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A Candidate Creep-Recovery Model for 2¼ Cr—1 Mo Steel and Its Experimental Implementation

D. N. Robinson

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Liquid Metal
Fast Breeder Reactors

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REACTOR DIVISION

A CANDIDATE CREEP-RECOVERY MODEL FOR 2 1/4 Cr–1 Mo STEEL AND ITS EXPERIMENTAL IMPLEMENTATION

D. N. Robinson

OCTOBER 1975

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A CANDIDATE CREEP-RECOVERY MODEL FOR 2 1/4 Cr—1 Mo STEEL AND ITS EXPERIMENTAL IMPLEMENTATION

D. N. Robinson

ABSTRACT

A uniaxial, creep-recovery model due to R. Lagneborg is recast into a form which facilitates its implementation through a testing program which is entirely phenomenological. The model is being considered under the High-Temperature Structural Design Program at ORNL as a candidate model for 2 1/4 Cr—1 Mo steel at high temperature. A specific experimental program is proposed for the determination of the several material parameters of the model. Generalization of the uniaxial formulation to that for general states of stress is briefly discussed.

Key Words: creep, constitutive relation, high-temperature, inelastic, deformation, reactor metals, anelastic, ORNL-HTSD Program

INTRODUCTION

The primary objective of the High-Temperature Structural Design Program at ORNL is to develop and recommend structural design methods and criteria for high temperature reactor system components. Recently, recommendations arising from the program were put forth concerning the inelastic structural analysis of 2 1/4 Cr—1 Mo steel reactor components.[1,2] As indicated in Refs. 1 and 2, the methods and procedures cited there are considered tenuous and it is recommended that they be employed only on an interim basis, that is to say, until more extensive experimental and theoretical investigations have been conducted concerning the material. The present work is a part of the continuing effort under the High-Temperature Structural Design Program to refine and update the constitutive equations and methods of analysis for use with 2 1/4 Cr—1 Mo steel components.

In Refs. 1 and 2 the strain-hardening theory of creep has been recommended for use in the high-temperature analysis of 2 1/4 Cr—1 Mo steel components, and although capable of providing accurate predictions

of creep behavior in many loading situations, the theory appears to have shortcomings in certain other situations. Two apparent inadequacies of the strain-hardening theory relative to the uniaxial creep behavior of 2 1/4 Cr—1 Mo steel are the following:*

i) it tends to over-predict the hardening which occurs under histories of increasing stress (see Figs. 1 and 2), and

ii) it does not account for the significant "state recovery"† which is observed following stress reductions at high-temperature (Fig. 3).

Each of these effects, if ignored, could lead to gross underestimates of creep strain under loading histories not unlike those encountered by some reactor components in service.

According to the now widely accepted hypothesis of Orowan,[3] high-temperature, inelastic deformation of metals occurs under the action of two (or more) competing mechanisms; a hardening process (proceeding with accumulated inelastic deformation), and a recovery or softening process (proceeding with time). In a simple creep experiment the hardening process dominates during the early (primary) stage of creep with an eventual balance of the two competing mechanisms finally being attained and resulting in a steady-state (secondary) creep stage. According to this view, the strain-hardening theory cannot predict phenomena such as the observed softening or state-recovery cited above, since it does not embody a recovery-like mechanism. Clearly, if such phenomena are to be accounted for in analysis it is necessary to adopt more comprehensive creep-recovery constitutive laws.

Several physically based theories of creep of metals have been proposed which follow the Orowan hypothesis. One is that of Gittus[5] whose formulation employs the density of crystal dislocations as an internal state variable. Gittus' model predicts phenomenological effects such as primary and secondary creep and state-recovery, however, in predicting an ever increasing dislocation density during constant load creep, it appears to contradict the results of existing microscopic observations.

---

*The same shortcomings of the strain-hardening theory have been found relative to other metals as well, see for example, Marriott.[4]

†"Recovery of state" is evidenced in Fig. 3 in as much as the creep strain rate following the zero stress segment is close to a full order of magnitude greater than that just prior to unloading.

