicts creep parameters such as rupture life, minimum creep rate, Monkman-Grant constant and various measures of remanent lifetime. The model is fitted to data obtained on 1Cr ½Mo (13CrMo44) ferritic steel to establish a model parameter database for prediction of these creep properties for this material.

2 MODEL DEVELOPMENT

2.1 Physical Basis for the Model

Ferritic steels of the type relevant to this work, are strengthened primarily by precipitates of carbides of the minor alloying elements. At high temperatures, coarsening of these precipitates is believed to result in a progressive loss in creep resistance of the material. The mechanism causing such weakening has previously been assumed to be a consequence of creep being controlled by a threshold stress which decreases with increased particle spacing (3,4). The results of the current COST 501 exercise could not be explained using this theory but success has been achieved using a new mechanism of deformation, which has been incorporated into a physically-based tertiary damage model. The continuum damage mechanics framework, within which the model has been developed, describes the evolution of strain-rate during creep as a function of a number of physically-based state variables, the evolution rate of each being a function of some unique combination of stress, temperature and the state variables.

2.1.1 New creep mechanism and constitutive equation

Dyson (7) has proposed a new mechanism of creep in particle-hardened alloys such as ferritic steels whereby dislocations by-pass particles by a process of "diffusive-drift". The essential new idea is that the drift process of climb and glide occurs as a parallel rather than a sequential process. Although simple in concept, it leads to predictions of the slopes of logarithm of minimum strain-rate as a linear function of stress which agree well with those found experimentally in ferritic steels and also nickel-base superalloys. The monotonic decrease with decreasing stress of the so-called Norton stress exponent, n , is a natural consequence of the Reaction Rate Theory formulation, but the important new consequence

is the role played by the particle spacing. It no longer operates as a threshold stress but, as will be seen below, acts in a manner that allows creep at any finite stress, provided that there is no other source of threshold stress.

NOMENCLATURE

b	Burger's vector
D_0	Pre-exponential constant for volume diffusion
H	State variable controlling primary creep
H^*	Saturation value for H
h'	Hardening coefficient for primary creep
$\bar{H}$	Loading mode sensitivity switch for cavitation
k	Boltzmann's constant
k'	Coefficient for tertiary creep due to cavitation
K'	Coefficient for time evolution of state variable, S
l, l_i	Precipitate spacing (subscript 'i' denotes initial value)
$\bar{L}$	Two dimensional Orowan spacing for precipitates
Q_v	Activation energy for volume diffusion
Q_s	Heat of solution of carbon in iron
Q_c	Activation energy for particle coarsening ($Q_c = Q_v + Q_s$)
R	Gas constant
S	State variable for tertiary creep controlled by particle coarsening $\left(S = 1 - \frac{l_i}{l} \right)$
$\bar{S}_{ij}$	Deviatoric stress tensor
T	Absolute temperature
V_p	Volume fraction of precipitates
α	Geometric term for dislocation glide
β	Loading-mode dependent parameter relating maximum principal to effective strain rate
γ	Loading-mode dependent parameter relating maximum principal stressing rate to effective straining rate
γ	Loading-mode dependent parameter relating effective to maximum principal stress
ϵ_{fu}	True uniaxial failure strain
$\dot{\epsilon}$	Current true uniaxial strain rate
$\dot{\epsilon}_i$	Initial true strain rate
$\dot{\epsilon}_{ij}$	True strain rate tensor
$\dot{\epsilon}_0$	Stress independent coefficient for true strain rate
$\dot{\epsilon}_e$	True von Mises equivalent strain rate
$\dot{\epsilon}_m$	True minimum strain rate
μ	Shear modulus
ν	Stress state sensitivity index for cavitation
ρ_m	Mobile dislocation density
σ	Applied uniaxial stress
σ_{ij}	True stress tensor
σ_0	'strength' parameter in creep rate equation
σ_e	von Mises equivalent stress
σ_{or}	Orowan stress for particle bypass
σ_1	Maximum principal stress
ω	State variable for creep-constrained cavitation
t_f	Rupture lifetime
C_m	Monkman-Grant parameter ($= \dot{\epsilon}_m t_f$)
λ	Damage tolerance parameter (ϵ_{fu}/C_m)

The initial creep rate, $\dot{\epsilon}_i$, of a dispersion of volume fraction V_p and interparticle spacing, l , is given by

$$\dot{\epsilon}_i = \dot{\epsilon}_o \sinh \frac{\sigma}{\sigma_o} \quad (1)$$

where $\sigma_o = 2\sqrt{3}kTf(V_p)/b^2\ell$ (2a)

or $\sigma_o = \frac{2\sqrt{3}kTV_p^{1/2}\sigma_{or}}{\alpha\mu b^3}$ (2b)

and $\dot{\epsilon}_o = \frac{4\sqrt{3}kT\rho_m D_o \exp\left[-\frac{Q_v}{RT}\right]}{b^3\sigma_o}$ (2c)

The Orowan stress, σ_{or} , has been approximated by $\alpha\mu b/\bar{L}$ where $\bar{L}$ is the 2-Dimensional Orowan spacing. The parameter V is taken to be 2.5 from the work of McLean (8), who correctly reformulated previous work of Lagneborg (9) and Evans and Knowles (10) to calculate the effective stress acting on a dislocation undergoing climb and glide as $0.4 (=1/2.5)\sigma$. Thus

$$\sigma_o = \frac{5kTV_p^{1/2}\sigma_{or}}{\alpha\mu b^3} \quad (3a)$$

and $\dot{\epsilon}_o = \frac{10kT\rho_m D_o \exp\left[-\frac{Q_v}{RT}\right]}{b^3\sigma_o}$ (3b)

The evolution of the interparticle spacing (and therefore σ_{or}) is calculated from standard coarsening theory using the state variable $S = 1 - \frac{\ell_i}{\ell}$. Thus from Eq (1), the strain rate $\dot{\epsilon}$, neglecting primary creep, is given by

$$\dot{\epsilon} = \dot{\epsilon}_o \sinh \frac{\sigma}{\sigma_o(1-S)} \quad (4)$$

with $\dot{S} = \frac{K'}{3} (1-S)^4$ (5)

where K' is a rate function that depends exponentially on temperature and inversely on ℓ_i^3 .

Primary creep has been modelled here in a manner similar to, but not the same as, that given by Ion et al (11), using the state variable $H(= \sigma_i/\sigma)$. Similarly, cavitation can be introduced

as described previously using the state variable ω . Thus the proposed new constitutive and damage accumulation equations are given by the following equation-set (6).

$$\begin{aligned}
 \dot{\epsilon} &= \dot{\epsilon}_0 \sinh \left[\frac{\sigma(1-H)}{\sigma_0(1-S)(1-\omega)} \right] \\
 \dot{H} &= \frac{h'\dot{\epsilon}}{\sigma} \left[1 - \frac{H}{H^*} \right] \\
 \dot{S} &= \frac{K'}{3} (1-S)^4 \\
 \dot{\omega} &= k'\dot{\epsilon}
 \end{aligned}
 \tag{6}$$

The parameter, k' , has a maximum value of $\frac{1}{3\epsilon_{fu}}$ where ϵ_{fu} is the uniaxial strain to failure and cavitation is fully constrained.

In constant load tests, the stress evolves with strain and thus, under constant volume deformation, there is the additional constraint that:

$$\dot{\sigma} = \sigma \dot{\epsilon} \tag{6a}$$

Thus creep under constant load (or stress) and temperature of a material undergoing primary creep followed by tertiary creep due to thermal coarsening and cavitation can be completely characterised by the parameters $\dot{\epsilon}_0$, σ_0 , h , H^* , K' , and ϵ_{fu} .

The model may be extended to apply to multiaxial stress-states by modification of equation set (6) to become:

$$\begin{aligned}
 \dot{\epsilon}_{ij} &= \frac{3}{2} \left(\frac{\bar{S}_{ij}}{\sigma_e} \right) \dot{\epsilon}_e \\
 \dot{H} &= \frac{h' \dot{\epsilon}_e}{\sigma_e} \left[1 - \frac{H}{H^*} \right] \\
 S &= \frac{K'}{3} (1-S)^4 \\
 \dot{\omega} &= \frac{1}{3\epsilon_{fu}} \left(\frac{\sigma_1}{\sigma_e} \right)^v \bar{H} \dot{\epsilon}_e
 \end{aligned}
 \tag{7}$$

where:

σ_e is the von Mises effective stress; $\dot{\epsilon}_{ij}$ the deviatoric strain rate tensor; $\bar{S}_{ij}$ the deviatoric stress tensor; ϵ_{fu} the uniaxial failure strain; σ_1 the maximum principal stress; v the stress state sensitivity index; and $\bar{H}$ controls the sensitivity of cavitation to the mode of loading: it takes the value 1 for $\sigma_1 > 0$ and 0 when $\sigma_1 < 0$.

$\dot{\epsilon}_e$ is given by the expression

$$\dot{\epsilon}_e = \dot{\epsilon}_0 \sinh \left[\frac{\sigma_e (1-H)}{\sigma_0 (1-S) (1-\omega)} \right]$$

In equation set (7), it is apparent that without the operation of $\dot{\omega}$, behaviour would be controlled totally by σ_e .

3 MODEL VALIDATION

One data set, 1Cr 1/2Mo (13CrMo44), was chosen from the COST 501 exercise as having the most suitable combination of tests for analysis: creep tests had been carried out on material in the as-received condition (designated NEW) and also on material that had been withdrawn from service after approximately 250000h under the nominal conditions 525°C and 110bar (designated SERVICE EXPOSED). A summary of the data is given in Table 1(a) and (b) and the lifetime data are plotted in Fig. 1 as a function of stress using logarithmic coordinates.

Table 1a Creep data (COST 501) for 1Cr ½Mo material in the as-received condition

TESTPIECE CODE	STRESS (MPa)	TEMP (°C)	FAILURE STRAIN (%)	FAILURE TIME (h)	MIN CREEP RATE ($\times 10^{-6} \text{h}^{-1}$)
BLD022	140	530	22.4	9416	1.3
BLD012	160*	530	17.2	10451	1.5
BLD011	200	530	6.0	3491	6.3
BLD019	220	530	14.3	2500	14.5
BLD004	250	530	22.3	998	61.0
BLD017	250	530	4.4 C	1822 C	14.0
BLD003	320	530	35.6	48	2720.0
BLD115	98	550	4.5	10693	0.8
BLD015	100	550	31.9	15699	1.4
BLD112	120	550	31.9	7927	1.6
BLD013	150	550	21.6	6847	1.2
BLD005	165	550	32.7	3692	3.0
BLD006	200	550	15.7	1562	17.8
BLD010	250	550	16.1	476	78.0
BLD018	250	550	20.2	432	142.0
BLD016	280	550	35.7	90	1420.0
BLD008	80	600	56.9	1345	45.6
BLD001	110	600	57.9	354	83.7
BLD020	130*	600	49.6	168	280.0
BLD007	50	625	58.9	1580	14.1
BLD002	65	625	67.6	967	39.3

* = poor data C = test continuing

Table 1b Creep data (COST 501) for 1Cr ½Mo material in the service exposed condition

TESTPIECE CODE	STRESS (MPa)	TEMP (°C)	FAILURE STRAIN (%)	FAILURE TIME (h)	MIN CREEP RATE (10 ⁻⁶ h ⁻¹)
BKA115	90*	530	3.7C	24199C	1.1
BKA112	100	530	44.8	20794	4.3
BKA209	120	530	38.4	6979	13.0
BKA013	120	530	43.4	6945	14.0
BKA022	130	530	50.5	2785	40.0
BKA002	150	530	50.9	769	162.5
BKA007	150	530	52.9	543	217.5
BKA005	175	530	51.3	136	1167.0
BKA001	175	530	58.0	93	1243.0
BKA106	70.2*	550	2.2C	26103 C	0.6
BKA101	82.1	550	41.3C	23652 C	1.9
BKA025	110	550	52.0	2777	37.0
BKA004	130	550	59.8	686	154.0
BKA003	130	550	42.1	791	124.0
BKA009	150	550	43.6	157	720.0
BKA008	150	550	42.6	190	640.0
BKA014	50	600	2.3	7602	2.0
A3014	50	600	41.7	9789	2.0
BKA005	92	600	64.9	620	106.3
BKA010	92	625	59.1	598	145.0
AB011	50	625	52.3	2285	12.1
BKA011	55	625	53.9	2229	13.3
BKA012	55	625	65.1	1966	14.7
AB010	50	650	58.6	544	58.0
AB009	50	675	83.4	133	370.0

* = poor data C = test continuing

Figure 1 illustrates the effect of prior service exposure on subsequent creep behaviour. Lifetimes of SERVICE EXPOSED (S.E) material are always less than those of NEW material at 530 °C and 550 °C, but there is a distinct convergence at the lower stresses. In marked contrast, lifetimes of NEW and S.E material are approximately identical at 600 °C and 625 °C. These lifetime similarities and differences are reflected in the shapes of creep curves, as characterised by the Monkman-Grant parameter, $C_m = \dot{\epsilon}_m t_f$, and the damage-tolerance parameter, $\lambda = \epsilon_{fu}/C_m$, Figure 2. Using the diagnostic procedure for identifying damage mechanisms proposed by Ashby and Dyson (12), it is concluded that the combination of low C_m and high λ found in Figure 2 for NEW material and the combination high C_m /low λ for S.E material can only be rationalised by assuming a dominant thermally-induced damage mechanism. This is taken to be the coarsening of transgranular carbides.

Quantifying these similarities and differences in behaviour between new and service exposed material pose a severe test for any constitutive model.

During the process of determining a self consistent set of values for the model parameters in Equation set (6), constant values were assigned to certain parameters. Specifically: $h' = 10^5$ MPa and $H^* = 0.4$

These values for h' and H^* are near to upper bound figures found through experience of fitting the primary equation to these alloy types.

Table 2 Global model parameters used for creep lifetime predictions

Temperature, °C	$\dot{\epsilon}_0$ (h ⁻¹)	σ_0 (MPa)	h' (MPa)	H^*	K' (h ⁻¹)
New Material					
530	1.3×10^{-9}	13.5	10^5	0.4	1.1×10^{-4}
550	3.5×10^{-9}	13.0	10^5	0.4	2.8×10^{-4}
600	5.0×10^{-8}	8.0	10^5	0.4	8.0×10^{-4}
625	1.7×10^{-7}	6.5	10^5	0.4	1.4×10^{-3}
Service-Exposed Material					
530	2.2×10^{-9}	8.0	10^5	0.4	2.2×10^{-5}
550	6.0×10^{-9}	7.5	10^5	0.4	5.4×10^{-5}
600	5.7×10^{-8}	7.0	10^5	0.4	5.4×10^{-4}
625	1.7×10^{-7}	6.5	10^5	0.4	1.4×10^{-3}
650	5.1×10^{-7}	5.5	10^5	0.4	2.4×10^{-3}
675	1.5×10^{-6}	4.5	10^5	0.4	3.8×10^{-3}

Values for σ_0 were determined from plots of minimum creep rate as a function of stress using the primary creep equation and are tabulated in Table 2 to the nearest 0.5 MPa. These data could be fitted to the following functions to a reasonable approximation.

For new material in the temperature range studied and for service exposed material at temperatures > 883 K:

$$\sigma_o = 8.0 \times 10^{-3} \exp \left[\frac{30,000}{RT} \right] \text{ MPa}$$

For service exposed material at temperatures > 800K but < 883K:

$$\sigma_o = 1.62 \exp \left[\frac{10,600}{RT} \right] \text{ MPa}$$

A better, physically-based representation is under investigation in order to provide a better fit at 550 and 530 °C.

The decrease of σ_o with temperature is believed to be primarily caused by the rapid decline in the yield stress (reflected in the Orowan Stress in Eq 3a) which is a well-known feature of the alloy above 550 °C. In these dilute alloys ($V_p = 2\%$), this means that the initial particle spacing $\ell_i \propto 1/\sigma_{or}$ must increase with temperature. Accordingly, in formulating the rate constant for coarsening, K' , which is inversely proportional to ℓ_i^3 , by definition, we have incorporated a term σ_o^3 , thus

$$K' \propto \sigma_o^3 \exp - \left[\frac{Q_c}{RT} \right]$$

The following equations for ϵ_o and K' as a function of temperature are proposed for both new and service exposed materials using a value of $Q_v = 260 \text{ kJ mole}^{-1}$ for volume diffusion in ferritic alloys and a heat of solution of carbon in iron, $Q_s = 40 \text{ kJ mole}^{-1}$

$$\epsilon_o = \frac{1.44 \times 10^9}{\sigma_o} \exp - \left[\frac{260,000}{RT} \right] \text{ h}^{-1}$$

$$K' = 1.4 \times 10^{12} \sigma_o^3 \exp - \left[\frac{300,000}{RT} \right] \text{ h}^{-1}$$

Table 2 reports the values of ϵ_o and K' for each temperature and material condition, along with those for σ_o and the constant values of h' and H^* . They have been termed "global" model parameters to reflect their approximate nature. The set of global parameters appropriate to each temperature and material condition has been used as input in Equation set (6) and integrated numerically, using a fifth order Runge Kutta routine with adaptive step control, to give strain/time trajectories at different stress levels. Cavitation damage was not included ($\omega = 0$) and lifetimes were determined by terminating the calculation at a specific value of strain. Rupture ductilities were taken to be 50% in all cases except for the new material at 530 °C and 550 °C where the values 20% and 30% respectively were used, to correspond with the mean rupture elongations of the data at each temperature.

3.1 Lifetime Prediction

Fig 3 shows a comparison of predicted rupture lives for new and service exposed material with the supplied data. Excellent agreement has been achieved at the higher temperatures but at 530 °C the model predictions overestimate lifetimes by a significant amount at the lower stresses. As demonstrated below, this discrepancy is due to the role of cavitation at this temperature: the data (Table 2a) show a distinct ductility trough with a minimum ductility ~6% at 200 MPa. Since the model predictions have assumed a rupture ductility of 20% and the influence of cavitation on strain trajectories and lifetime has not been included, the model would be expected to predict lifetimes greater than the actual data.

A comparison between model predictions and data for the strain/time trajectories of tests carried out on new material at 150MPa and 550°C and service exposed material at 50MPa and 625°C are shown in Figs 4 and 5 respectively. In both cases the use of the global parameters approximates the form of the creep curve, but the fit has been improved by making small independent adjustments of just two parameters, σ_0 and h , with the consequent changes in $\dot{\epsilon}_0$ and K' due to the change in σ_0 .

