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A substructure characterizing parameter in creep

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A substructure characterizing parameter which is the ratio of applied stress, σ, in creep test to yield stress, σ_{YS}, of the material at the test temperature is introduced. A correlation is found to exist between this parameter and the creep rate for the data obtained in the temperature range 820–975 K when the initial yield strength is modified by (i) introducing different amounts of prior cold work by two modes of deformation at room temperature in a type 316 LN stainless steel and (ii) grain size, chemistry and grain size variation in a type 316 stainless steel. The correlation was found to exist also for a Cr–Mo–V steel at 823 K, in which different yield strengths were due to different heat treatments. Minimum creep rate when plotted against the substructure characterizing parameter yields an exponent similar to Norton's creep exponent and it is postulated that the value of the exponent reflects on the type of substructure developed in creep. Another parameter σ/F_{ys} where F_{ys} is a function of the ratio of the yield strength of a given microstructure to that of a reference microstructure (zero cold work for cold worked material, largest grain size when the microstructure variation is through grain size and solution annealed microstructure among heat treatments) also gives a unique correlation with the minimum creep rate at a test temperature with the exponent identical to Norton's creep exponent.

1. Introduction

Plastic deformation of a material occurs when the applied stress is sufficient to enable the mobile dislocations to overcome the inherent obstacle structure in the material resisting the motion of dislocations. The initial structural obstacle to mobile dislocations can be considered to be other dislocations, solutes, second phase particles etc. arranged in a three-dimensional irregular network of mobile dislocation segments. The mobile dislocation segments must free themselves from the nodes in the network to be able to participate in the glide process. In the dislocation creep regime, the generation of mobile dislocations and their glide motion represent the process of creep. The minimum creep rate, $\dot{\varepsilon}_{min}$ during creep is often correlated with the applied stress at a given temperature by the equations

$$\dot{\varepsilon}_{min} = A_1(\sigma/G)^{n_1} \quad (1)$$

$$\dot{\varepsilon}_{min} = A_2\tau^{3.5}(\sigma/G)^{n_2} \quad (2)$$

$$\dot{\varepsilon}_{min} = A_3(\sigma - \sigma_0)^{n_3} \quad (3)$$

$$\dot{\varepsilon}_{min} = A_4[(\sigma - \sigma_0)/\sigma_{ys}]^{n_4} \quad (4)$$

where A_i are constants, n_i are creep exponents, τ is the stacking fault energy, σ is the applied stress, σ_0 is the internal stress, σ_{ys} is the appropriate yield or proof stress and G is the shear modulus.

It is known that microstructure influences the creep rate. The first two relations do not take this into account (the shear modulus is not influenced by substructure). There is general acceptance that the mechanical deformation of materials at elevated temperatures is not driven by the full applied stress, σ, but rather by an effective stress, $(\sigma - \sigma_0)$ where σ_0 is the back stress opposing the motion of dislocations [1–8]. The friction stress σ_0 represents the deformation resistance of the microstructure. In pure metals σ_0 might arise from subcells or tangles [7] while in precipitation-hardened systems a major factor will be the precipitates themselves [9, 10]. σ_0 also may depend on temperature, strain, composition, thermomechanical history, imposed stress and strain rate. The stress $(\sigma - \sigma_0)$ controls the creep rate by determining the size of the three-dimensional dislocation network developed during creep [11]. The creep data for a variety of materials may be superimposed [10] by dividing by the appropriate yield or proof stress. σ_{ys} is a function of the substructural features like stacking faults, dislocation density, interparticle spacing, grain size, texture, cell or subgrain size etc. influencing the creep deformation process. The description of the high temperature deformation in terms of the effective stress $(\sigma - \sigma_0)$ is based on the need to rationalize the commonly observed high stress dependence on the creep rate which cannot be explained otherwise. The latter

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two relations in some sense account for the microstructural dependence of creep rate. In this case the temperature dependence would arise from the effect of temperature on σ_0 and σ_{ys}.