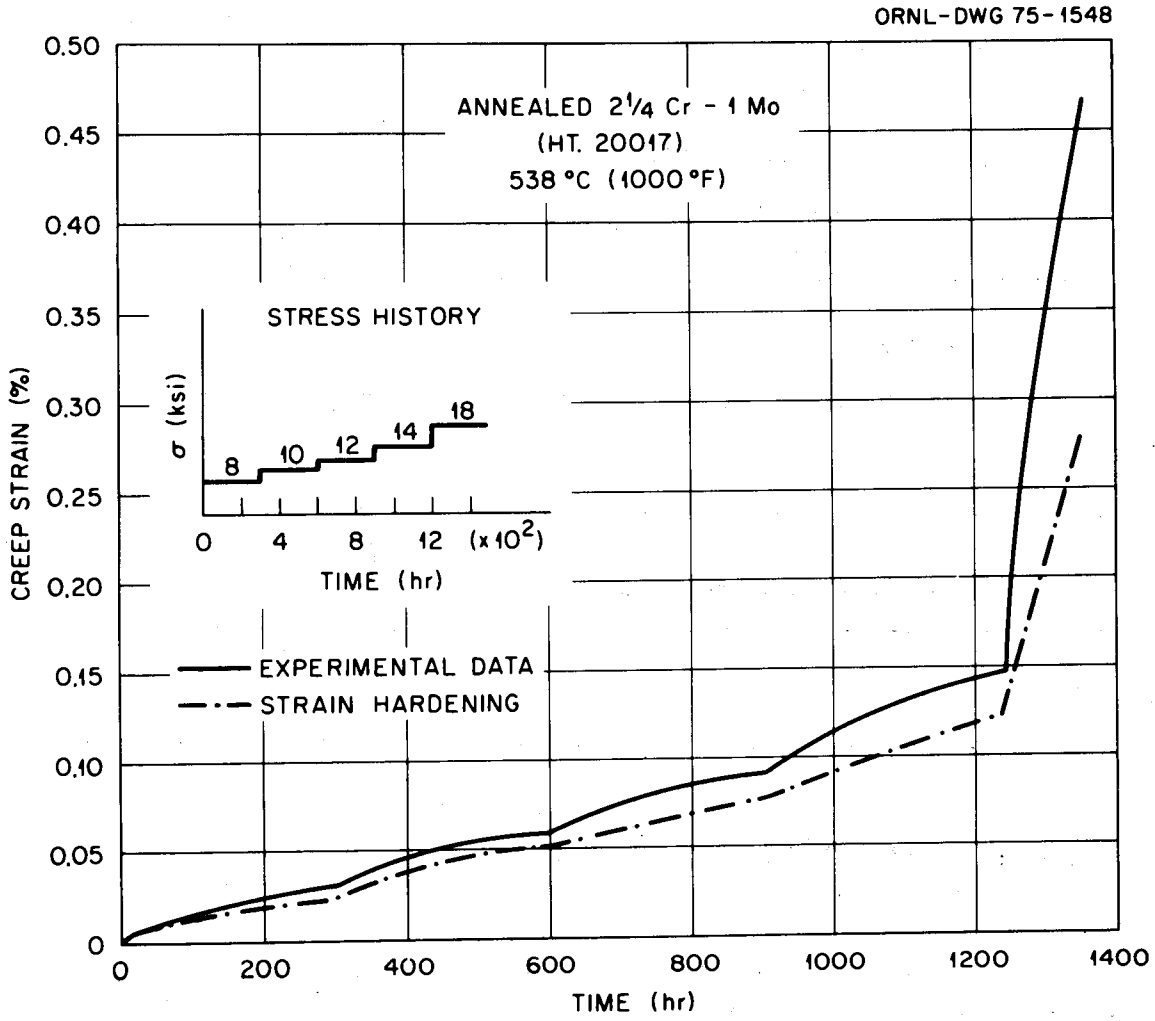

Fig. 1. Comparison of the actual and predicted (strain-hardening) response of uniaxial tensile specimens of annealed 2 1/4 Cr–1 Mo steel. Prediction is based on creep-time data obtained in constant stress creep tests.