The most severe test of the model is to fit the strain/time behaviour of the low ductility test at 200 MPa at 530 °C. Fig 6 demonstrates the procedure that was necessary. The first stage was to adjust the global parameters, as in Figs 4 and 5, to give agreement with the initial stages of creep. This resulted in a greater discrepancy in lifetime prediction. Secondly the maximum possible influence of cavitation was introduced by using the damage rate equation:

$$\dot{\omega} = \dot{\epsilon}/3\epsilon_{fu}$$

This resulted in a predicted lifetime slightly shorter than observed, but an excellent fit was obtained by reducing the influence of cavitation (changing 1/3 to 1/5).

The primary objective of this modelling exercise was to predict lifetimes, particularly at low stresses, which is why the primary parameters h' and H^* were taken as being independent of stress and temperature. This occasionally led to a poor representation of strain trajectory with time at low fractional lifetimes, particularly at high and low stresses. Figure 6 gives one example: the primary stage is very poorly represented using the global parameters and requires h' to be adjusted by one order of magnitude while slightly increasing σ_0 . This enormous change in h' affects lifetime only slightly and so gives some justification for the use of constant values of h' and H^* for lifetime predictions in situations dominated by tertiary damage.

3.2 Prediction of Minimum Creep Rates

The comparison of experimentally determined minimum creep rates with model predictions is shown in Fig 8. Solid lines represent new material predictions and dashed lines, service exposed. The agreement is reasonable on a global scale, but detailed examination shows that at low stresses, particularly at 550 °C, there are major discrepancies. These are mirrored for the most part by the rupture data in Fig 3 (underestimates of minimum creep rate corresponds with overestimates of lifetime). In view of this, it seems unlikely that the discrepancies in predicted and measured values of minimum creep rate are due solely to the use of constant values of h' and H^* . Rather, they are likely to reflect scatter in material properties, as will be discussed in section 4.4.

4 MODEL PREDICTIONS

The model represented by Equation set 6 and shown in Section 3 to give an adequate description of creep deformation and fracture of 1 Cr ½ Mo steel, using the COST 501 database, has been extended in the following to predict other aspects of material behaviour.

4.1 Monkman-Grant and Damage Tolerance Parameters

The Monkman-Grant parameter ($C_m = \dot{\epsilon}_m t_f$) and the damage-tolerance parameter ($\lambda = \epsilon_{fu}/C_m$) are useful for diagnosing the tertiary damage mechanisms operating in engineering alloys (12). Different mechanisms lie in restricted fields of a map of C_m as a function of λ although, since more than one mechanism can occupy a given field, an unambiguous identification of the primary damaging mechanism cannot be achieved using this technique alone. The underlying physics of Equation set 6 requires that the primary damage mechanism at high stress is that due to a progressively increasing stress (low λ and $C_m \rightarrow \sigma_o/\sigma$) while at low stress, the primary damage mechanism will be due to particle-coarsening (high λ , low C_m). Figure 9 shows predicted creep curves at 300 and 150 MPa using the 530 °C model parameters and illustrates the preceding point. At intermediate stress levels, the two damage mechanisms produce tertiary curves with shapes that vary continuously with stress.

4.2 Influence of Grain Boundary Cavitation

It has already been demonstrated in a specific way in Section 3 how creep-constrained cavitation can interact strongly with strain rate and change the shape of creep curves: Fig 10(a) illustrates this in a more systematic way using the 550 °C model parameters for new material at 200 MPa. At this stress and temperature, the dominant tertiary mechanism is the increasing stress due to the constant load boundary condition. As the stress level approaches that found in service, the coarsening mechanism dominates and this has the effect of restricting the ability of cavitation to reduce lifetimes, Fig 10(b).

4.3 Influence of Prior Thermal Exposure

One of the most important pieces of information in the COST 501 programme, from the viewpoint of modelling, was the creep behaviour of material previously exposed to a low stress (~ 10 MPa) for 250000 h at 525 °C. Since the prior creep strain of this service exposed material was likely to be negligible (it was not reported), any degradation recorded in subsequent creep behaviour must be a consequence of thermal events alone. The present model treats this as being due to particle-coarsening and has satisfactorily accounted for the COST dataset. There is not total agreement on all counts since the numerical values used in the coarsening rate equation require that the service exposure temperature should be ~ 50 °C lower than reported. This may be the case of course, since temperature monitoring is not the usual practice and there may have been errors of this magnitude. Alternatively, when an optimised parameter set is available, better agreement may be achieved.

The softening effect of prior service exposure in this steel was reported 30 years ago by Toft and Marsden (13). Their results are shown in Fig 11 and can be compared with Fig 12 which shows the predictions made by the present model at the same temperature of 565 °C. Although not exact, the agreement provides strong further evidence that the new model adequately represents creep behaviour in this material.

4.4 Prediction of behaviour under constant strain rates and constant total strain (stress relaxation)

Provided that constitutive and damage evolution equations such as Equation set 6 have been validated, they can be used to predict behaviour under a variety of imposed boundary conditions such as variable stress and/or temperature, constant total strain rate and constant total strain. Examples of the latter two types of test are shown in Figs 13-15. Comparison of Fig 13 with Fig 8 shows that the peak tensile stress at a given straining rate corresponds closely to the stress associated with a minimum creep rate in a creep test. The marked reduction in stress at the lower of the two straining rates is due to particle-coarsening and corresponds to tertiary creep in a creep test.

Fig 14 demonstrates the effect of initial stress on stress trajectory up to 25000 h in a stress relaxation test. It is worth noting that the stress levels differ by only about 10% after this time interval. Fig 15 demonstrates the influence of particle-coarsening on stress trajectory: curve (b) has been calculated under the assumption of no coarsening while in (a) K' has been set to the value appropriate to 530 °C. Clearly the higher the temperature, the more important particle-coarsening will be in determining stress relaxation. A corollary is that it is the primary parameters that are of major importance at the lower temperatures (≤ 530 °C) and so the details of the present results using global primary parameters should be viewed with caution. More precise predictions would require an "adjusted" database as described in Section 3.

4.5 Remanent Lifetime Methods

Remanent Lifetime techniques can be classified as either mechanical or metallographic. A brief account will be given of the predictions made by the present model and the prognosis for successful implementation.

4.5.1 Mechanical

One possibility that has just been explored in the COST 501 programme is to measure hardness as a function of life fraction consumed. The present model can only relate to hardness if it is assumed to be linearly proportional to yield stress. In the model, σ_0 is proportional to the Orowan stress which is proportional to the yield stress, however defined. Thus the model can predict the decline in yield stress as a function of life fraction, Fig 16. There is no unique relationship between yield stress and life fraction consumed - as is apparent from a cursory inspection of the underlying physics of Equation set 6. It is recommended that such a technique should not be used in a quantitative way for lifetime monitoring.

Another method that has been proposed to assess remanent life is to take out a sample of material from the critical component and test it in the laboratory at either the same temperature but higher stress, or at the same stress, but at a higher temperature. Predictions have been made using the present model of the likely outcome of such tests. The service conditions were assumed to be 50 MPa at 550 °C which gives a computer lifetime of 118237 h. Interruptions at various fractions of lifetime were made and simulations of laboratory creep tests made in the computer at either 50 MPa/625 °C or 200 MPa/550 °C. The conditions of respective temperature and stress were chosen to give a non-damaged lifetime of approximately 2000 h, which was regarded as being a reasonable timescale for remanent life work. The results are shown in Fig 17(a) and (b).

Neither method was a complete success but accelerated stress testing was a disaster, Fig 17(b). Its only redeeming feature was that it was conservative: it predicted zero life remaining in the service exposed material after less than 10% of actual life consumed! In

contrast, the predictions for accelerated temperature testing were always non-conservative: only when the service life expired was around 50% did the predictions come within 10% of actual.

It is predicted by the model that better (but still non-conservative) estimates of remanent life would be made by the accelerated temperature method if lower temperatures (longer lifetimes) were to be used.

4.5.2 Metallographic

Remanent lifetime assessments using metallographic techniques have concentrated on monitoring either the development of cavitation or some aspect of the carbide dispersion - usually the spacing. The parameter being monitored with any technique should ideally be a single valued function of life fraction, t/t_f , and in practice should deviate as little as possible from this ideal. The present model predicts that monitoring carbide spacing has nothing to recommend it, as far as stress excursions are concerned. The reasons being the same as those discussed for stress-enhanced mechanical testing, Fig 17(b). If only temperature excursions occurred in plant, then carbide monitoring could be used in principle: even then the problem would be to standardise the technique at the service stress.

An initial assessment has been made of the possibility of successfully monitoring the cavitation. Fig 18 plots the damage parameter, ω , (fraction of boundaries cavitated) against t/t_f for different stress levels, giving lifetimes ranging from 10^3 - 10^5 h. The calculations used the 550 °C model parameters and the material was assumed to fail at $\epsilon_{fu} = 10\%$. The figure given for experimental resolution was based on NPL's experience. From 5×10^3 to 10^5 h, the monitoring of cavitation damage would appear to be feasible *provided* ϵ_{fu} remained constant. In practice, this is rarely true: Fig 19 shows the effect of increasing ϵ_{fu} to 20% and Fig 20, the effect of varying ϵ_{fu} between 2.5 and 20% for a constant life of 10^5 h. A much more comprehensive assessment would be required before a conclusion could be drawn on a specific alloy system.

4.6 Multiaxial Loading

Many of the critical components in power generation plant that are made from ferritic steels are subject to multiaxial stressing - most frequently triaxial and induced by a combination of loading and geometry. The stress fields are thus statically-indeterminate and require a combination of finite element analysis and material models such as Equation-set 7 to predict lifetimes with precision. Historically, laboratory data have been mainly acquired under statically-determinate stressing - uniaxial tension, torsion and thin-walled tubes under internal pressure are some examples. To predict component lifetimes in service under statically-indeterminate stressing, a combination of the reference stress technique with a "multiaxial stress rupture criterion", determined from the laboratory tests, has been used as an approximate method. However the determination of the "multiaxial stress rupture criterion" from laboratory data is fraught with difficulties, as Hurst (14) has recently pointed out. Uniaxial tension tests and internal pressure tests are usually conducted in such a manner that the material experiences an increasing stress as strain accumulates, whereas a torsion test is under constant stress (albeit with principal stress axes that rotate with strain). If strains to failure are large, Hurst has argued that this may lead to misinterpretation of the "multiaxial stress rupture criterion". The point may be demonstrated using Equation set 7, re-written in a way more appropriate for uniaxial/biaxial loadings:

$$\begin{aligned}
 \varepsilon_1 &= \beta \varepsilon_e \\
 \dot{H} &= \frac{h' \varepsilon_e}{\sigma_e} \left[1 - \frac{H}{H^*} \right] \\
 \dot{S} &= \frac{K'}{3} (1-S)^4 \\
 \dot{\omega} &= \frac{1}{3\varepsilon_{fu}} \left[\frac{\sigma_1}{\sigma_e} \right]^v \bar{H} \varepsilon_e \\
 \varepsilon_e &= \varepsilon_o \sinh \left[\frac{\delta \sigma_1 (1-H)}{\sigma_o (1-S)(1-\omega)} \right] \\
 \sigma_1 &= \gamma \sigma_1 \varepsilon_e
 \end{aligned}
 \tag{8}$$

The parameters β , γ , δ , σ_1/σ_e and $\bar{H}$ are given in Table 3 for constant load/stress uniaxial tension; thin-walled torsion; and constant load/stress thin-walled tubes under internal pressure with no end-load.

Table 3: Values of parameters for use in equation-set (8) for 3 different loading modes

Loading mode	Parameters				
	β	γ	δ	σ_1/σ_e	$\bar{H}$
Uniaxial tension					
1) Constant load	1	1	1	1	1
2) Constant stress	1	0	1	1	1
Thin-walled torsion	$\frac{\sqrt{3}}{2}$	0	$\sqrt{3}$	$\frac{1}{\sqrt{3}}$	1
Internal pressure with no end-load					
1) Constant load	$\sqrt{3}$	$\sqrt{3}$	$\frac{\sqrt{3}}{2}$	$\frac{2}{\sqrt{3}}$	1
2) Constant stress	$\frac{\sqrt{3}}{2}$	0	$\frac{\sqrt{3}}{2}$	$\frac{2}{\sqrt{3}}$	1

As noted previously, Equation set 7 (and 8) will predict that lifetimes are a unique function of the von Mises stress, σ_e , if $\dot{\omega} = 0$ and testing is carried out under constant stress. This is demonstrated in Fig 21a, using the model parameter database for the 1Cr ½ Mo steel at 530 °C. A constant failure criterion of maximum principal strain, ϵ_1 , of 40% was used and numerical integration performed under constant stress tension, thin-walled torsion and constant "stress" internally-pressurised tubes with no end-load, using the parameters given in Table 4 as appropriate. In Fig 21b, the same numerical lifetime data are plotted as a function of the maximum principal stress. On this basis, lifetimes are longest under constant "stress" internal pressure and shortest under torsion. Clearly, if tests in the laboratory were to be conducted under constant stress conditions, this "material" could be categorised unambiguously as a "von-Mises type".

It is usual practice to test under constant load tension and constant "load" internally-pressurised tubes (ie the pressure within the tube remains constant as strain accumulates). Equation set 8 has been integrated using the appropriate parameters listed in Table 4 and the lifetime data plotted against σ_e in Fig 22a and against σ_1 in Fig 22b. Laboratory tests conducted under these conditions would thus lead to an incorrect assessment of the material's "multiaxial stress rupture criterion", confirming the views of Hurst (14). The numerical data given in Figs 22a and b suggest some mixed stress criterion which would change with stress level and loading mode, even though the correct "multiaxial stress rupture criterion" is von Mises.

The strain trajectories with time in the three stress states indicate the dominant mechanism of damage accumulation at high and low stress levels, Figs 23a and b respectively. At 350 MPa, strain is almost a linear function of time in torsion indicating no thermal coarsening while tertiary creep under uniaxial tension and internal pressure are both caused by the increasing stress. Although the initial maximum principal (hoop) strain rate, $\dot{\epsilon}_1$, for the pressurised tube test is less than that for uniaxial tension, the increase in stress is greater in the former (see γ values in Table 4) and causes a transition in strain accumulation at approximately 4%, resulting in the reduced life, Fig 23a. At 60 MPa, Fig 23b, the torsion curve illustrates the importance of the particle-coarsening damage in controlling lifetime, while again the shorter life in the pressurised tube test is a consequence of the *additional* damage caused by the different rates of increase of stress.

5 CONCLUDING REMARKS

A new mechanism and constitutive model of creep in particle-hardened materials has been shown to describe the creep deformation and fracture behaviour of a 1 Cr ½ Mo steel database generated in the COST 501 programme. The model parameters database was arrived at by using well accepted values for certain of the variables (for example the activation energy for volume diffusion) and determining others by fitting techniques using the strain evolution data. The term "Global" parameters has been coined to distinguish these model parameters from the "optimised" set that will be determined at a future date using optimising mathematical routines. It is anticipated that the usual problems of local minima using these techniques will be overcome by putting tight bounds on the values of nearly all the parameters, as a consequence of the physics basis of the model.

The most important material parameter in the model is σ_0 : having dimensions of stress, it is a measure of the "strength" of the material and declines in value with time and temperature (not strain) as the precipitate dispersion or volume fraction changes. Only changes in the spacing of precipitates are considered in this report: at constant temperature, changes with time are calculated using the well-established equation for coarsening-kinetics, but the *initial* value of σ_0 at each temperature has had to be determined empirically. Work is in progress to establish a better physical understanding of the initial value of $\sigma_0(T)$ using

as a working hypothesis, that the volume fraction of particles decreases with temperature in the temperature range under study.

Strong additional support for the model comes from its ability to predict very closely the results obtained by Toft and Marsden (13), who demonstrated that testpieces of a material heat-treated to give vastly different creep lifetimes at high stresses, were projected to have similar creep lifetimes under service conditions.

The model, because of the way in which it has been constructed, leads naturally to predictions of behaviour under external constraints other than a constant stress/load and temperature: a number of examples are given in the text including stress relaxation, remanent life assessment and biaxial stressing.

6 ACKNOWLEDGEMENTS

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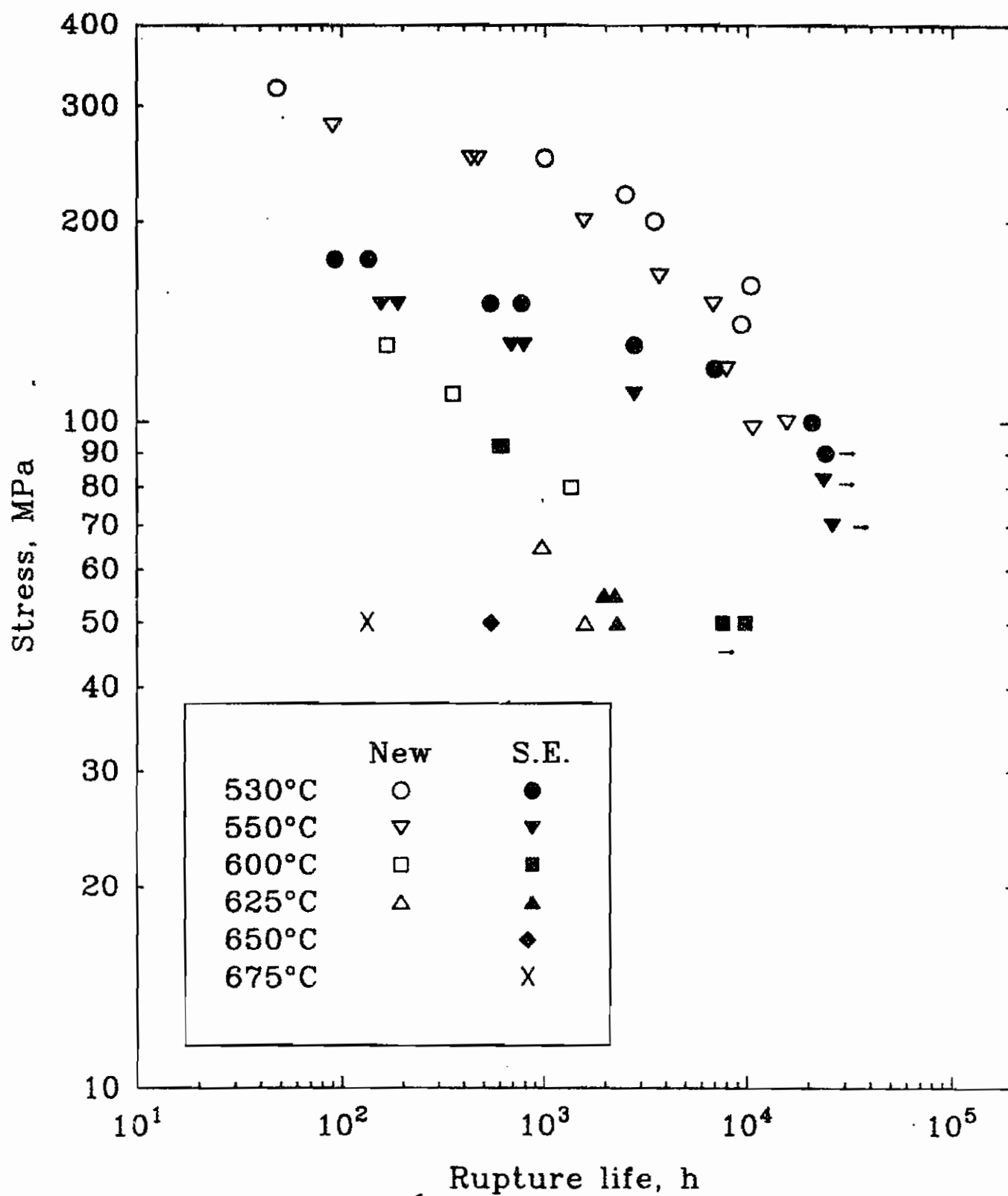


Fig 1. Creep rupture data from COST 501 programme.