It was shown by Ajaja [12] that the high temperature flow strength of aluminium during steady state creep (measured in a tensile test on a specimen in which creep test was interrupted) was higher than the applied stress. Similarly Alden [13] showed that the low temperature flow stress of precrept lead samples increased with prior creep strain. Mathew *et al.* [14] showed that after prior creep deformation at 923 K, the room temperature yield strength and tensile strength of a type 316 stainless steel increased with increasing creep time while thermal ageing alone at 923 K did not lead to any significant changes in room temperature yield or tensile strength. There is evidence for work hardening during creep deformation. Kestenbach *et al.* [15] studied the development of substructure during creep of a type 316 stainless steel in the temperature range 873–973 K. Their results suggest that it is the stress level which controls the development of substructure at a given temperature during creep in this material. The creep-induced substructure under high applied stresses was similar to that found after tensile tests in this temperature range. From the available transmission electron microscopy results they have concluded that at 873 K, subgrains formed below an applied stress of 200 MPa and a subgrainless dislocation substructure for applied stress levels above 200 MPa. At this transition stress level they have noted that the measured minimum creep rate would be roughly equal to 10^9 D where D is the coefficient of self-diffusion.

Since the initial substructure determines the creep deformation under an applied stress, it is thought to be appropriate and meaningful to compare the creep properties of a material having substructural modifications (1) through prior cold work (2) through grain size and heat to heat variation, and (3) by heat treatments at a given temperature in terms of a parameter which characterizes the relative strength of the initial substructure and gives an indication of the ease of dislocation mobility. A correlation of minimum creep rate with this parameter may throw more light on the nature of the substructure developed at steady state. In this paper the dependence of minimum creep rate of a type 316 LN stainless steel subjected to different levels of prior cold work, a type 316 stainless steel with different grain sizes, and from different heats, and a Cr–Mo–V steel subjected to different heat treatments is reported in terms of the substructure characterizing parameter.

2. Substructure characterizing parameter

In all materials, the mechanism of creep deformation depends on the operative stress level. At high creep stresses ($> \sigma_{ys}$) the grains will deform and the matrix deformation processes are important. At low stresses, low strain rates and high temperatures, grain boundary phenomena are considered dominant in determining the creep properties.

The yield strength of the material can be thought of as a measure of the strength of the initial substructure. A given material can be strengthened by grain refinement or by prior cold work (PCW) or by suitable heat treatments. In all these cases the yield strength of the material is changed through substructural modifications. In grain refinement the dislocation density increases as the grain size is reduced. In PCW material the strengthening is ascribed to the increased dislocation density and their arrangement depending on the amount of cold work. The mean free path of mobile dislocations decreases with an increase in PCW and the mobile dislocation link length may be reduced, such that a higher applied stress is needed for their mobilization and subsequent participation in the glide process. By suitable heat treatments the amount of precipitates and phases present in the virgin material is altered, which could change the yield strength and the interaction forces on dislocations compared to the virgin material. As the plastic deformation proceeds, dislocations encounter other dislocations and obstacles and a new obstacle structure evolves. The evolution of substructure depends on the crystal structure and stacking fault energy (SFE), in addition to the imposed parameters like temperature, stress and strain rate. It also depends on the initial state of the substructure.

In a work-hardening condition where the flow strength increases with deformation, flow strength can be considered as a measure of the current strength of the obstacle structure or the substructure produced at that strain. This is particularly true since the material will normally start yielding at the same strength level on unloading and reloading. The ratio of flow strength $(\sigma)_\varepsilon$ at a given strain to the initial yield stress σ_{ys} is considered as an index of the work hardening in the material at a given temperature and structure or as a relative measure of the strength of the evolved obstacle structure relative to the original structure. Since the yield strength and flow strength are characteristic of the initial and current dislocation substructures, their ratio could be considered as a substructure characterizing parameter.

In high temperature creep deformation, thermal activation in addition to the applied stress assists the mobile dislocations in overcoming the obstacle structure and creep is possible at stress levels below the yield stress of the material. Evolution of the substructure during creep will depend on the strength of the initial substructure (yield strength, σ_{ys}). A steady state dislocation substructure results from dynamic equilibrium of generation (work hardening) and annihilation (recovery) of dislocations resulting in a steady state creep rate. When a specimen is loaded in creep, an instantaneous plastic deformation occurs and the flow strength of the material is equal to the applied stress and therefore the initial value of the substructure characterizing parameter is the ratio of the applied stress to the yield stress of the material at that temperature. The activation of dislocation sources by the applied stress at the test temperature to cause creep strain will depend on the magnitude of the applied stress relative to the yield stress. During further creep

TABLE I Chemical composition (wt%) of various materials investigated in the study