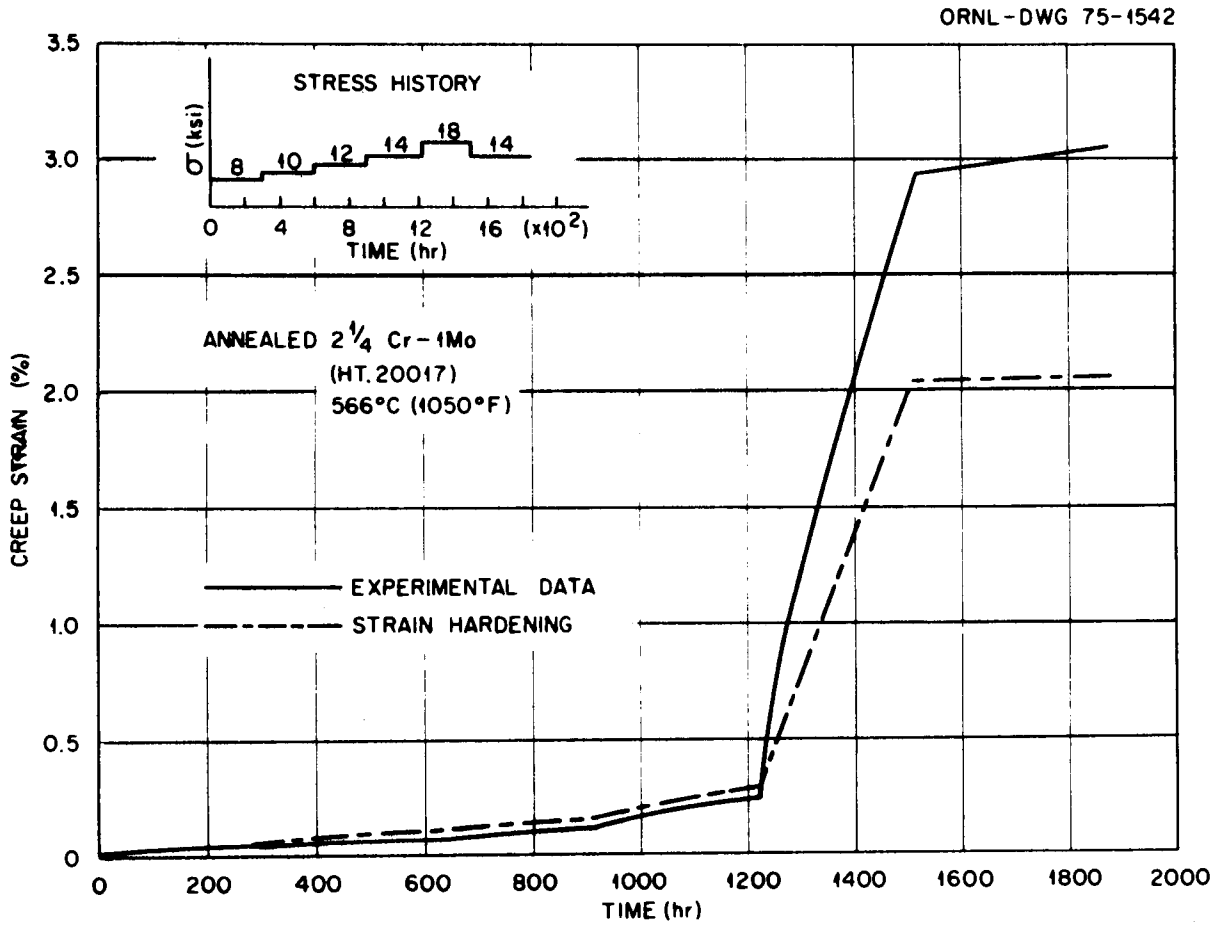

Fig. 2. Comparison of the actual and predicted (strain-hardening) response of uniaxial tensile specimens of annealed 2 1/4 Cr–1 Mo steel. Prediction is based on creep-time data obtained in constant stress creep tests.

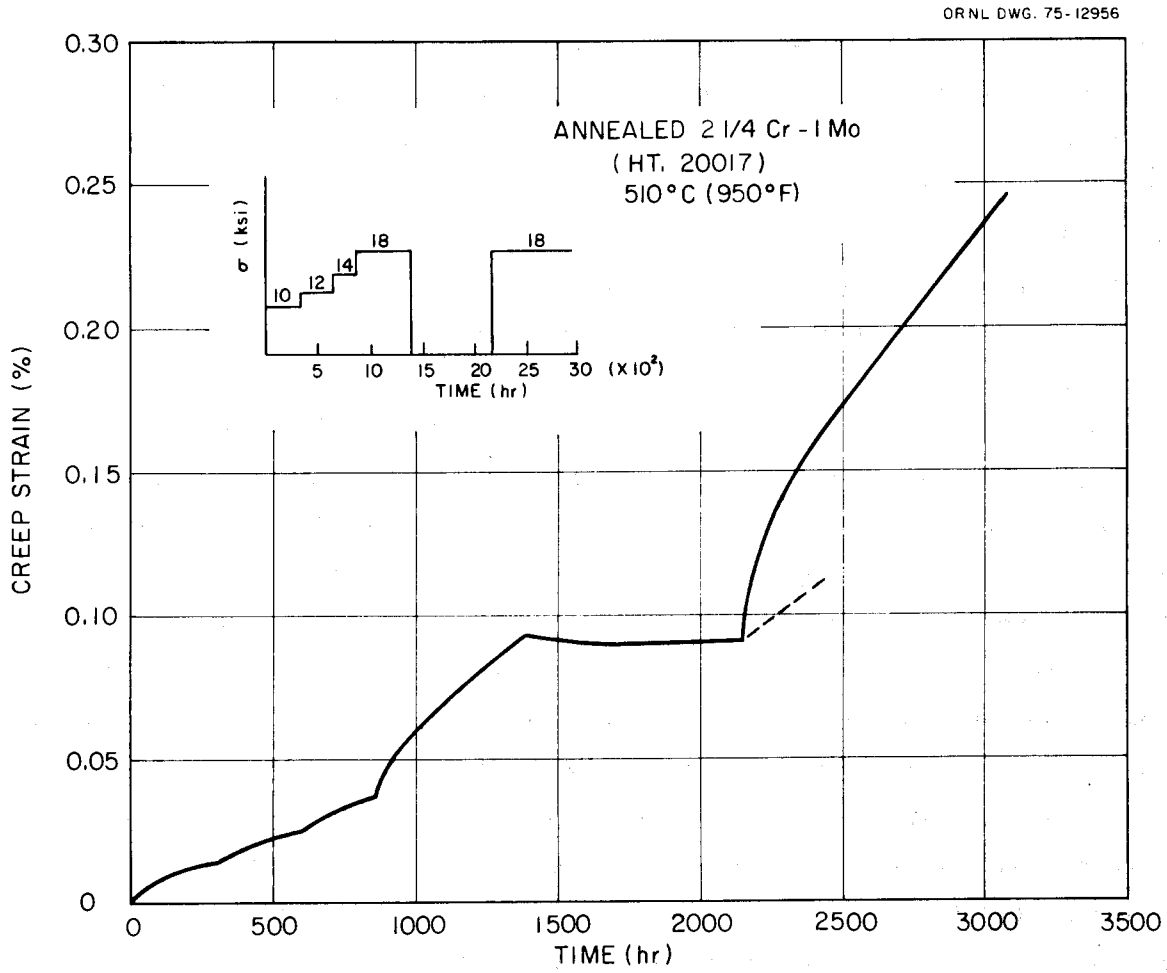

Fig. 3. Response (solid line) of annealed 2 1/4 Cr–1 Mo tensile specimen to a stress history including a recovery (zero stress) segment of about 800 hr at 510°C (950°F). The dotted line segment indicates the approximate creep strain rate predicted by strain-hardening theory upon reloading.