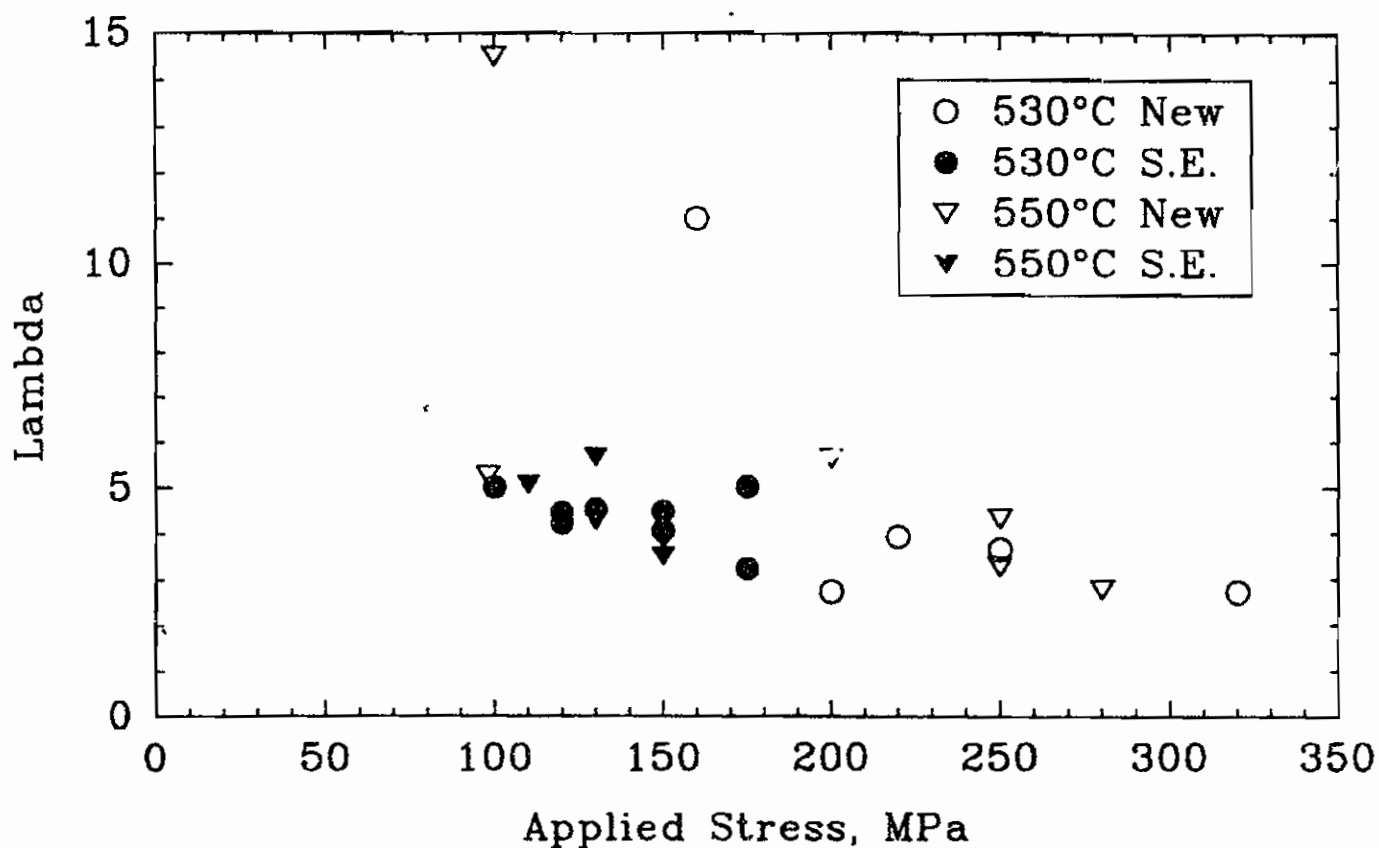
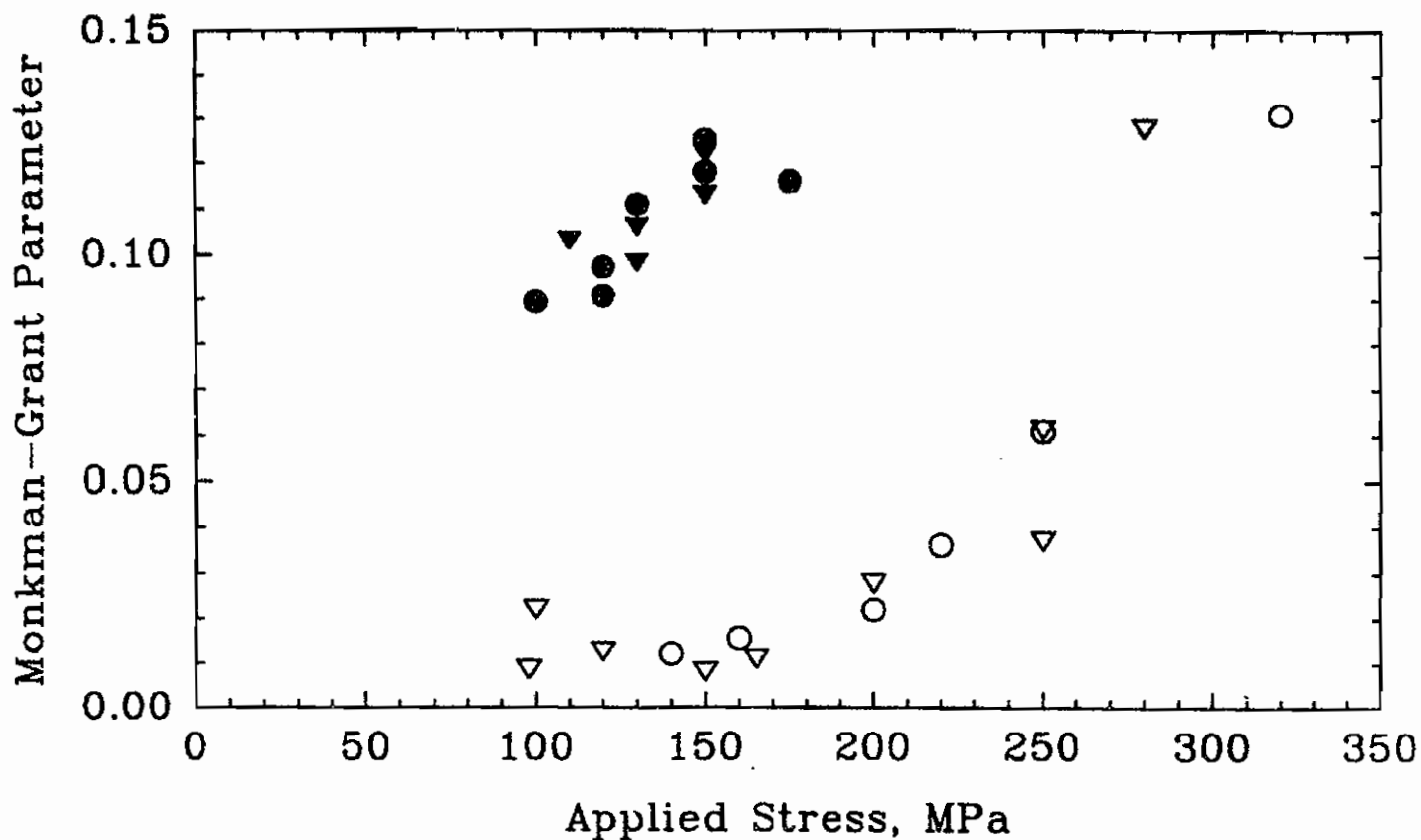


Fig. 2 Monkman-Grant and damage tolerance parameters as a function of stress and temperature for New and Service-Exposed materials at 530 and 550°C.

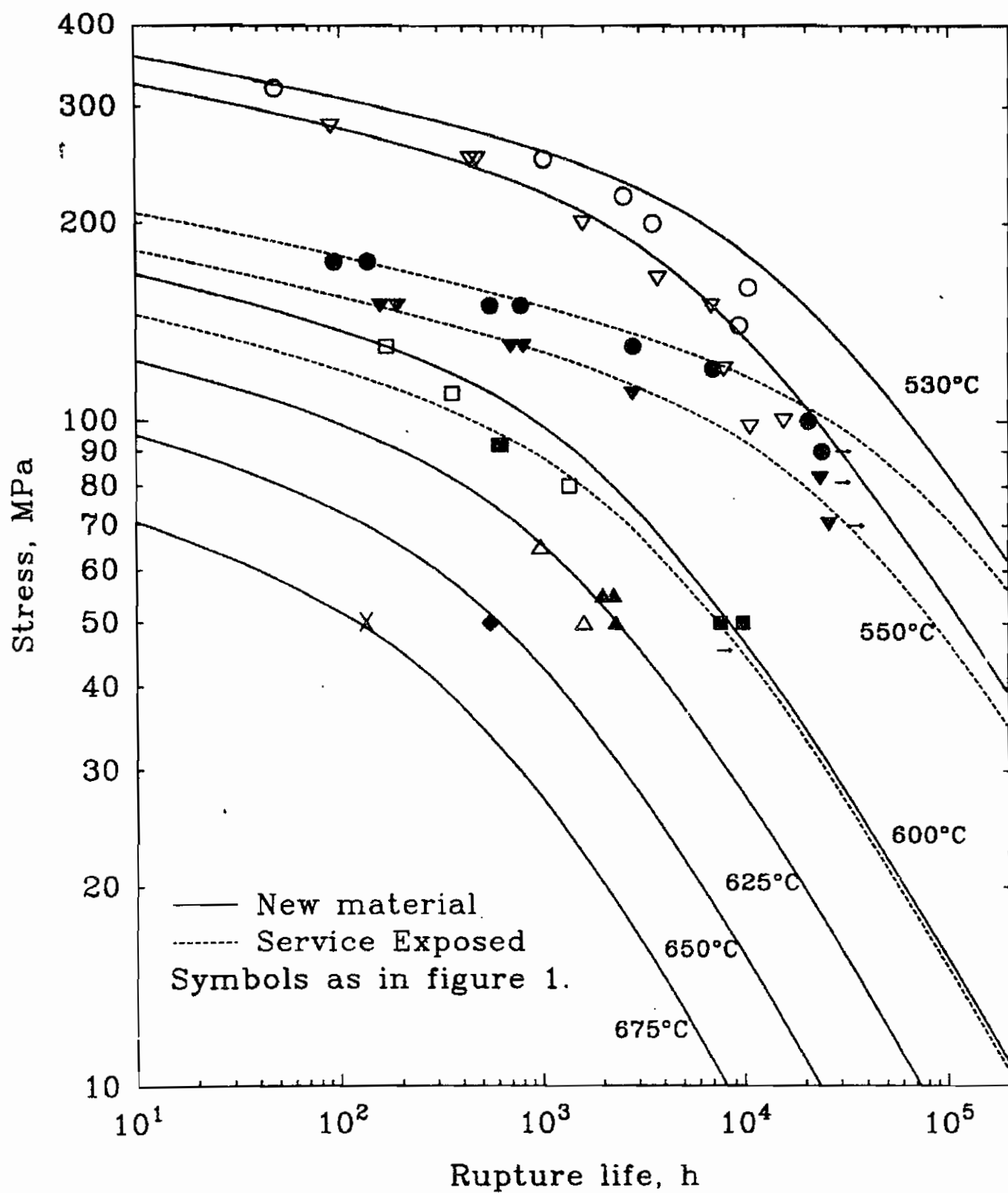
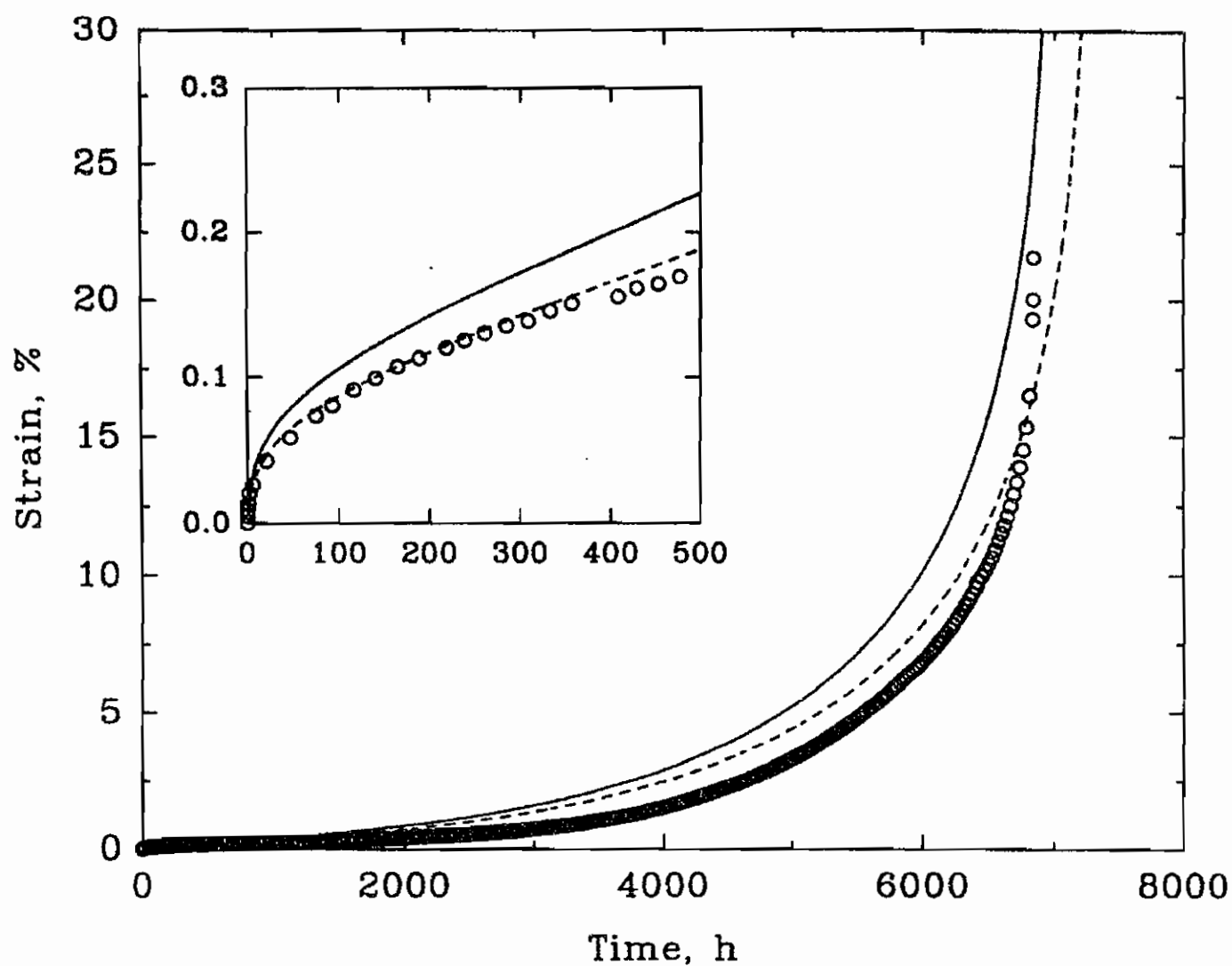
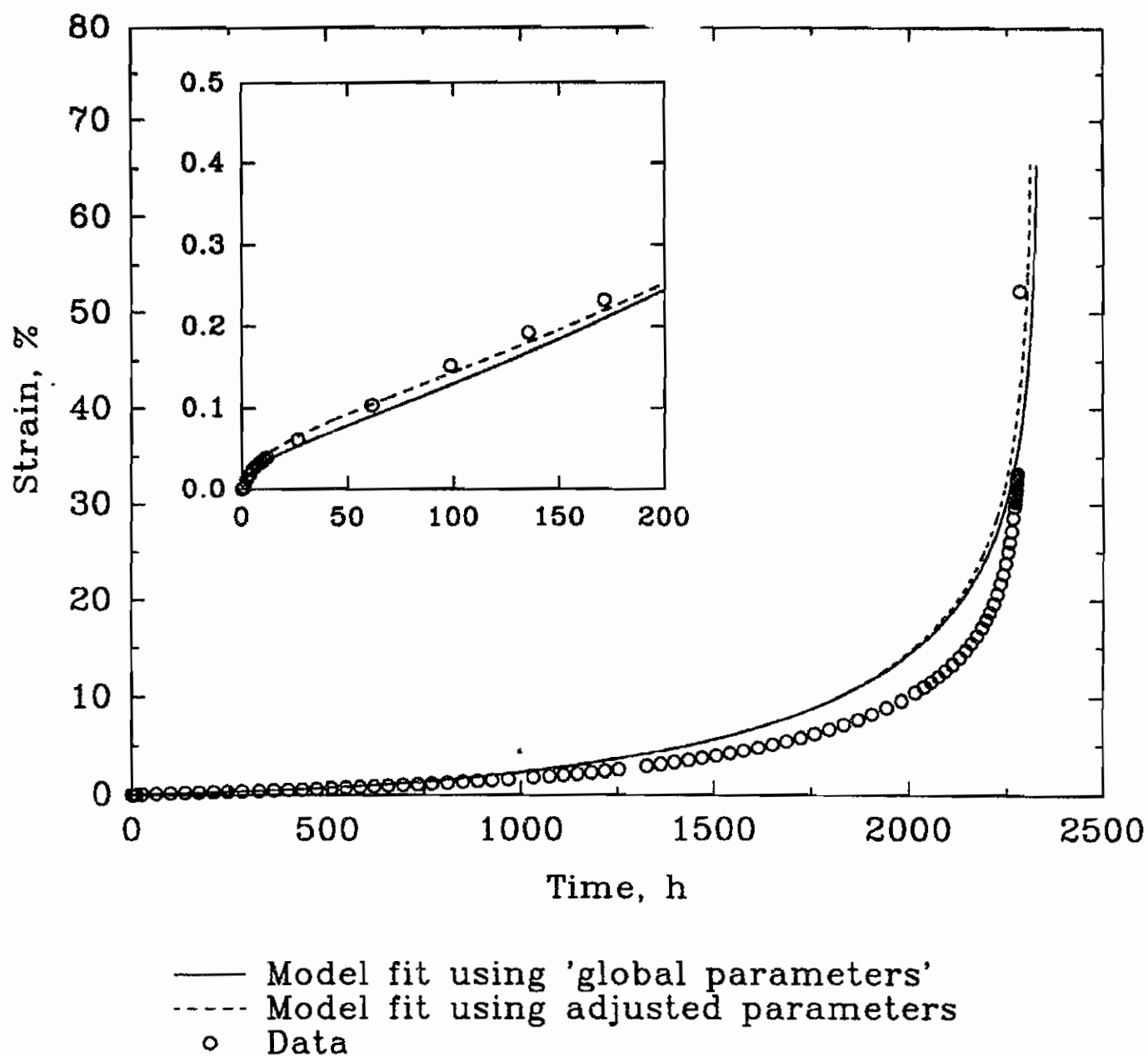


Fig 3. Comparison of Model Predictions for Creep lifetime with data.
 Model Parameters given in Table 2.