	C	Mn	Ni	Cr	Mo	N	S	P
316	0.06	1.93	12.60	16.36	2.22	0.0335	< 0.01	0.034
316 (A)	0.057	1.8	12.51	16.7	2.23	0.045	0.013	0.025
316 (B)	0.057	1.65	12.44	16.62	2.32	0.032	0.007	0.025
316 (C)	0.048	1.73	12.48	16.1	2.11	0.031	0.016	0.030
316 LN	0.021	1.74	12.0	17.0	2.40	0.078	0.02	0.023
Cr–Mo–V	0.11	0.30	–	1.01	0.28	0.023	0.007	0.01

deformation, the material undergoes work hardening and recovery simultaneously and at steady state a balance is obtained and a characteristic substructure is evolved. The relative strength of the substructure developed at steady state could be determined by tensile testing a specimen crept up to the steady state region and by determining the ratio of the yield stress of the crept specimen to the yield stress of a specimen having the initial substructure at the test temperature.

In a specimen that has already undergone rupture, as there is no way to find the relative strength of the substructure at steady state, the value of the ratio of applied stress, σ (the creep test stress) to the yield stress at the test temperature, σ_{ys}, is considered as the initial value of the substructure-characterizing parameter and it is assumed that further substructure development depends on this parameter.

3. Experimental procedure

The chemical composition of the materials used in this investigation is shown in Table I. The details of the treatments to produce various grain sizes were described elsewhere [17]. Type 316 LN material was subjected to 10, 20 and 30% cold work at room temperature by two modes, rotary swaging and by tensile deformation by pulling in an Instron model 1195 universal testing machine. The amount of cold work was measured as the percentage reduction in cross-sectional area. Creep specimens were fabricated with their tensile axis in the original rolling direction of the plate or in the direction of deformation in the case of swaged or tensile deformed specimens. Creep tests were carried out in an EMEC multilever creep testing unit with lever ratio 20:1. Creep elongation was measured using a dial gauge accurate to 0.005 mm. Test temperature was controlled within 1 K. Data on type 316 stainless steel from the earlier work in our laboratory [16, 17] and on Cr–Mo–V steel [8] from the literature have also been analysed according to the concept described earlier.

4. Results and discussion

4.1. Yield strength modified by prior cold work

Cold working introduces plastic deformation which results in increased dislocation density and substructural changes. These changes increase the yield strength of the material; the percentage increase in yield strength at 300, 873 and 948 K after the two modes of prior cold working is shown in Table II. The

TABLE II Percentage increase in yield strength at various temperatures for prior cold-worked type 316 LN stainless steel

		% increase in yield strength		
Method of PCW	%PCW	300 K	873 K	948 K
Tensile	10	65.3	81.4	70.3
	20	120.7	151.4	119.8
	30	160.8	200.5	141.8
Swaging	10	94.6	102.2	75.8
	20	135.6	162.3	130.7
	30	162.4	206.0	147.3

variation of the relative increase in strength of the initial substructure as a function of temperature for different amounts of prior cold work is shown in Fig. 1. These changes should reflect on the creep performance of the material.

Dislocation substructure developed after prior cold work can be either tangles, cells or subgrains depending upon the amount of deformation and the working temperature. Ajaja and Ardell [18–21] based on their study of creep behaviour of prior cold-worked type 304 stainless steel conclude that the important microstructural feature governing the creep behaviour at 950 and 1023 K is the dislocation density. The initial dislocation substructure developed after 20% PCW in type 316 LN stainless steel at room temperature has been reported [22] to contain dense dislocation tangles. The mode of cold work may also be expected to show some variation in the dislocation population or cell size which may affect the relative strength of the substructure.

The results of creep tests on type 316 LN stainless steel deformed to different amounts of cold work (0–30%) by tensile and swaging modes of deformation and tested at 948 K in the stress range 160–220 MPa are shown in Fig. 2. The minimum creep rate decreases with increase in PCW for both modes of cold work. The creep exponent was found to be 8.5 irrespective of the amount of PCW. The decrease in minimum creep rate with increase in PCW is attributed to the decrease of mobile dislocation density with increasing amount of PCW and the relative strength of the different initial dislocation substructure developed prior to creep testing and the consequent steady-state substructure evolved. This is evident when the above results are plotted against (σ/σ_{ys}) and shown in Fig. 3. For as-received and cold-worked materials the slopes are comparable but data points shift to lower values of abscissa as the cold work is increased. Mode of cold work also influences the initial substructure. This is