Other authors have proposed theories of creep in the spirit of Gittus, employing the dislocation density as a state variable; the essential difference in these theories is in the form of the growth law, i.e., the law(s) expressing the rate of change of the state variable(s). One work of this kind which appears attractive from several points of view is that of Lagneborg.[6] Lagneborg's model includes the concept of an internal flow stress, taken proportional to the square root of the average dislocation density, and only when it is exceeded are inelastic strains incurred. The growth law in Lagneborg's model accounts for the reduction in dislocation density by recovery, the recovery process being assumed to occur through a climb-controlled growth of the mesh of the dislocation network.

An application of Lagneborg's creep-recovery model to an austenitic steel (20 Cr—35 Ni) has been made by Lagneborg[6] and Modéer.[7] The several parameters in the model were determined on the basis of constant load creep tests and microscopic examinations of dislocation density under steady state creep. It is noteworthy that, in spite of the fact that the parameters were found largely on the basis of macroscopic observations, those which possess definite physical interpretations on the microscale were found to have quite realistic values.

A major drawback of the Lagneborg model, from the viewpoint of engineering applications, lies in the necessity of performing microscopic measurements of dislocation density to provide part of the data base. Of course, it is desirable that the fitting of a given constitutive model to a particular material be based solely on simple phenomenological experiments.

It is the purpose of the present work to recast the Lagneborg model into a form which facilitates determining the material parameters from simple macroscopic measurements. This involves identifying the internal flow stress, itself, as the relevant state variable, rather than the density of dislocations. The growth law which results from this transformation of the state variable is stated and an experimental procedure is proposed for determining the several model parameters.

Finally, a brief discussion is given concerning a generalization of the present uniaxial model to a multiaxial formulation. A simple J_2-form

of the theory is stated, which is analogous to the multiaxial form of the presently recommended strain-hardening theory.[1,2]

THE CREEP RECOVERY MODEL OF LAGNEBORG

The uniaxial creep-recovery law proposed by Lagneborg[6] reads

$$\dot{\varepsilon}^p = \begin{cases} A(\sigma - B\rho^{1/2})^n & ; \quad \sigma > B\rho^{1/2} \\ 0 & ; \quad \sigma \leq B\rho^{1/2} \end{cases} \tag{1}$$

in which $\dot{\varepsilon}^p$ is the inelastic strain rate and σ is the stress. The coefficients A and n are Arrhenius type temperature-dependent quantities; $B = \alpha bG$ with $0 < \alpha < 1$ a material constant; b, the Burger's vector; and G, the elastic shear modulus. The quantity ρ is the average dislocation density whose rate of growth is governed by

$$\dot{\rho} = C'\dot{\varepsilon}^p - D'\rho^2 \ . \tag{2}$$

The first term on the right hand side of Eq. (2) represents the hardening mechanism which proceeds with accumulated inelastic strain, whereas the second term represents the recovery process which proceeds with time. The coefficients C' and D' have physical origins, as discussed by Lagneborg,[6] and generally depend upon both temperature and stress.

RESTATEMENT OF THE LAGNEBORG MODEL

Phenomenologically, it is not convenient to employ ρ, the average dislocation density, as an internal state variable; a more convenient choice is the so-called internal flow stress $S = B\rho^{1/2}$ itself. Under this choice Eqs. (1) and (2) may be rewritten as

$$\dot{\varepsilon}^p = \begin{cases} A(\sigma - S)^n & ; \quad \sigma > S \\ 0 & ; \quad \sigma \leq S \end{cases} \tag{3}$$

and

$$S\dot{S} = C\dot{\varepsilon}^p - DS^4 \ . \tag{4}$$

Another way of arriving at the growth law expressed in Eq. (4) is provided for by the work of Mitra and McLean.[8] The results of their experiments are consistent with the Orowan hypothesis, stated in terms of the internal flow stress S as

$$\dot{S} = h(S)\dot{\varepsilon}^p - r(S) \tag{5}$$

in which h(S) is regarded as the coefficient of strain-hardening and r(S) the rate of thermal recovery. Mitra and McLean have found that h and r in Eq. (5) may be adequately described by

$$1/h \propto S^{\beta} \tag{6}$$

$$r \propto S^{n-\beta} \tag{7}$$

where n and β are constants. Taking, in particular, n=4 and β=1 in Eqs. (6) and (7) and substituting into Eq. (5), the growth law Eq. (4) results.