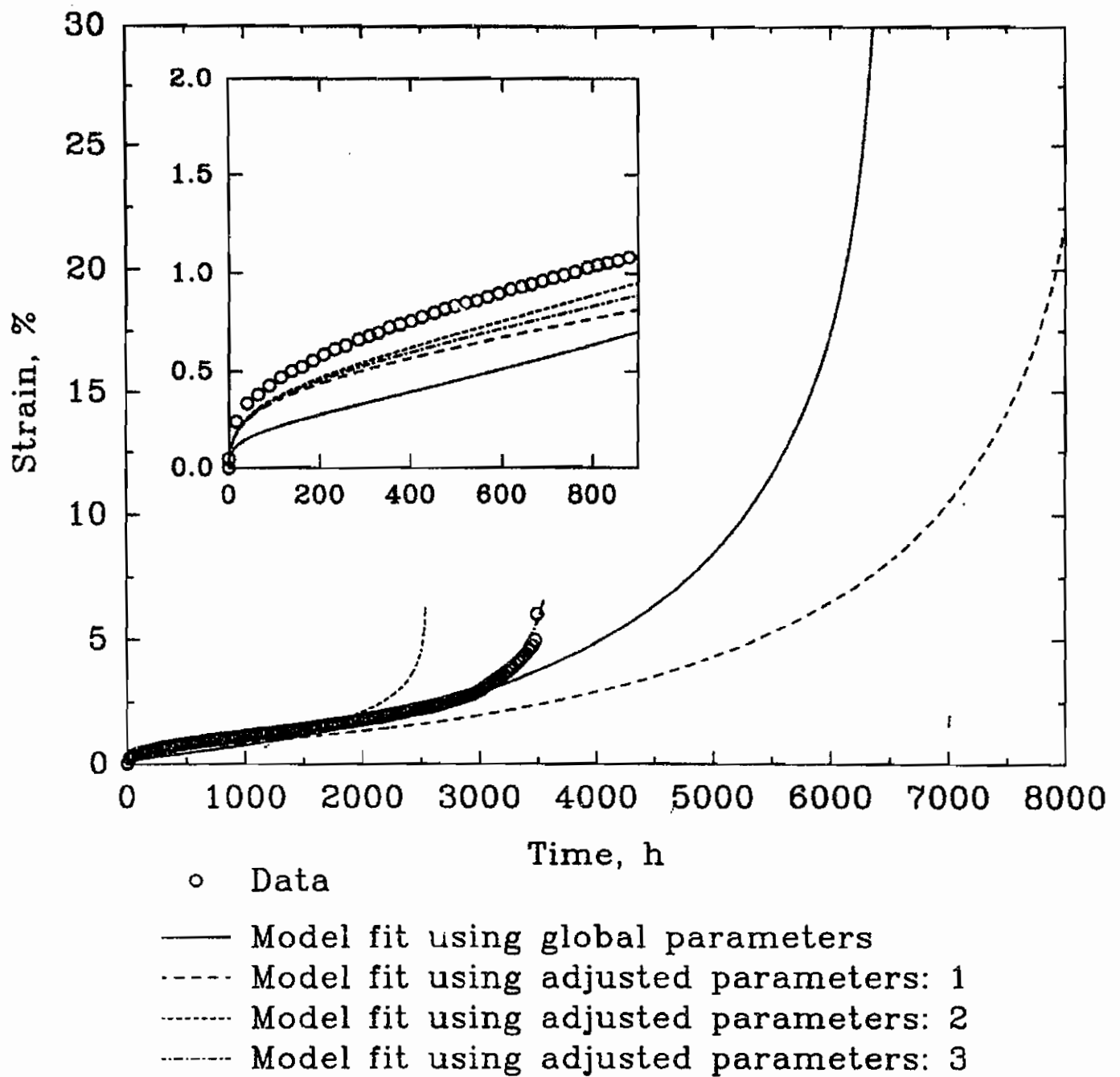
	$\dot{\epsilon}_0$	σ_0	H^*	h'	K'
Global	3.5×10^{-9}	13.0	0.40	10×10^4	2.8×10^{-4}
Adjusted	3.4×10^{-9}	13.3	0.40	12×10^4	3.0×10^{-4}

Fig 4. Comparison of model predictions and strain/time data for New material tested at 150MPa and 550°C.



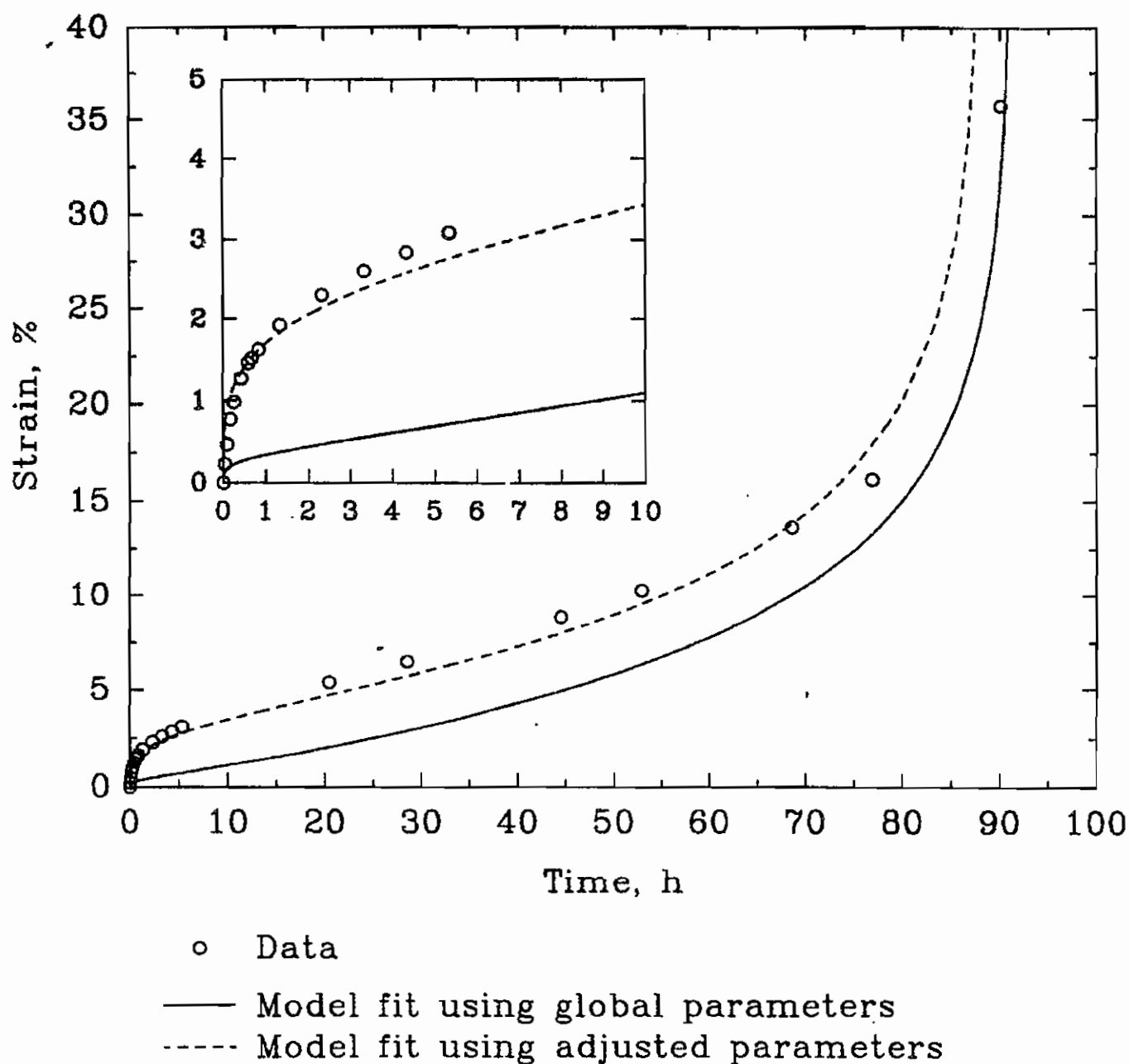
	$\dot{\epsilon}_0$	σ_0	H^*	h'	K'
Global	1.7×10^{-7}	6.5	0.40	10×10^4	1.3×10^{-3}
Adjusted	1.7×10^{-7}	6.6	0.40	6×10^4	1.4×10^{-3}

Fig 5. Comparison of model predictions and strain/time data for Service Exposed material tested at 50MPa and 625°C.



	$\dot{\epsilon}_0$	σ_0	H^*	h'	K'	Cav.
Global	1.3×10^{-9}	13.5	0.40	10×10^4	1.1×10^{-4}	0
Adjusted 1:	1.21×10^{-9}	14.5	0.40	3×10^4	1.3×10^{-4}	0
Adjusted 2:	1.21×10^{-9}	14.5	0.40	3×10^4	1.3×10^{-4}	$1/3 \epsilon_f$
Adjusted 3:	1.21×10^{-9}	14.5	0.40	3×10^4	1.3×10^{-4}	$1/5 \epsilon_f$

Fig 6. Comparison of model predictions and strain/time data for New material tested at 200MPa and 530°C.



	$\dot{\epsilon}_0$	σ_0	H^*	h'	K'
Global	3.5×10^{-9}	13.0	0.40	10×10^4	2.8×10^{-4}
Adjusted	3.4×10^{-9}	13.3	0.40	1×10^4	3.0×10^{-4}

Fig 7. Comparison of model predictions and strain/time data for New material tested at 280MPa and 550°C.

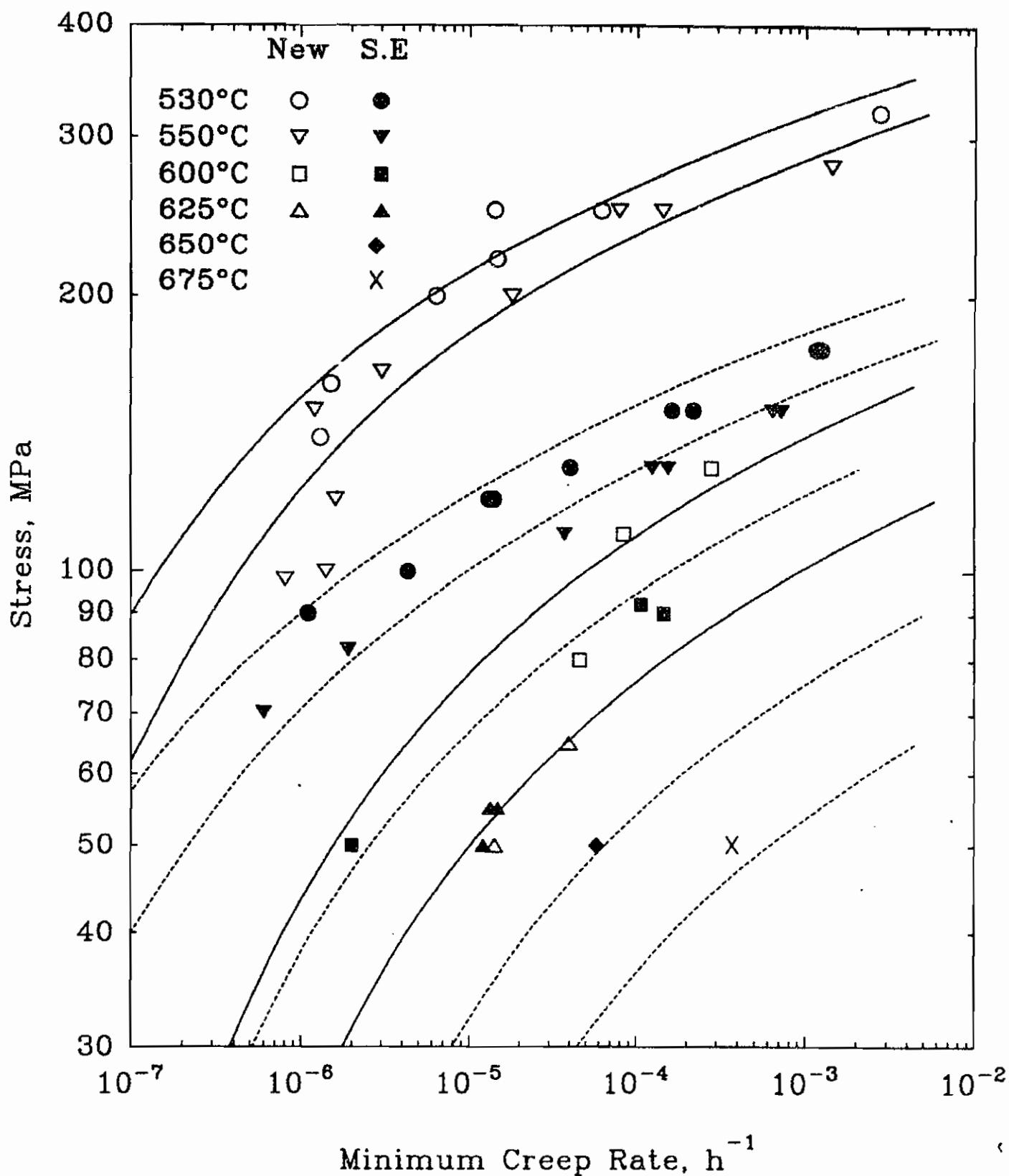
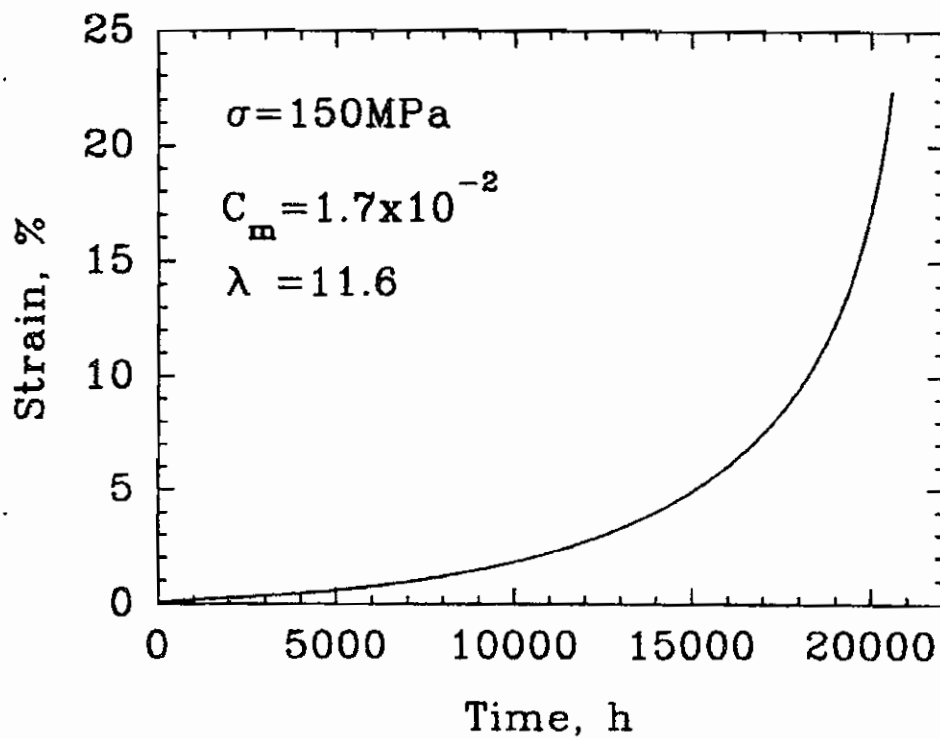
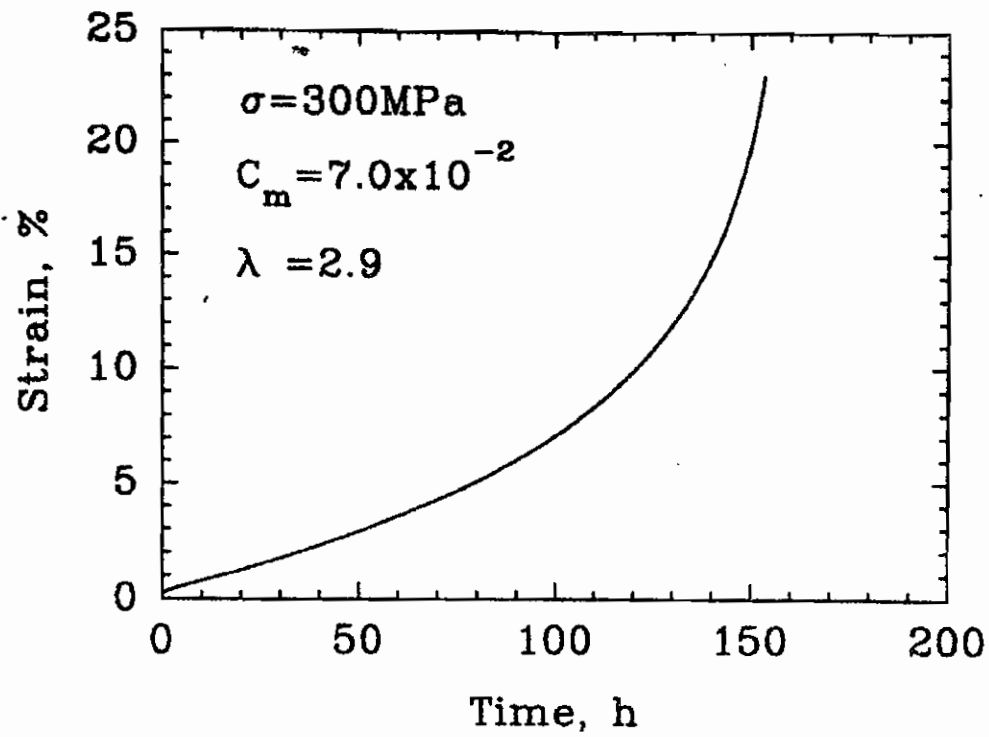