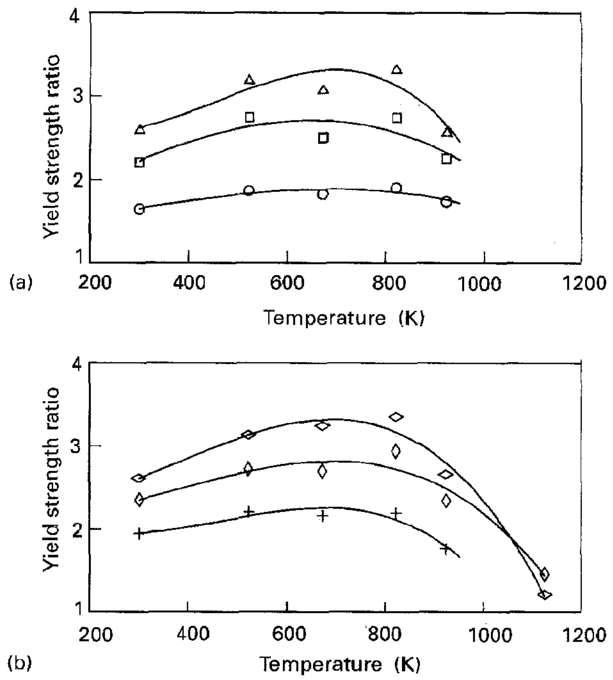

Figure 1 Dependence of yield strength ratio with temperature for different amounts of prior cold work (PCW). (a) PCW by tension (O—O 10%, □—□ 20%; Δ—Δ 30%); (b) PCW by swaging (+—+ 10%; ◊ —◊ 20%; ◇—◇30%).

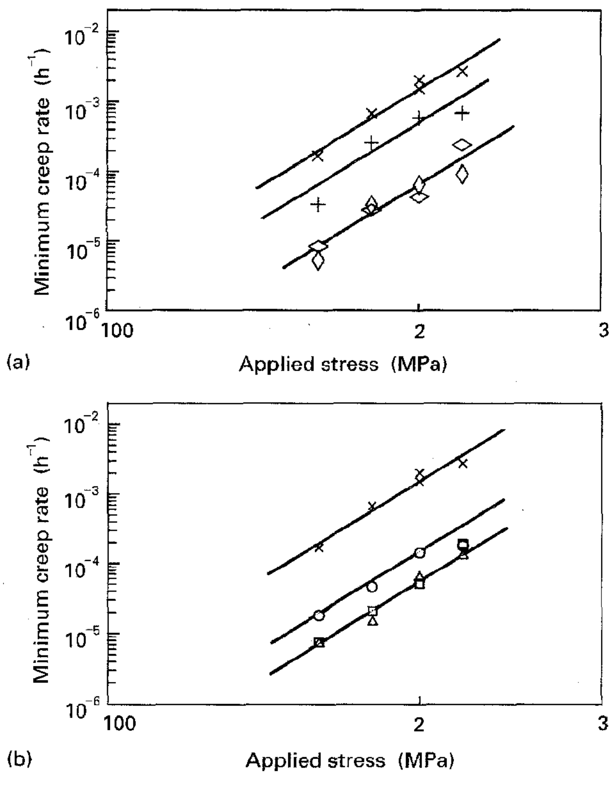

Figure 2 Dependence of minimum creep rate with applied stress for different amounts of prior cold work (PCW) at 948 K. (a) PCW by tension (×—× 0%; +—+ 10%; ◊ —◊ 20%; ◇—◇30%) (b) PCW by swaging (×—× 0%; O—O 10%; □—□ 20%; Δ—Δ 30%).

evident from the percentage increase in the room-temperature and high-temperature yield strength as shown in Table II.