It is helpful for the following development to state the creep law expressed in Eqs. (3) and (4) in the form discussed by Onat and Fardshisheh,[9] i.e., in terms of a vector growth law

$$\dot{\bar{q}} = \bar{G} + \bar{H}\dot{\varepsilon} \ , \tag{8}$$

with the following identifications made:

$$\bar{q} = \begin{bmatrix} \sigma \\ S \end{bmatrix} \tag{9}$$

$$\dot{\bar{q}} = \begin{bmatrix} \dot{\sigma} \\ \dot{S} \end{bmatrix} \tag{10}$$

$$\bar{G} = \begin{cases} \begin{bmatrix} -AE(\sigma - S)^n \\ CA\dfrac{(\sigma - S)^n}{S} - DS^3 \end{bmatrix} & ; \quad \sigma > S > 0 \qquad (11) \\ \begin{bmatrix} 0 \\ -DS^3 \end{bmatrix} & ; \quad \sigma \leq S \qquad (12) \end{cases}$$

$$\bar{H} = E \begin{bmatrix} 1 \\ 0 \end{bmatrix} . \tag{13}$$

In the above $\dot{\varepsilon}$ is the total strain rate (elastic plus inelastic) and E is the Young's modulus for the material.

In this representation σ and S play the role of internal state variables which may be thought of as the coordinates of a "state space." The quantity $\bar{q}$, then represents a vector in that space and $\dot{\bar{q}}$ its rate of change with time. According to Eq. (8), under rapid strain rates, the rate of change of the state vector $\dot{\bar{q}}$ essentially follows the $\bar{H}$ field. On the other hand, in a relaxation experiment $\dot{\varepsilon}$ equals zero so that $\dot{\bar{q}}$ is determined entirely by the $\bar{G}$ field. The trajectories of the $\bar{G}$ vector field are accordingly called the "relaxation lines" by Onat and Fardshisheh.[9]

The relaxation lines in (σ, S) space which roughly correspond to the austenitic steel examined by Lagneborg[6] and Modéer[7] are shown in Fig. 4.

AN EXPERIMENTAL PROCEDURE FOR THE DETERMINATION OF THE MATERIAL PARAMETERS

The image of an isothermal,* constant-stress creep test at the stress level σ_o is shown as the path OAB† in the state space of Fig. 4. The state point B corresponds to steady state creep ($\bar{G}$ field vertical) at the given stress with the creep rate at B taken to be $\dot{\varepsilon}_{so}$. The position S_{so} of B along the line $\sigma = \sigma_o$ is not known a priori but can be found experimentally by determining the (rapid) drop in stress, say $\Delta\sigma_o$, necessary to bring the state point from B to F on the line $\sigma = S$. Such a "stress drop test" is similar to that employed by Parker and Wilshire.[10] Once $\Delta\sigma_o$ is known, then

$$S_{so} = \sigma_o - \Delta\sigma_o \,, \tag{14}$$

since it is assumed that for a "rapid enough" drop in stress the points B and F lie on a vertical line, i.e., there is not sufficient time for the

---

*The present development will be restricted to isothermal conditions. The coefficients can, if necessary, be considered functions of temperatue and fit to date taken over a specific temperature range.

†Rapid loading at a realistic strain rate would probably follow a path such as OA′ rather than OA as discussed by Onat and Fardshisheh.[9] Since this has little or no bearing on the present development, however, the question will be presently ignored.