Fig 8. Comparison of model predictions for minimum creep rate and data as a function of stress and temperature for new and service exposed material. New —; Service Exposed

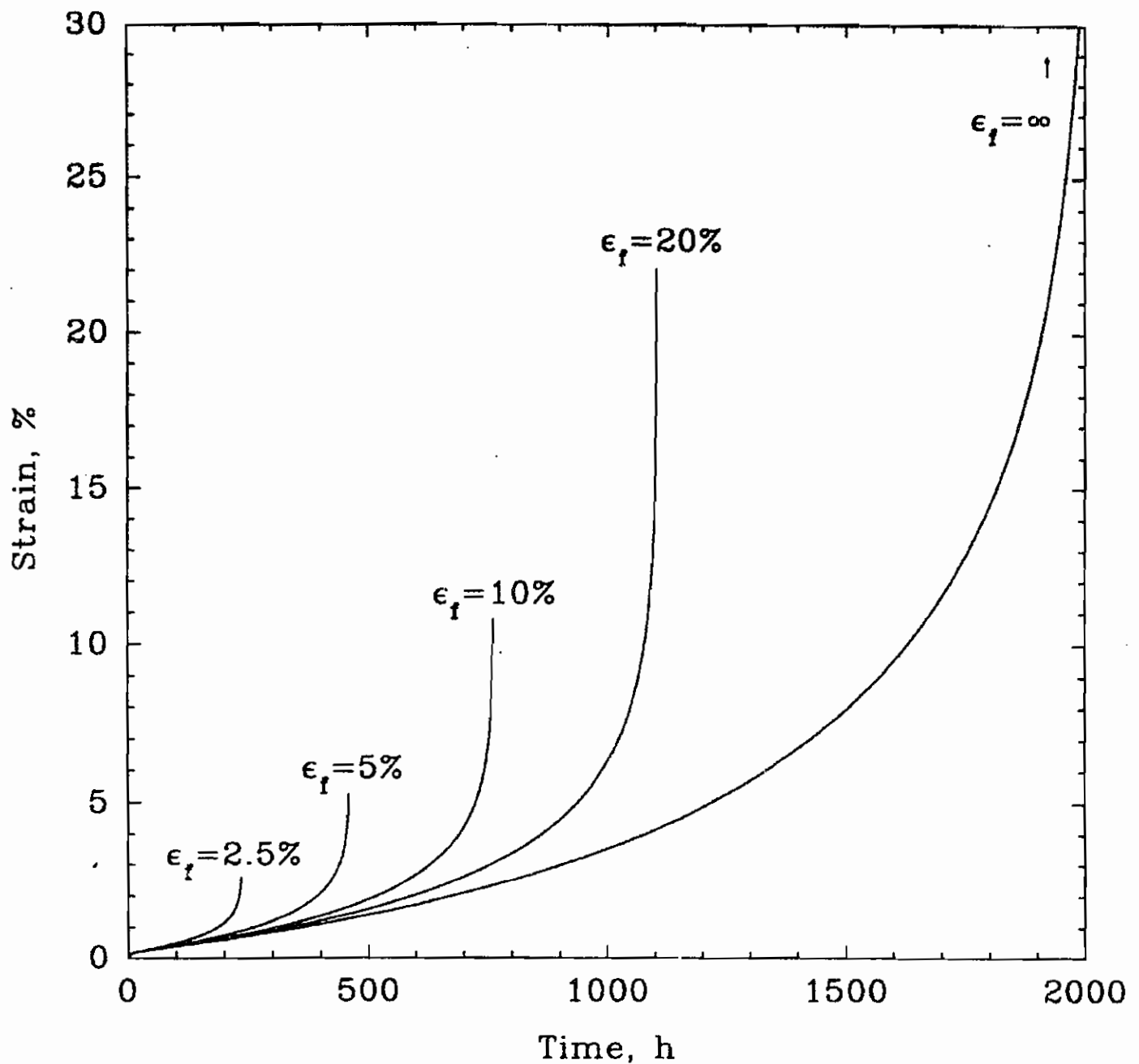


Model parameters:

$$\dot{\epsilon}_0 = 1.3 \times 10^{-9} \text{ h}^{-1}, \sigma_0 = 13.5 \text{ MPa}, h' = 10^5 \text{ MPa},$$

$$H^* = 0.4, K' = 1.1 \times 10^{-3} \text{ h}^{-1}, \epsilon_{fu} = 20\%.$$

Fig 9. The predicted influence of stress on the shape of creep curves at 530°C.



Model parameters:

$$\dot{\epsilon}_0 = 3.5 \times 10^{-9} \text{ h}^{-1}, \quad \sigma_0 = 13.0 \text{ MPa}, \quad h' = 10^5 \text{ MPa}$$

$$H^* = 0.4, \quad K' = 2.8 \times 10^{-4} \text{ h}^{-1}$$

Fig 10a. The influence of rupture ductility on the shape of creep curves predicted by the model at 200MPa and 550°C.

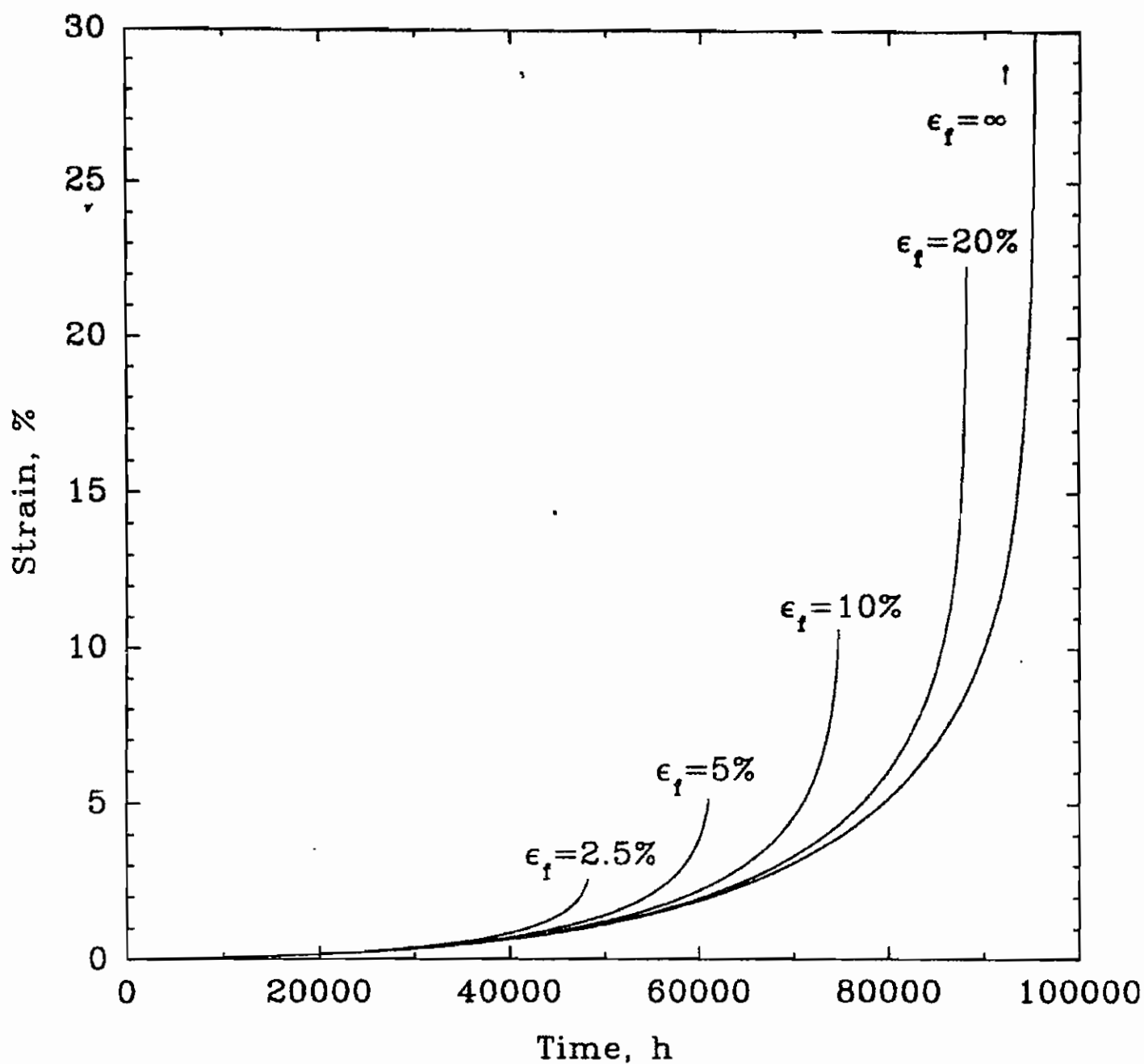


Fig 10b. The influence of rupture ductility on the shape of creep curves predicted by the model at 50MPa and 550°C.

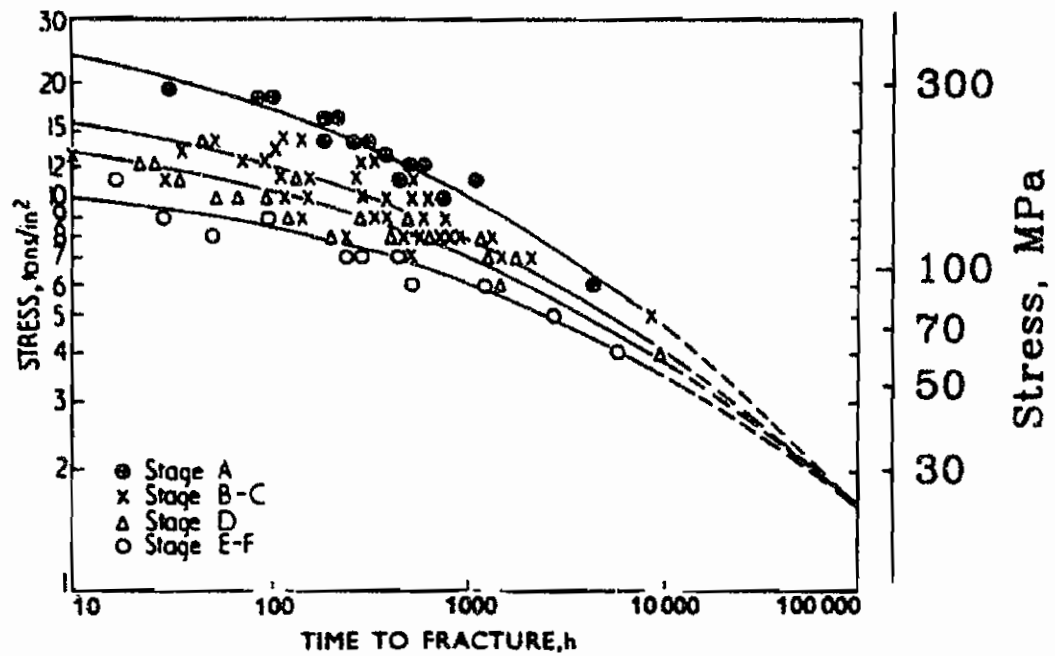


Fig 11. 1Cr 1/2Mo creep data after different thermal exposures, Toft and Marsden (13).

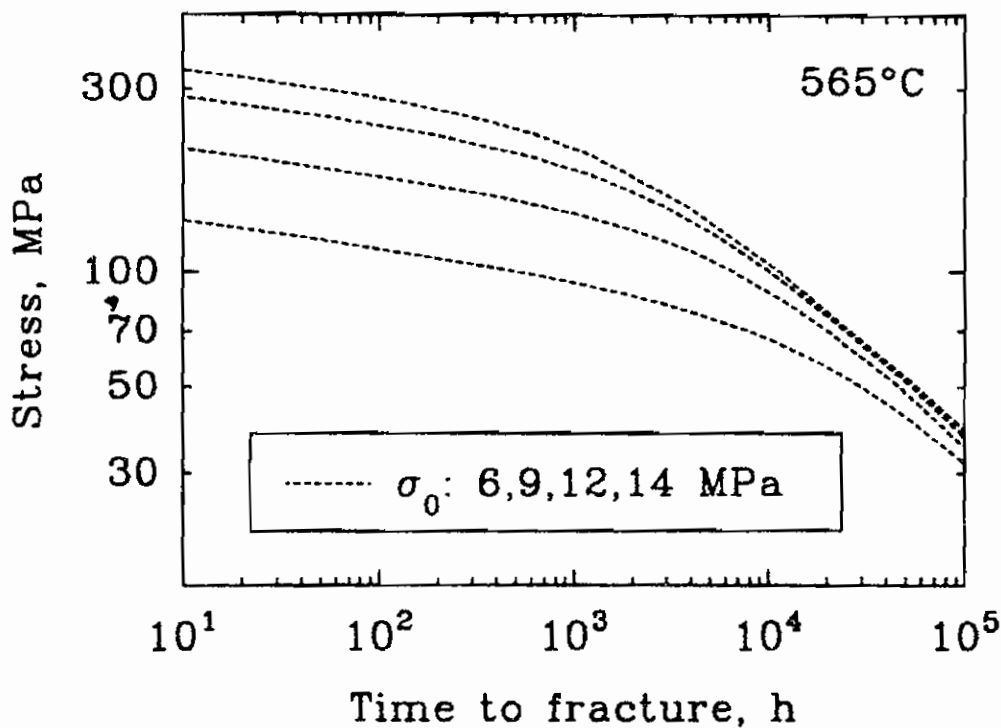


Fig 12. Predicted creep lifetime as a function of stress at 565°C with different initial thermal exposures (Characterised by σ_0).

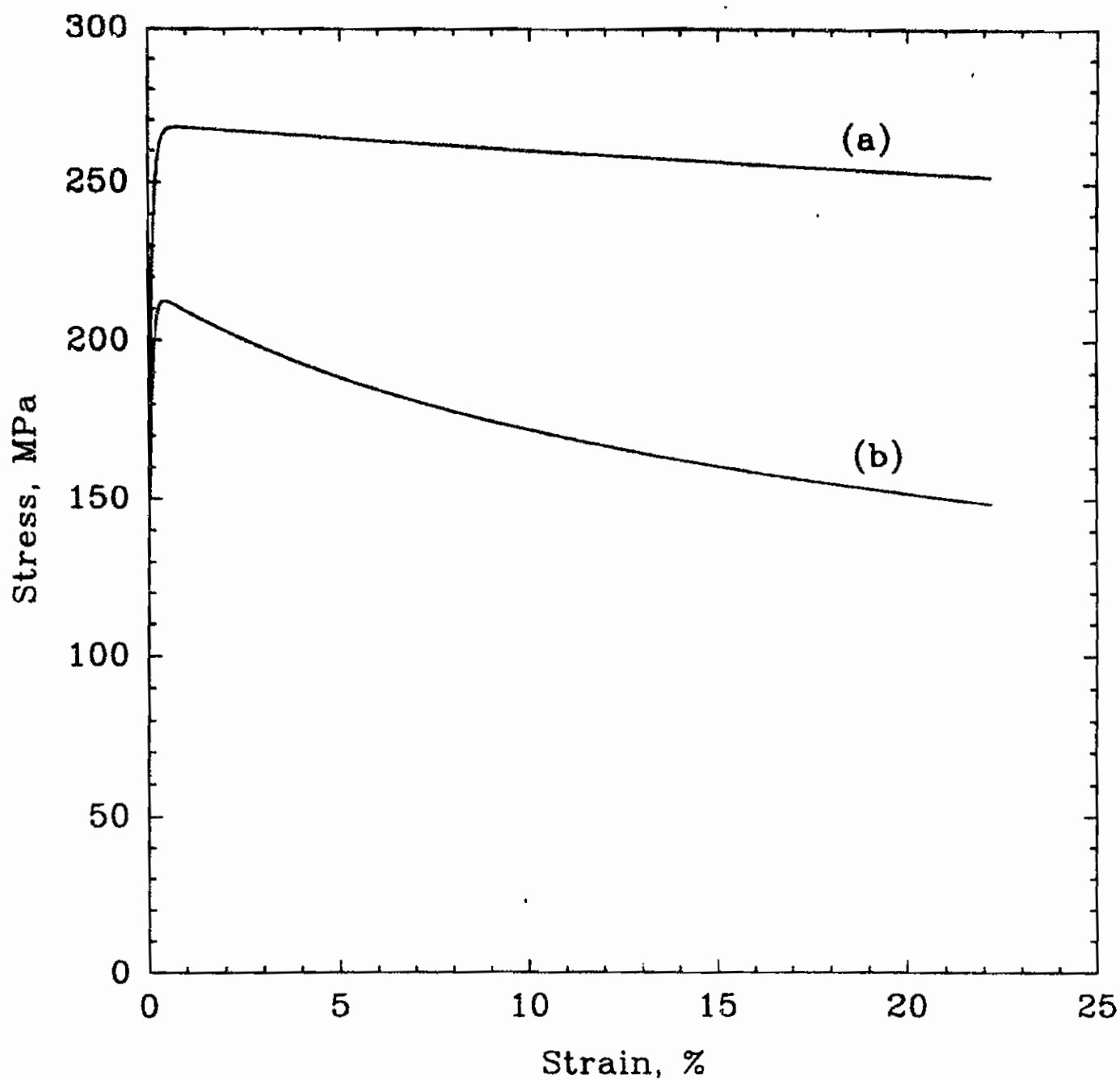


Fig 13. Predicted stress/strain curves for constant strain rate tests at 530°C.