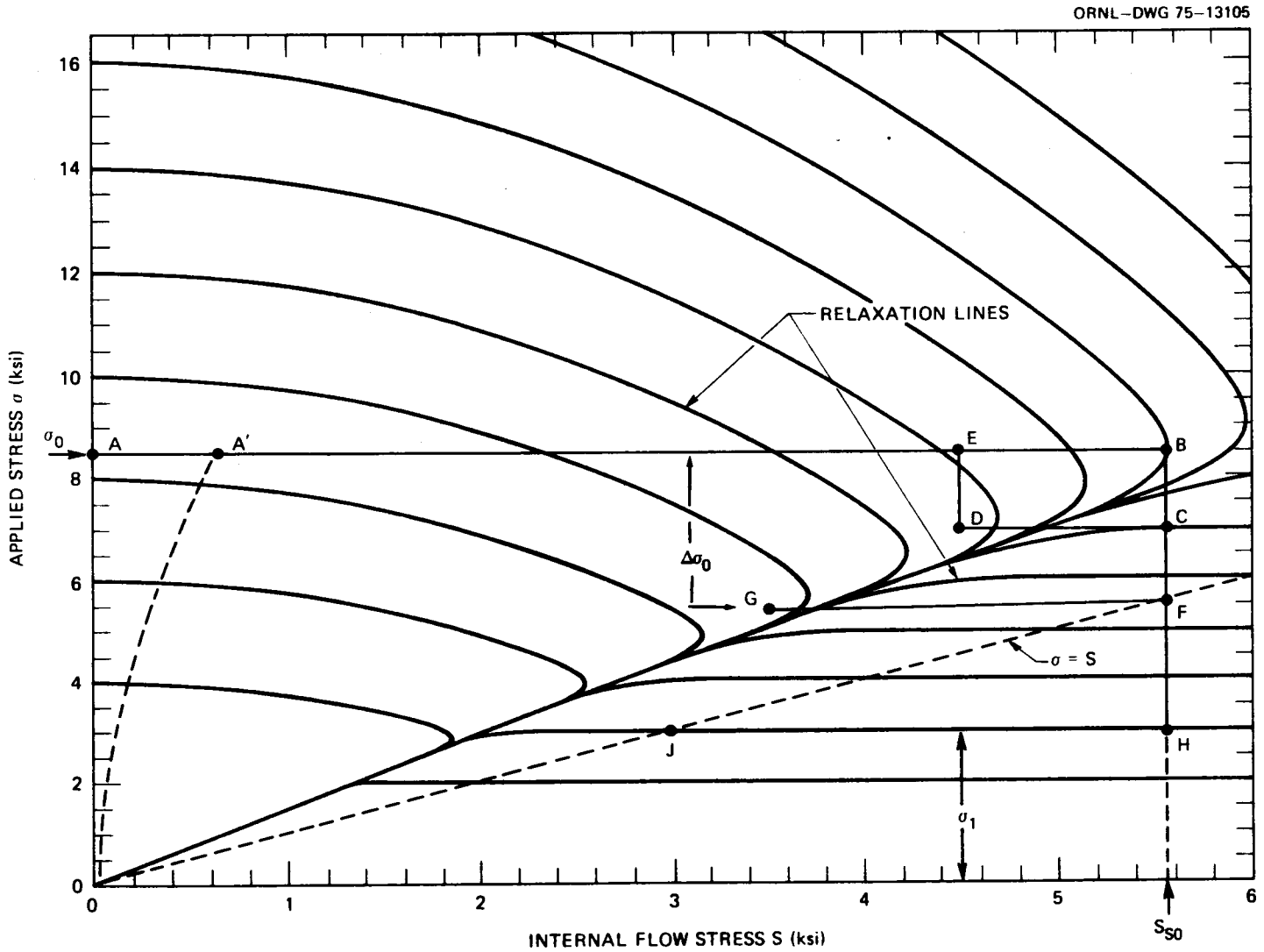

Fig. 4. State space σ vs S.

internal stress (or dislocation density) to alter significantly as the stress is suddenly reduced.

The question as to when the state point F is reached, is answered by observing the creep response immediately following the sudden drop in stress. At F the creep-strain rate is, according to this model, instantaneously zero following the stress reduction (Fig. 5 — curve II); it increases thereafter and finally reaches a new steady state value. This corresponds to a path such as FG in Fig. 4.

The creep responses predicted by the present model as the state point is changed abruptly from B to C and from B to H, are shown in Fig. 5 as the curves I and III, respectively. These are compared with the response corresponding to a stress reduction to state point F (curve II). Dropping the stress from B to points lying below F (e.g., H) results in a finite period $\Delta\tau$ of zero creep rate (curve III), the greater the stress drop the longer the hesitation period. Dropping to points above F (e.g., C), on the other hand, results in no hesitation period (curve I).

What would seem, on the basis of this idealized response, to be a reasonable procedure for determining $\Delta\sigma_o$ and hence S_{so}, approximately, is as follows: While under steady state creep at σ_o, suddenly reduce the stress from σ_o through a succession of increasing stress drops, each time reloading back to σ_o and allowing steady state conditions to be again realized. For example, the state path BCDEB in Fig. 4 might correspond to one such cycle. The stress reductions are successively increased until a critical hesitation period, say $\Delta\tau_c$, of a predetermined duration (finite but small) is observed just following the sudden unloading. The stress reduction necessary to produce the critical delay period $\Delta\tau_c$ is taken to be $\Delta\sigma_o$; correspondingly, S_{so} can be calculated using Eq. (14).

Several tests of this kind need to be conducted, with, for instance, the $k^{\underline{th}}$ test yielding the steady state creep rate $\dot{\varepsilon}_{sk}$ and the steady state internal flow stress S_{sk} corresponding to the applied stress σ_k. Since by Eq. (3),

$$\dot{\varepsilon}_{sk} = A(\sigma_k - S_{sk})^n \tag{15}$$

the coefficients A and n can then be determined for a "best fit" of the generated data.

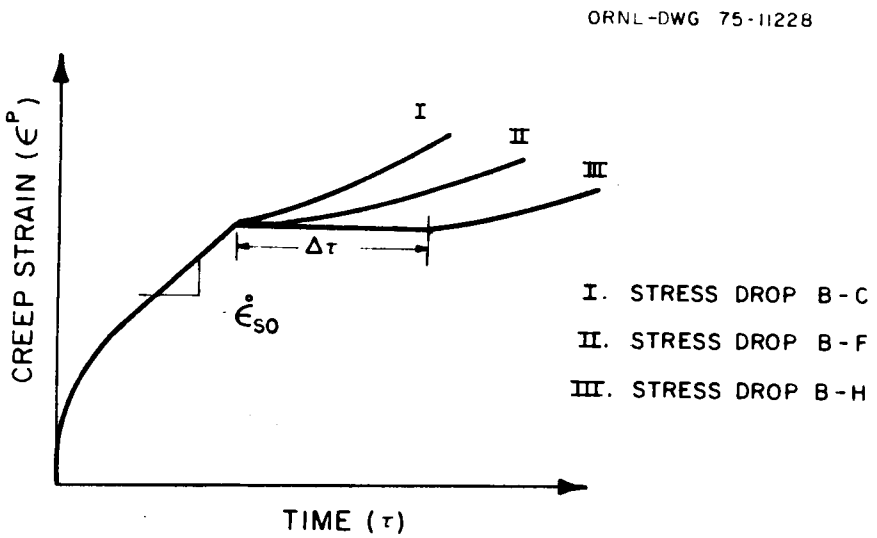

Fig. 5. Idealized creep strain response of the present model to abrupt stress reductions.

The present model does not account for creep-strain recovery, and its occurrence may, indeed, adversely affect the accuracy of the experimental procedure described above. This is not expected to cause difficulties of any great consequence however, since creep-strain recovery in 2 1/4 Cr—1 Mo steel is generally found to be quite small (see Fig. 3).

It will be assumed in the following that the material parameters C and D in Eq. (4) may be treated as constants, at least over a limited range of stress. How constant they remain over the stress range of interest is readily ascertained and if necessary they may be regarded as functions of stress and fit to the material of interest by performing a sequence of tests of the type described below.

Before discussing the additional experimentation, notice that under steady state creep conditions, i.e., with $\dot{S} = 0$ in Eq. (4), the following holds

$$\dot{\varepsilon}_s = (D/C)\ S_s^{\ 4}\ , \tag{16}$$

hence, the ratio D/C can be determined using the data already collected as described above. Also, the constancy of the ratio D/C with stress can be evaluated using the same data.

To determine the coefficient D individually, and thus C from Eq. (16), consider the following experiment. A uniaxial specimen is loaded into a steady state creep condition at a given stress level, say for convenience, the same stress as above, σ_o. This corresponds, once again, to the state point B in Fig. 4. However, now the coordinates of point B are assumed known. The stress is then dropped rapidly from σ_o, only this time by an amount sufficient to bring the state point well below point F (Fig. 4), say for example to point H at the reduced stress level σ_1. At H, $\sigma < S$, so that from Eq. (4),

$$\dot{S}\Big|_{\ } = -DS^3\ , \tag{17}$$

which remains valid as long as the state point remains below $\sigma = S$. According to the present model, the state point moves (with zero accumulated creep strain) from H toward point J at a rate dictated by the differential equation Eq. (17).

The solution of Eq. (17) corresponding to the state path H to J is

$$(1/S_J^2 - 1/S_H^2) = 2\ D\ \Delta\tau_{HJ}\ , \tag{18}$$

where S_H is the internal stress at H (here, $S_H = S_{so}$), S_J is the internal stress at state point J ($S_J = \sigma_1$) and $\Delta\tau_{HJ}$ is the duration of the observed hesitation period (curve III in Fig. 5). The coefficient D can thus be found as

$$D = \frac{(1/\sigma_1{}^2 - 1/S_{so}^2)}{2\ \Delta\tau_{HJ}} \tag{19}$$

C can then be determined from the known ratio D/C. In the event D varies significantly with σ_1, it may be considered a function of stress and fit to the material of interest using several such tests, dropping the stress in each case to a different level.

EXTENSION TO MULTIAXIAL STRESS STATES

A natural generalization of the scalar internal flow stress S is the internal flow stress tensor with Cartesian components S_{ij}. In terms of this quantity, it is possible to generalize Eq. (3) as follows

$$\dot{\varepsilon}_{ij}^p = f_{ij}(\sigma_{k\ell} - S_{k\ell})\ , \tag{20}$$

which states that the components of inelastic strain rate are functions of the so-called "effective stress tensor" $(\sigma_{k\ell} - S_{k\ell})$. Furthermore, if linearity in the inelastic strain rate is retained as in Eq. (4), the growth law can, in turn be generalized as

$$\dot{S}_{ij} = h_{ijk\ell}(S_{mn})\dot{\varepsilon}_{k\ell}^p - R_{ij}(S_{mn})\ . \tag{21}$$

The second term on the right hand side of Eq. (21) characterizes the recovery process and is regarded as a function of the components of internal flow stress.

It is interesting to note the formal resemblance of the present model, generalized in this way, to the theory of kinematic hardening in classical plasticity; the major difference occurs as a consequence of the recovery term in Eq. (21). Further study of the implications and interpretation of this observation is currently underway.