Applied strain rate = (a) 10^{-4} h^{-1} (b) 10^{-5} h^{-1}

Young's Modulus = 150GPa

Model parameters from table 2.

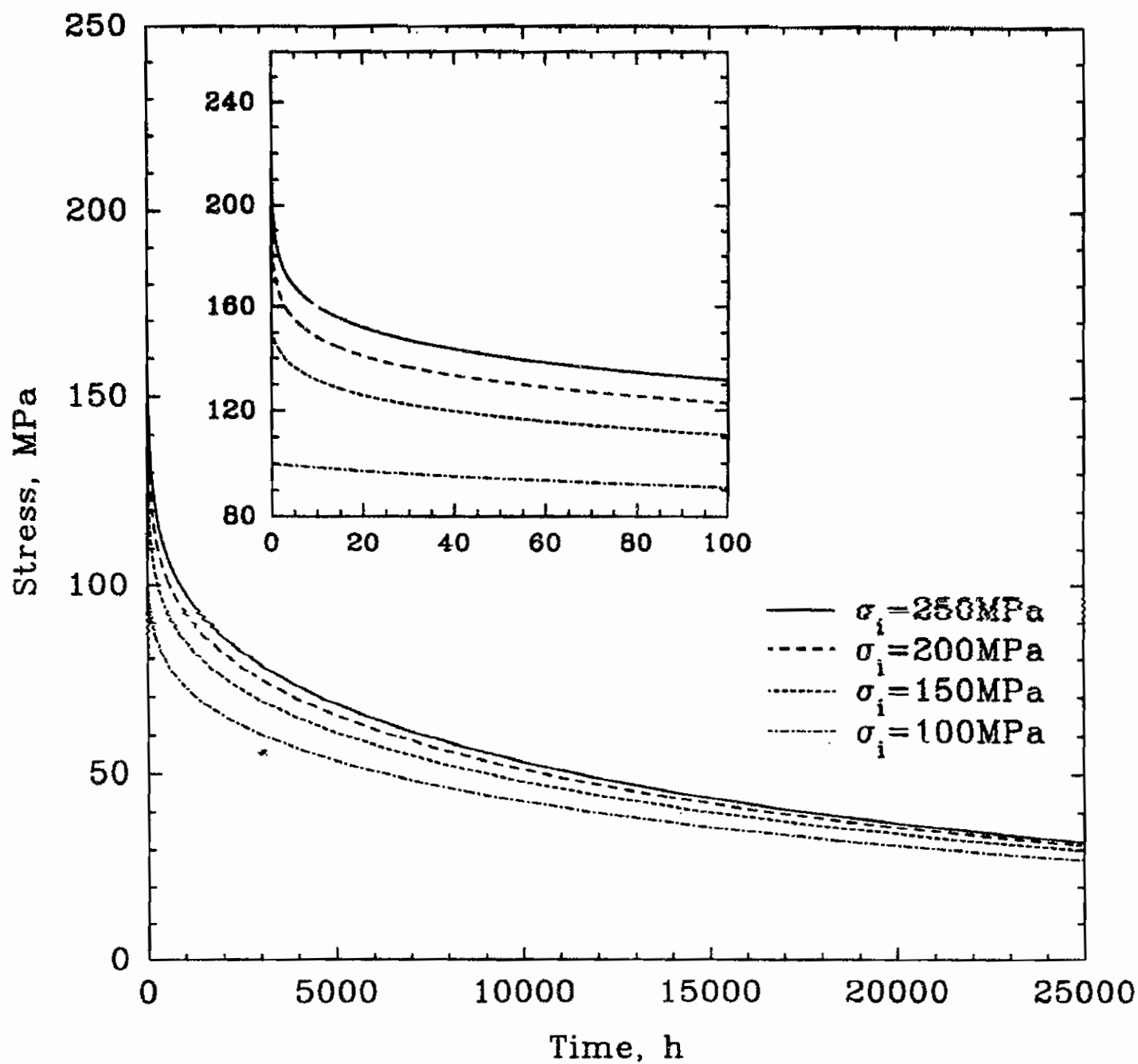


Fig 14. Predicted stress/time curves for stress-relaxation tests at 530°C.

Model parameters from Table 2.

Young's Modulus=150GPa

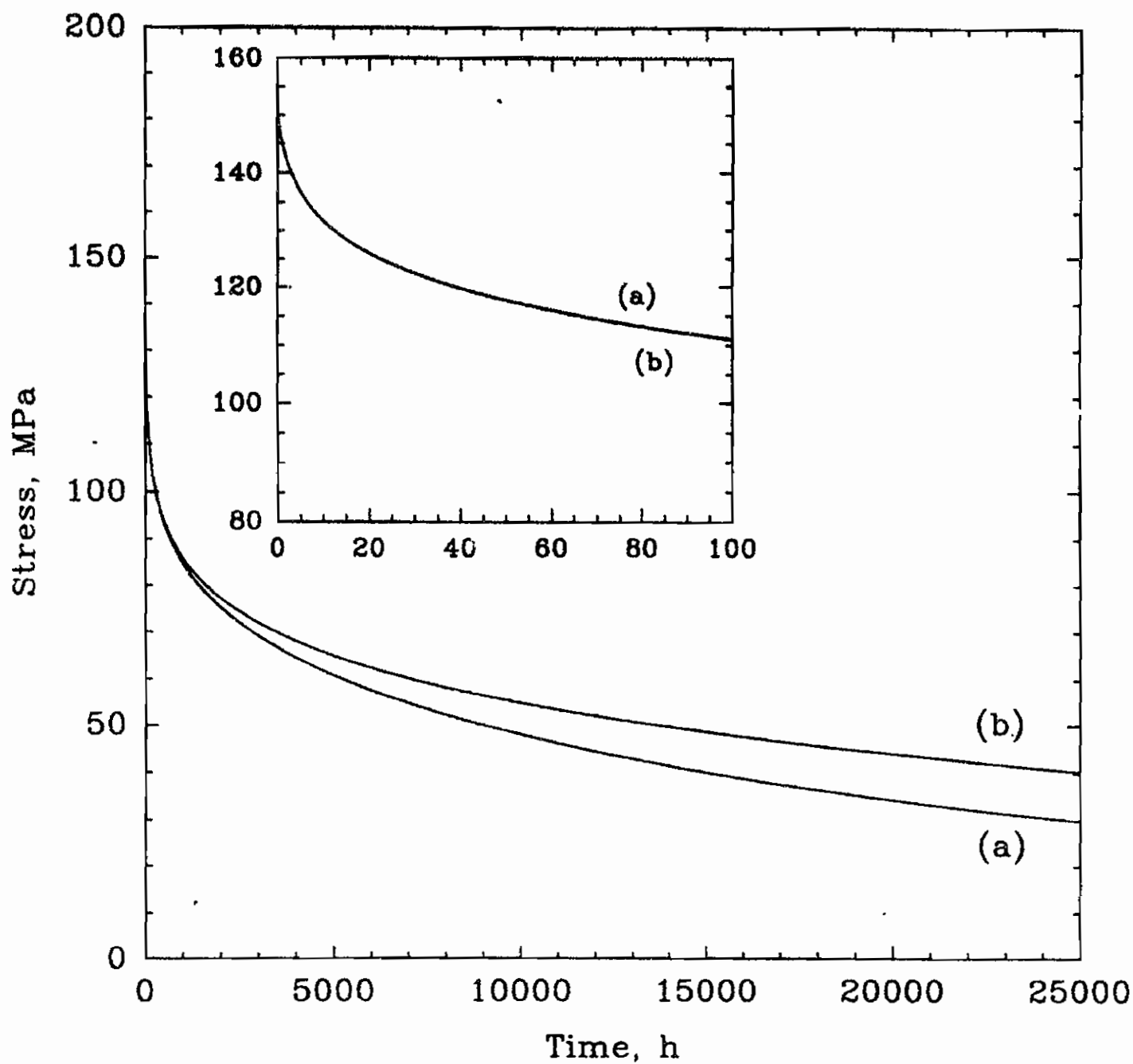


Fig 15. Predicted stress/time curves for stress-relaxation tests at 530°C

Initial Stress=150MPa.

(a) $K'=1.1 \times 10^{-4} \text{ h}^{-1}$, (b) $K'=0 \text{ h}^{-1}$

Model parameters from table 2.

Young's Modulus=150GPa

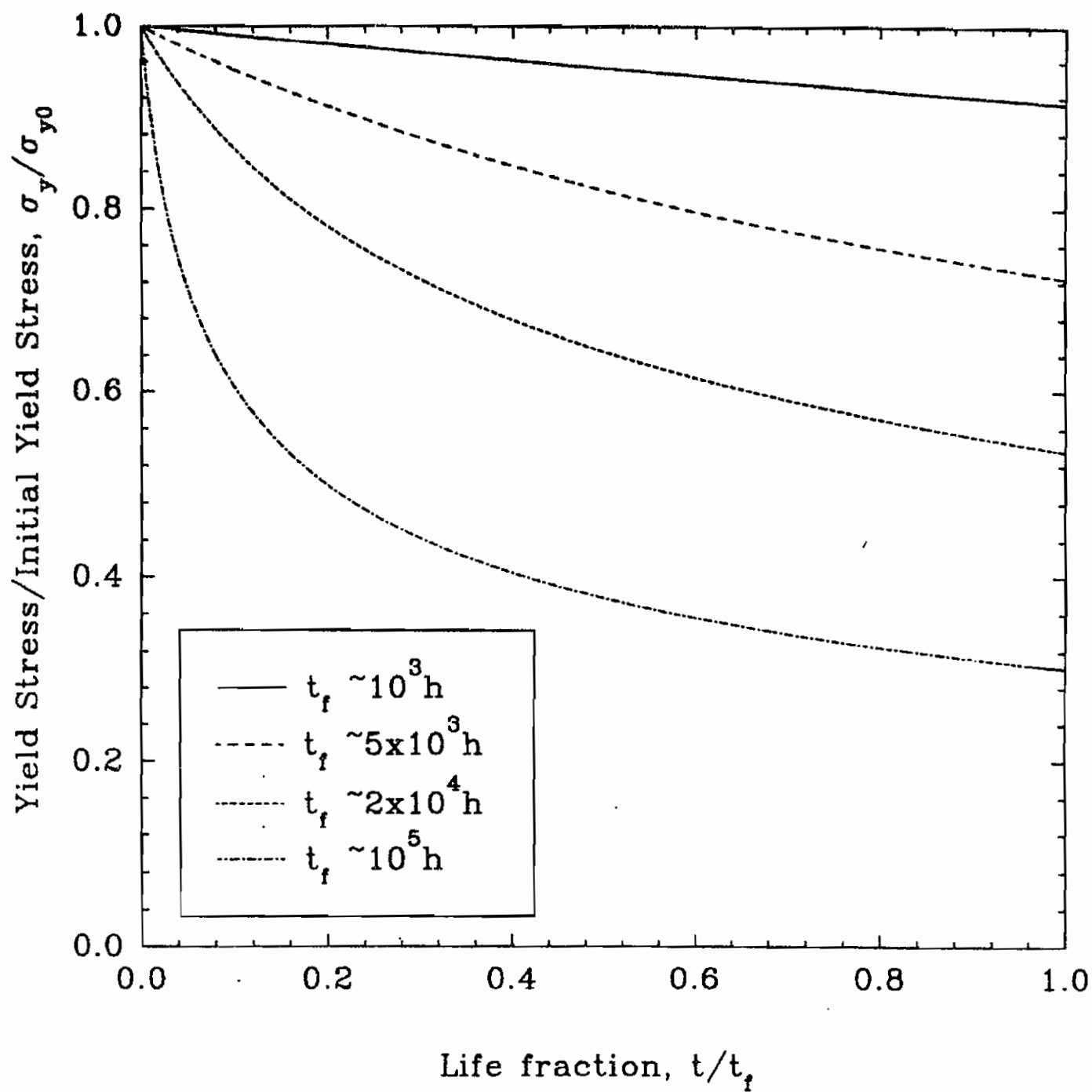


Fig 16. Predicted thermal softening as a function of fractional lifetime for material tested at 550°C.

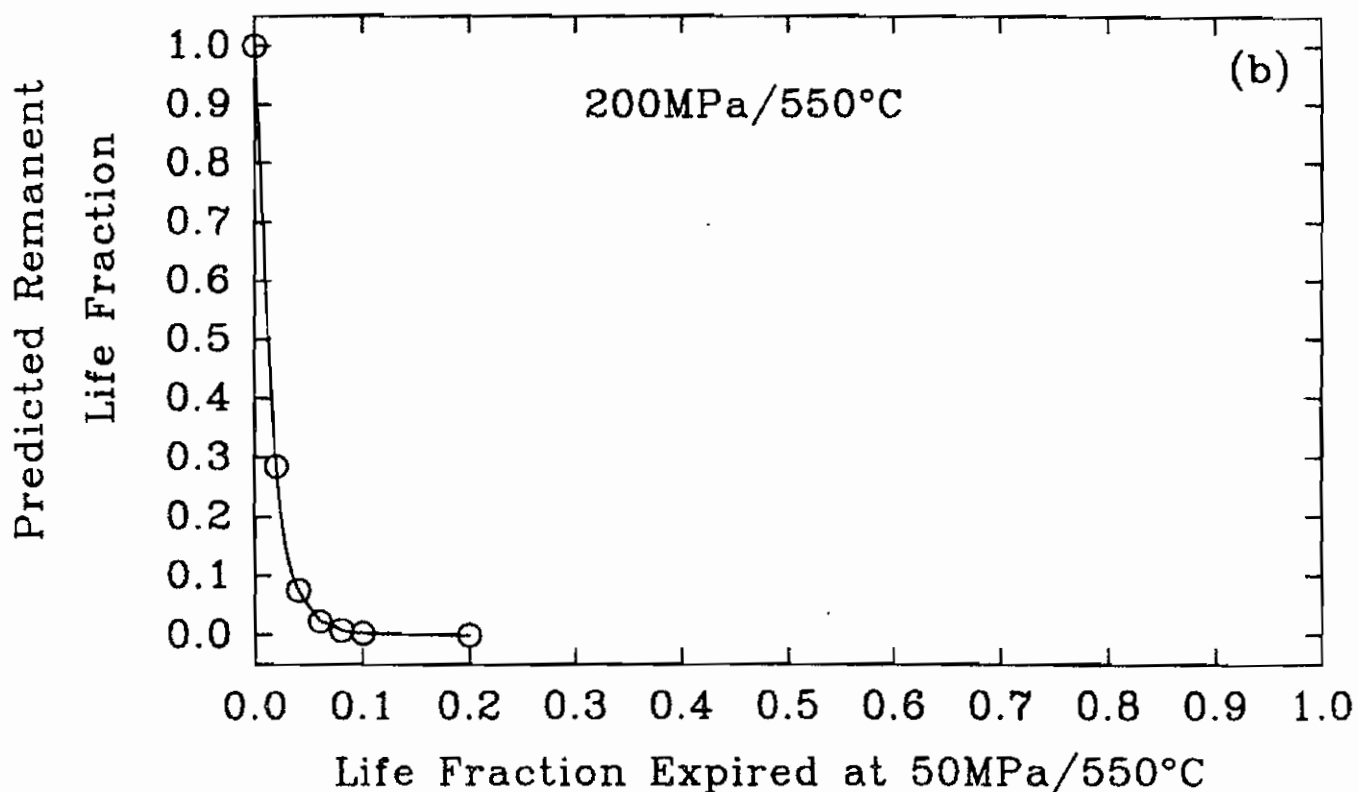
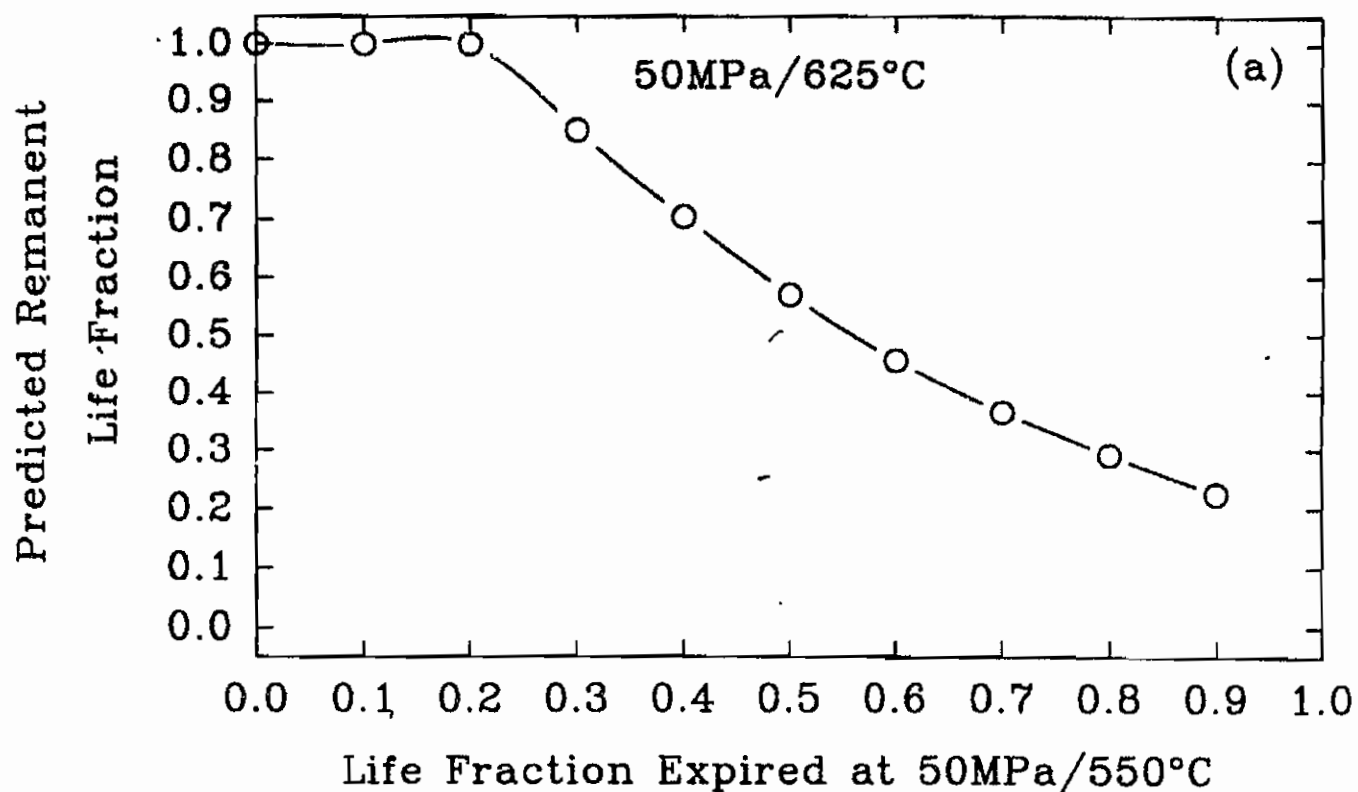