A more definite and more restrictive multiaxial representation can be set down for the present model by adopting a (Mises) J_2-form just as is done in the case of the currently recommended strain-hardening constitutive equations.[1,2] Such a J_2-formulation may be written as

$$\dot{\varepsilon}^p_{ij} = \begin{cases} (3/2)A(\bar{\sigma} - S)^n \, s_{ij}/\bar{\sigma} \; ; & \bar{\sigma} > S \\ 0 \; ; & \bar{\sigma} \leq S \end{cases} \tag{22}$$

and

$$S\dot{S} = C\dot{\bar{\varepsilon}}^p - DS^4 \; , \tag{23}$$

where S, as in the uniaxial formulation, is taken as a scalar measure of the state, s_{ij} are the components of the deviatoric stress tensor,

$$\bar{\sigma} = \sqrt{\frac{3}{2} s_{ij} s_{ji}} \tag{24}$$

and

$$\dot{\bar{\varepsilon}}^p = \sqrt{\frac{2}{3} \dot{\varepsilon}^p_{ij} \dot{\varepsilon}^p_{ji}} \tag{25}$$

SUMMARY AND CONCLUSIONS

No theory of inelastic material behavior is capable of accurately describing the behavior of even a single metal under all circumstances. Any physical or mathematical theory is, at best, an idealization applicable over a limited range of conditions. Input to the development of an engineering constitutive theory, therefore, must come not solely from isolated laboratory experimentation and principles of physics but from considerations regarding service conditions as well. A theory constructed for application under high loading rates, for example, may bear little resemblence to one intended for use under quasi-static loading of the same material.

As indicated above, the theory of strain-hardening provides good predictions of the creep response of 2 1/4 Cr—1 Mo steel under many loading situations at high temperature. As also pointed out, however, the theory suffers certain deficiencies in that it tends, under some stress histories, to over-predict hardening and moreover, does not account for the recovery of state which occurs under stress reductions. If these effects are judged significant, in light of the supposed service conditions encountered by 2 1/4 Cr—1 Mo steel reactor components, then they must be taken into account in the constitutive theory used in the structural analysis of these components.

The creep-recovery model considered here qualitatively predicts the apparent reduced hardening and state recovery effects addressed above and, furthermore, it provides for "strain-hardening like" behavior under conditions where the strain-hardening theory is known to accurately apply.

It is expected that the present model can be implemented on the basis of relatively simple macroscopic tests as outlined. To be sure, the testing called for may require somewhat more sensitive and sophisticated instrumentation than that necessary for standard creep tests, however, it seems reasonable to expect that more comprehensive constitutive laws may necessitate more complex experimentation. The feasibility of the proposed testing program must, of course, be demonstrated.

It is easily demonstrated that neither constant-stress creep tests nor pure relaxation tests, in themselves, constitute the necessary tools for complete exploration of the state space (Fig. 4) associated with the present, or for that matter, any creep-recovery model. This is equivalent to saying that creep and relaxation tests alone do not provide the basis for predicting the creep response to all possible loading histories. Evidently, hybrid creep-relaxation tests or recovery-like tests, such as those described above are necessary in this regard.

In spite of the many inherent assumptions, the so-called J_2 multiaxial form expressed in Eqs. (22) through (25) may well prove useful in predicting the creep response of 2 1/4 Cr—1 Mo steel under general stress states. In fact, these equations have already been implemented in the CREEP-PLAST computer code at ORNL by J. A. Clinard and predictions of response of this model under uniaxial and multiaxial stress states are

now being obtained. Many open questions still remain with respect to such a multiaxial formulation, e.g., the behavior under reversed stresses etc., nevertheless, it is encouraging to note the success the analogous J_2 form of the strain-hardening theory has enjoyed in its several applications to pertinent structural problems.

REFERENCES

1. Letter from D. B. Trauger (ORNL) to J. A. Lenhard (USERDA-ORO), *High-Temperature Structural Design Methods for 2 1/4 Cr–1 Mo Steel*, Jan. 17, 1975.

2. J. M. Corum et al., "Constitutive Equations for 2 1/4 Cr–1 Mo Steel," *High-Temperature Structural Design Methods for LMFBR Components Progr. Rept. Dec. 31, 1974*, ORNL-5048 (in preparation).

3. E. Orowan, "The Creep of Metals," *J. W. Scotl. Iron Steel Inst.* 54, 45-53 (1946).

4. D. L. Marriott, *A Theory of Creep Including the Effects of Thermal Recovery*, D/011/69, Univ. of Liverpool, Dept. of Mech. Engr. (April 1969).

5. J. H. Gittus, "A Model of Creep Embodying Dislocations Whose Movements Produce Work Hardening and Recovery," *Phil. Mag.* 21, 495-508 (March 1970).

6. R. Lagneborg, "A Modified Recovery-Creep Model and Its Evaluation," *Met. Sci. J.* 6, 127-33 (1972).

7. B. Modéer, *An Experimental Study of Creep in 20% Cr — 35% Ni Steel and Simulations of Creep Deformation by a Recovery-Creep Model*, Int. Rep. Div. Phys. Met., The Roy. Inst. of Tech., Stockholm, Sweden (1972).

8. S. K. Mitra and D. McLean, "Work Hardening and Recovery in Creep," *Proc. Roy. Soc. (London), Ser. A* 295, 288-99 (1961).

9. E. T. Onat and F. Fardshisheh, *Representation of Creep of Metals*, ORNL-4783 (August 1972).

10. J. D. Parker and B. Wilshire, "The Effect of a Dispersion of Cobalt Particles on High-Temperature Creep of Copper," *Met. Sci.* 9, 248-52 (1975).

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