Fig 17. Simulation of remanent life prediction using Robinson's Rule. (a) Service stress testing
(b) Service temperature testing

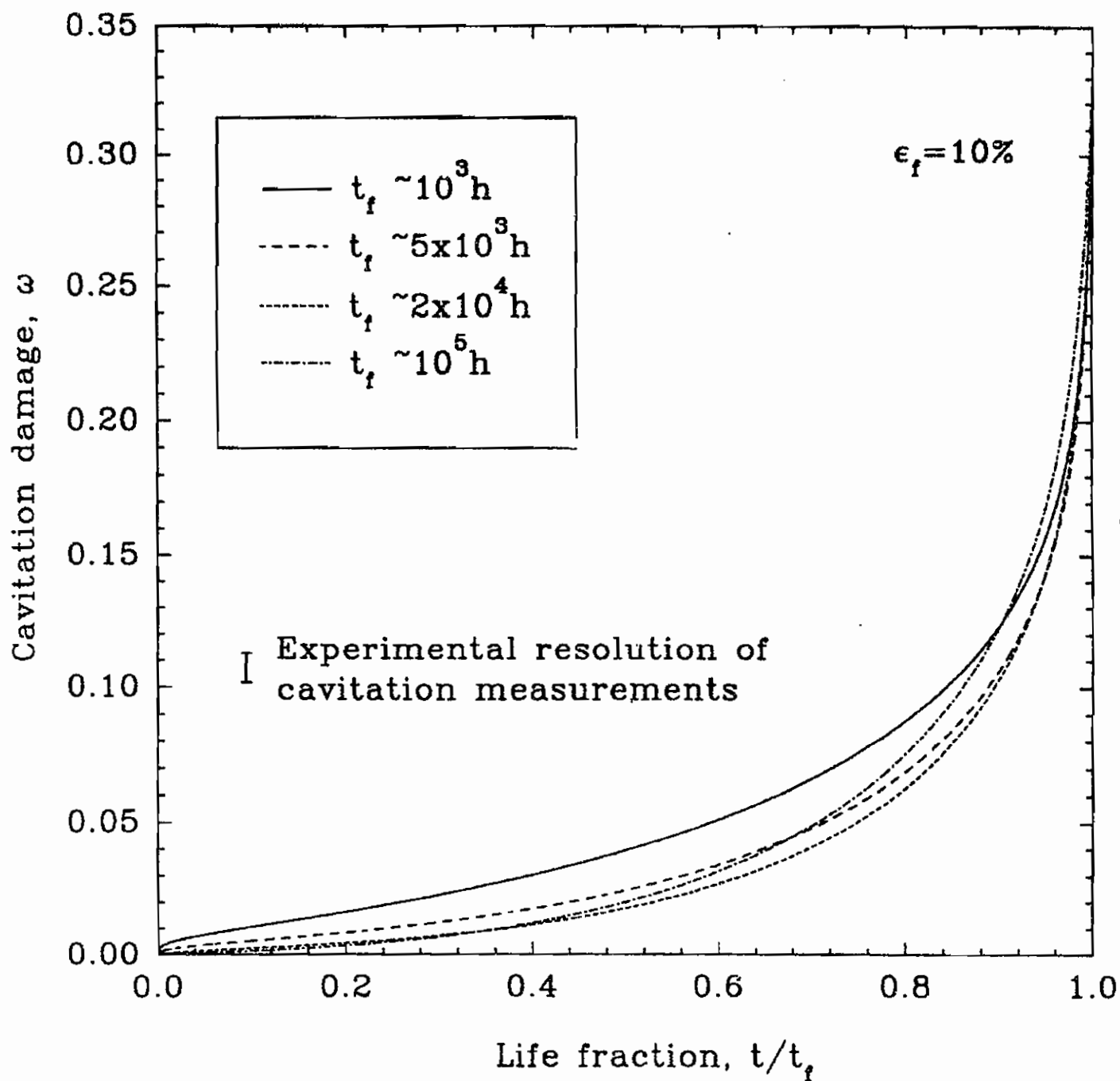


Fig 18. Predicted cavitation damage as a function of fractional lifetime for material tested at 550°C.

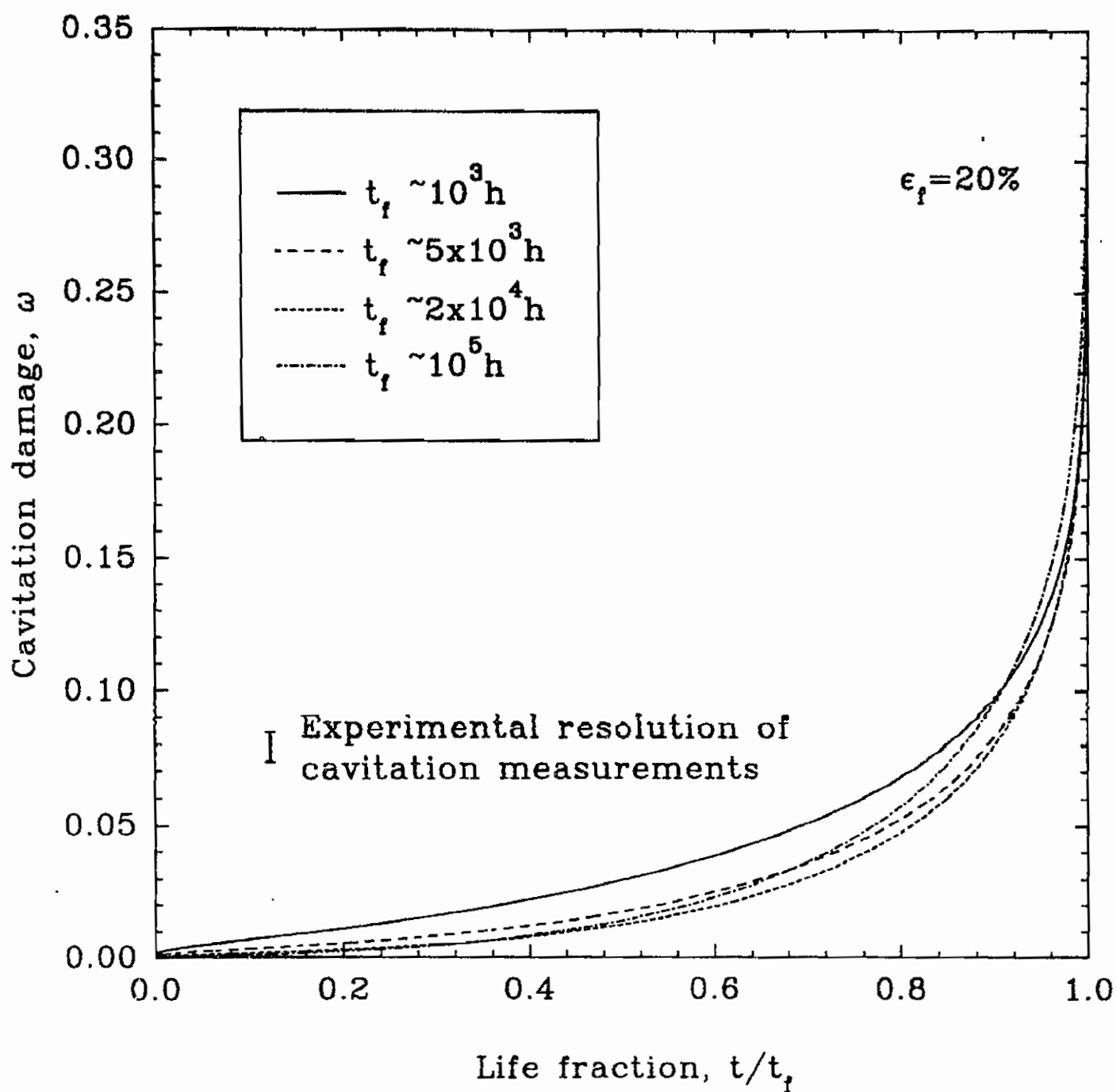


Fig 19. Predicted cavitation damage as a function of fractional lifetime for material tested at 550°C .

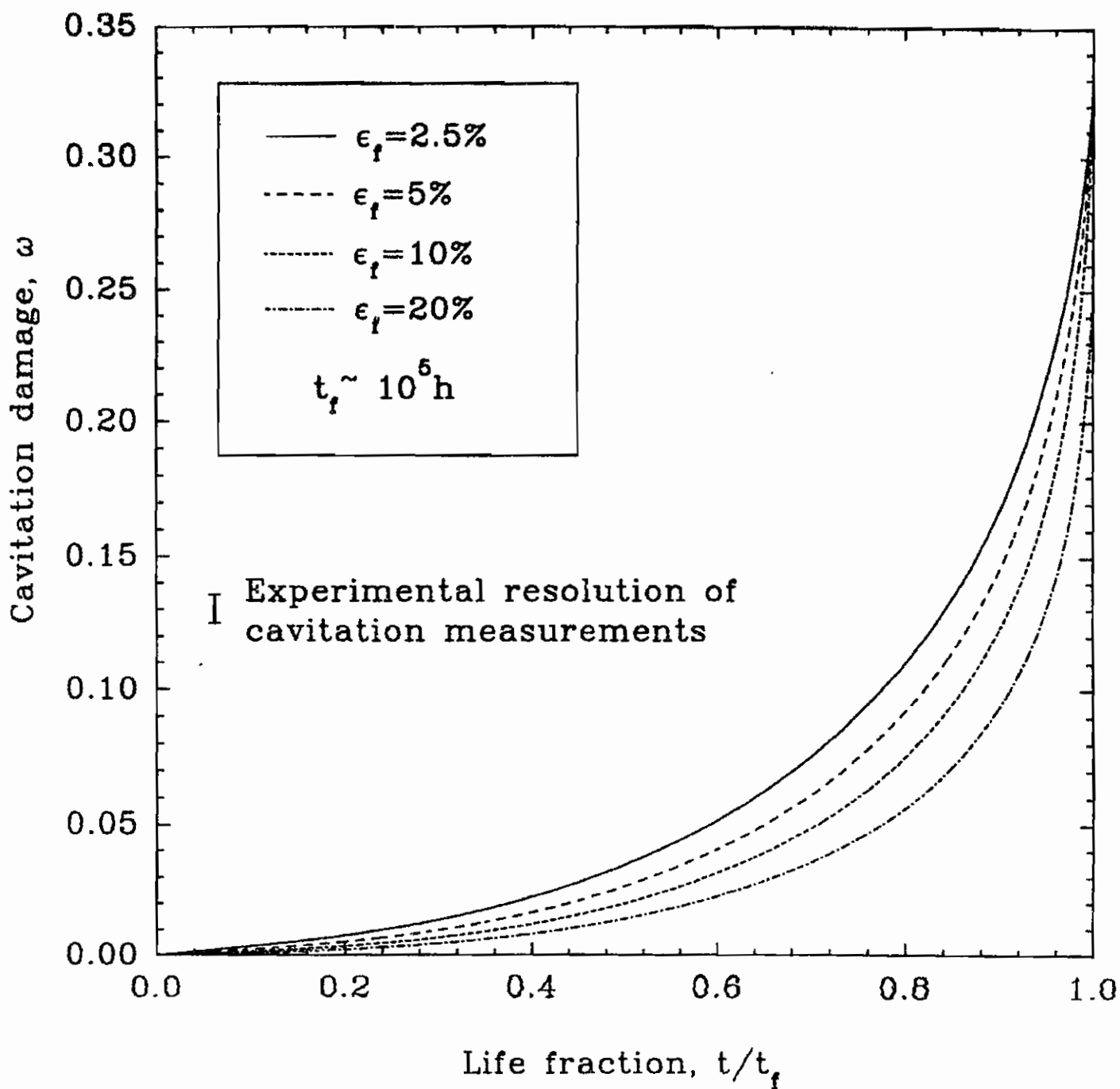


Fig 20. Predicted cavitation damage as a function of fractional lifetime for material tested at 550°C.

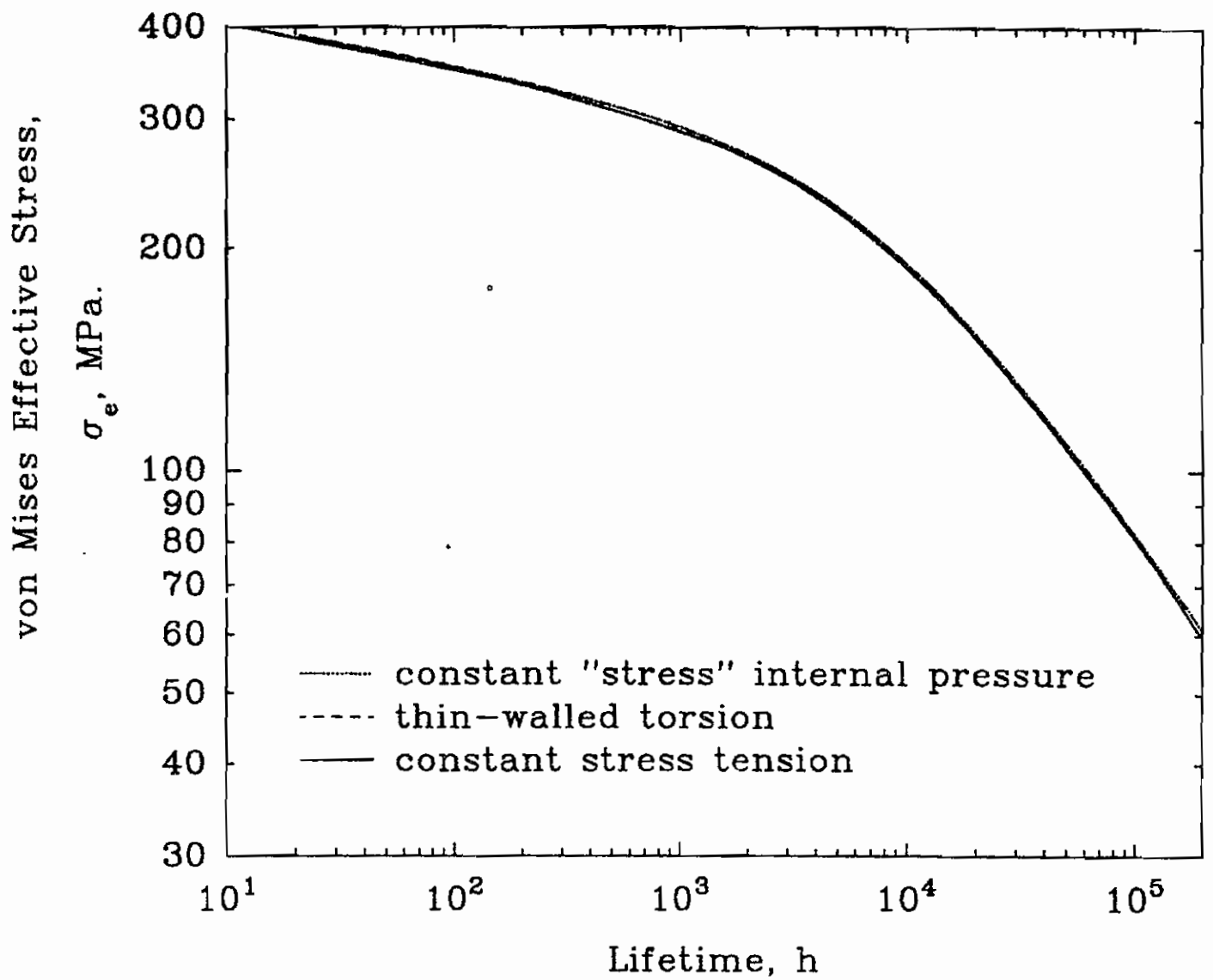


Fig 21a Model lifetimes at 530°C as a function of von Mises effective stress for three stress-states: constant stress tension, torsion and internally pressurised thin-walled tubes with no end-load under constant stress.

Constant failure criterion of $\epsilon_1=0.4$

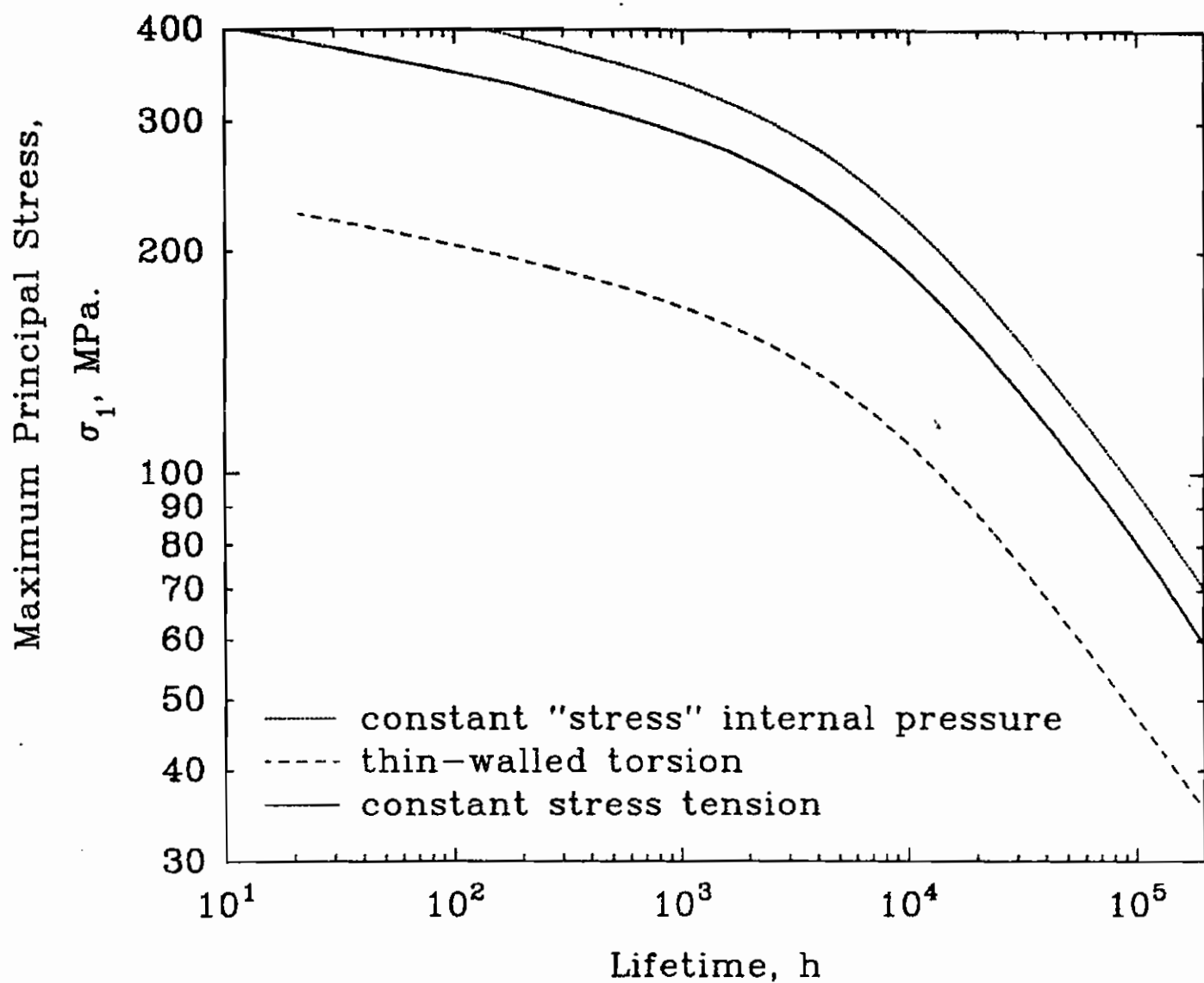


Fig 21b Model lifetimes at 530°C as a function of maximum principal stress for three stress-states: constant stress tension, torsion and internally pressurised thin-walled tubes with no end-load under constant stress.

Constant failure criterion of $\epsilon_1 = 0.4$

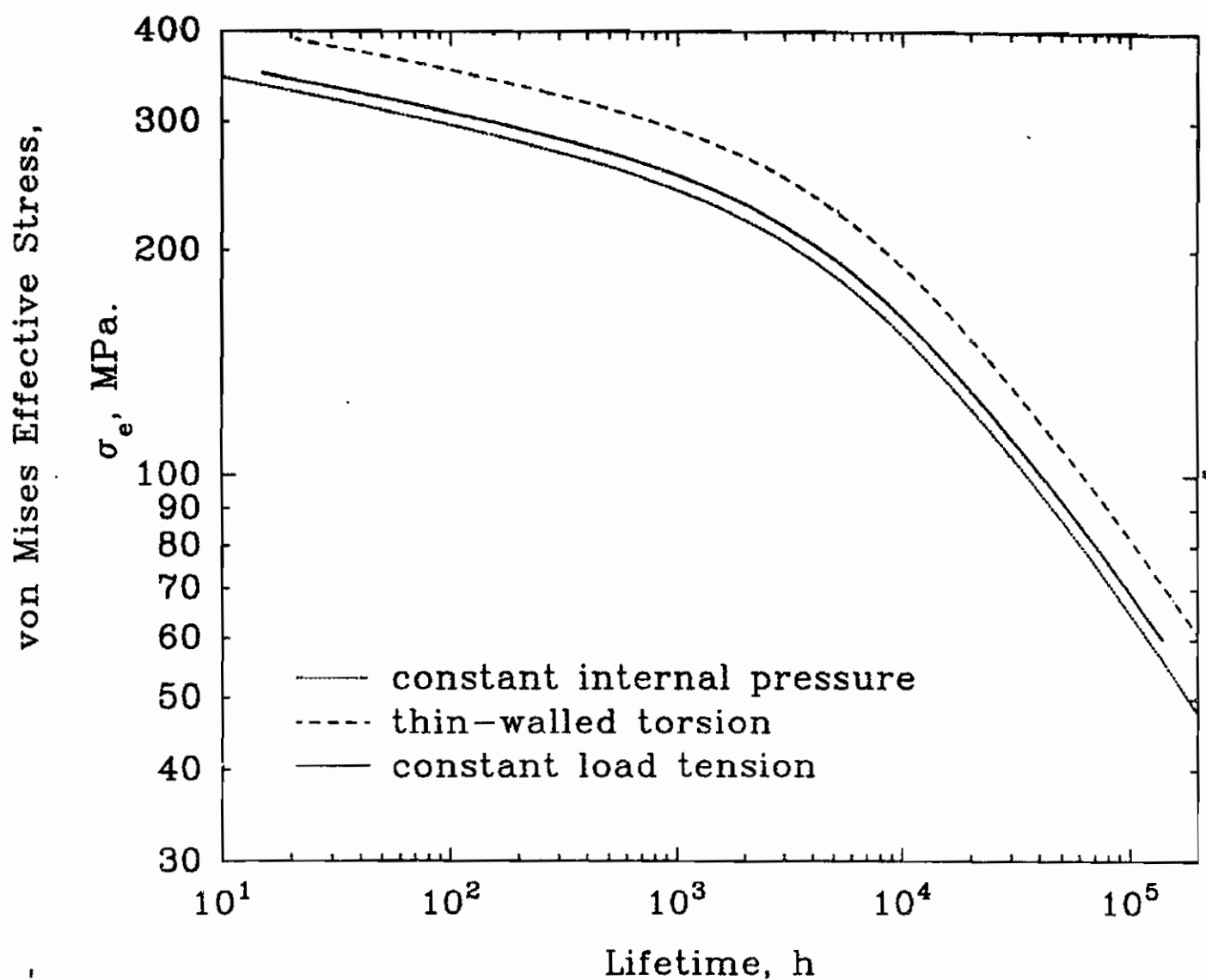


Fig 22a Model lifetimes at 530°C as a function of von Mises effective stress for three stress-states: constant load tension, thin-walled torsion and thin-walled tubes with no end-load under constant internal pressure.

Constant failure criterion of $\epsilon_1 = 0.4$

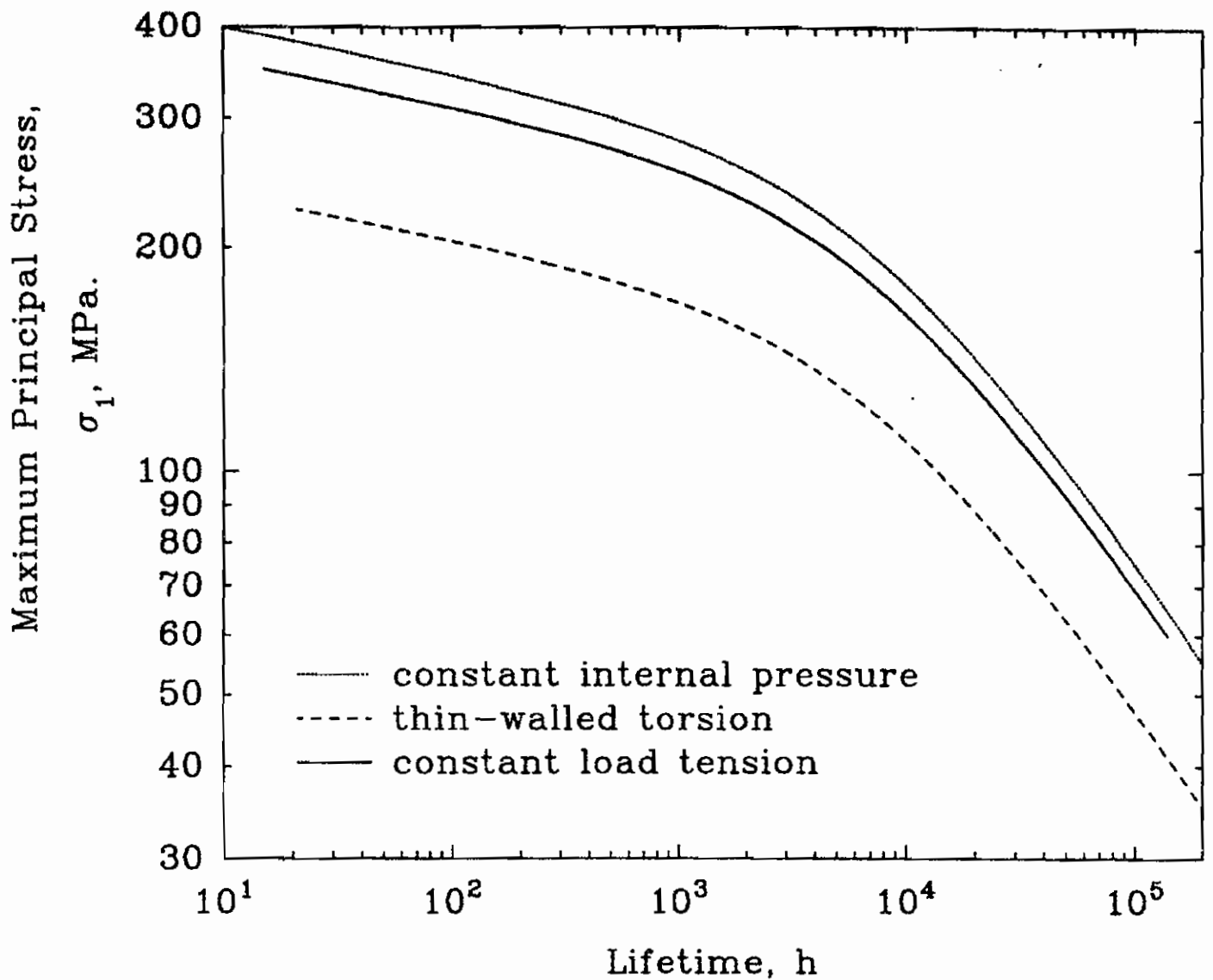


Fig 22b Model lifetimes at 530°C as a function of maximum principal stress for three stress-states: constant load tension, thin-walled torsion and thin-walled tubes with no end-load under constant internal pressure.

Constant failure criterion of $\epsilon_1=0.4$

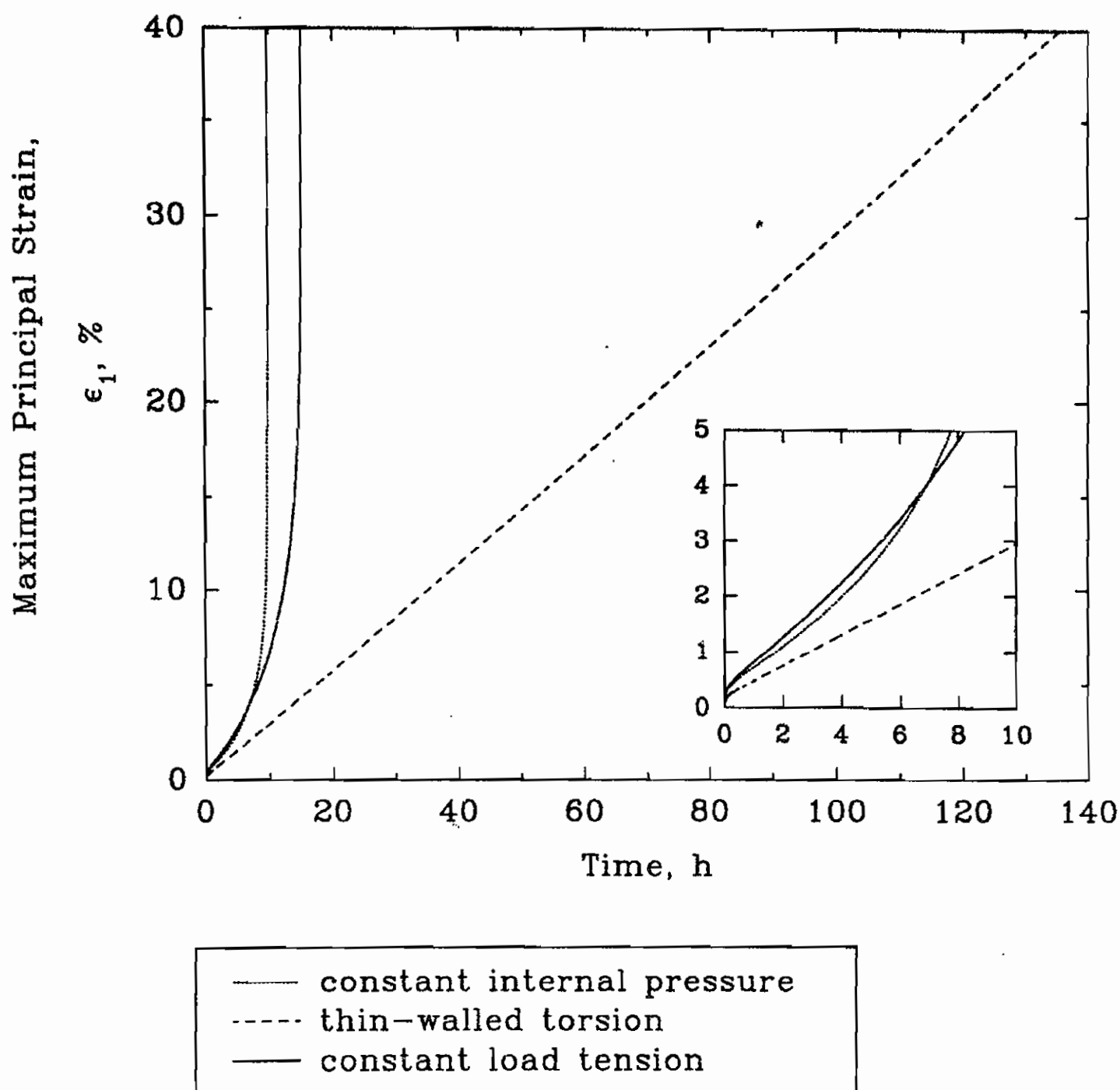


Fig 23a Model predictions of maximum principal true strain trajectories as a function of time for three stress-states at 530°C and 350 MPa von Mises effective stress: constant load tension, torsion and internally pressurised tubes under constant pressure.

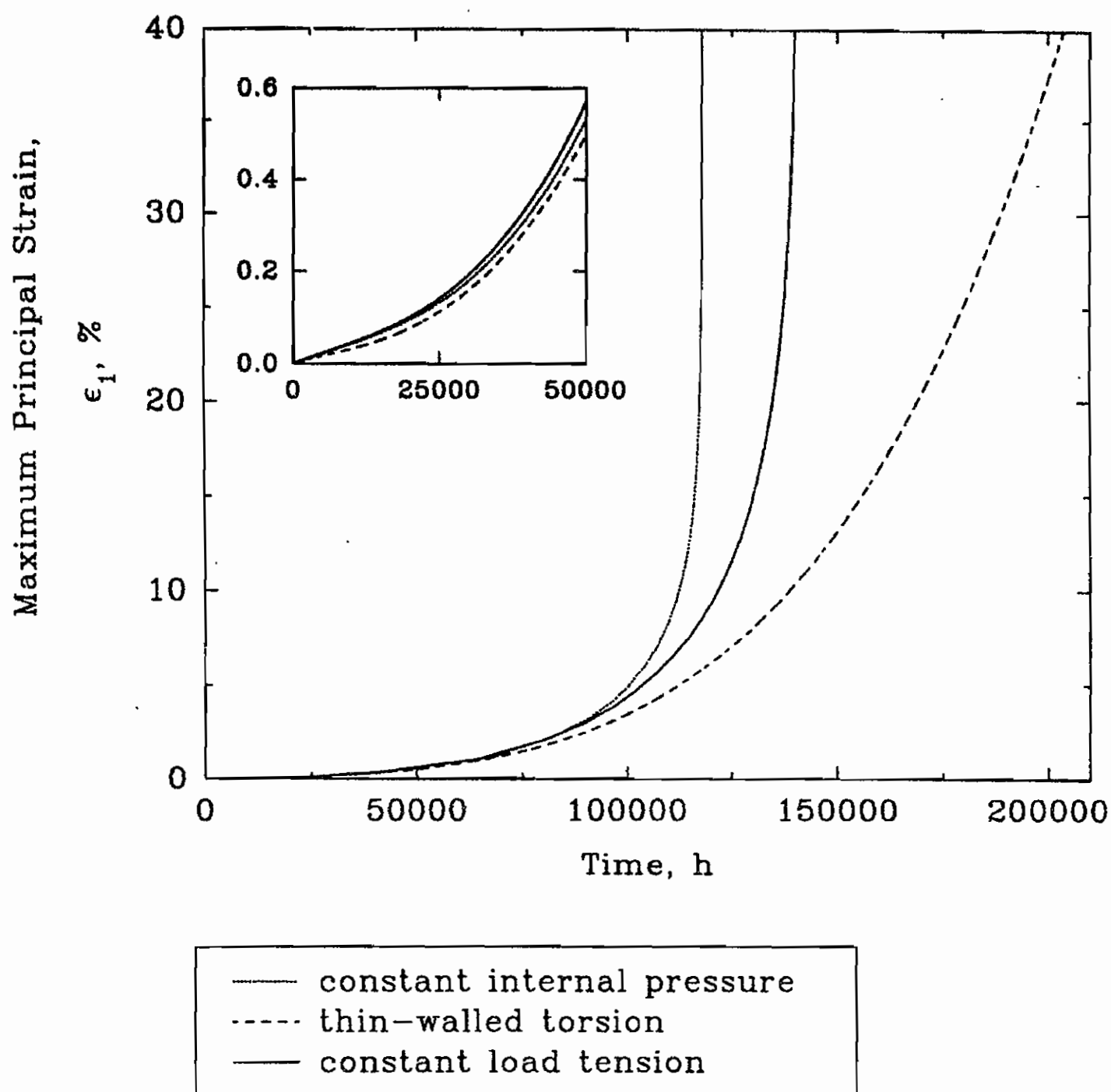


Fig 23b Model predictions of maximum principal true strain trajectories as a function of time for three stress-states at 530°C and 60 MPa von Mises effective stress: constant load tension, torsion and internally pressurised tubes under constant pressure.