The exponent obtained from the plot of minimum creep rate versus substructure characterizing parameter ($n = 8.5$) was the same as the Norton's creep exponent correlating the minimum creep rate and applied stress. However, it can be noted that as the cold work increases, the minimum creep rate decreases for a given applied stress. In this context one can correlate the minimum creep rate with another parameter, σ/F_{ys} as

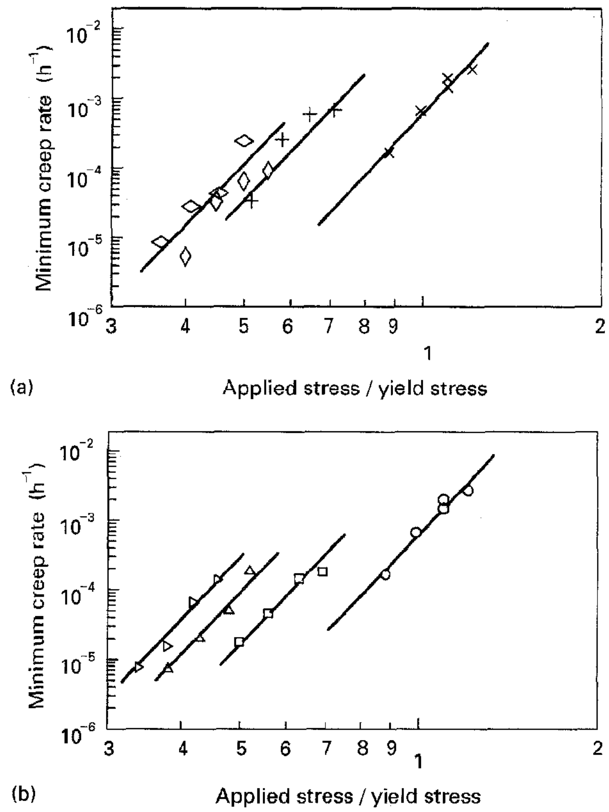

Figure 3 Dependence of minimum creep rate with ratio of applied stress to yield stress for different amounts of prior cold work (PCW) at 948 K (a) PCW by tension (×—× 0%; +—+ 10%; ◊ —◊ 20%; ◇—◇ 30%). (b) PCW by swaging (O—O 0%; □—□ 10%; Δ—Δ 20%; ▷—▷ 30%).

$$\dot{\varepsilon}_{min} = A(\sigma/F_{ys})^n \qquad (5)$$

where F_{ys} is a function of the ratio of the yield strength of a given microstructure to that of a reference microstructure. In the case of cold-worked material the reference microstructure is taken as that of zero percent cold-worked material. If we consider the function as a power function of the form [23]

$$F_{ys} = (\sigma_{ys(cw)}/\sigma_{ys(ref)})^m \qquad (6)$$

where m is a constant which may depend on the test temperature (in the present case m is found to be 0.4). The relation between the minimum creep rate and the ratio σ/F_{ys} yields a unique relation with a good linear fit on a log–log plot with a slope equal to 8.5 as shown in Fig. 4. For the material investigated the above relation can be used to determine the minimum creep rate at any given stress level within the range of the cold work considered.

4.2. Yield strength modified by grain size variation

In type 316 austenitic stainless steel, the influence of grain size on creep rate was more pronounced at

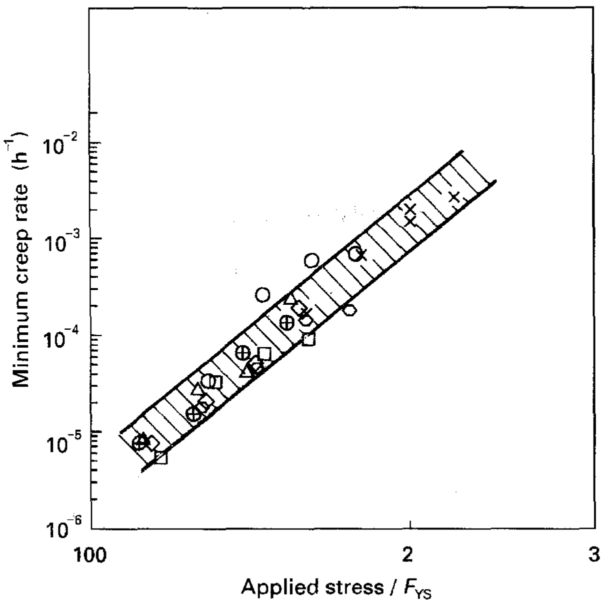

Figure 4 Variation of minimum creep rate with the ratio (σ/F_{ys}) at 948 K. (T) PCW by tension; (S) PCW by swaging. (×—× 0%; ○—○ 10%; (T); □—□ 20%; (T); △—△ 30%). (T); (⬡—⬡ 10% (S); ◇—◇ 20% (S); ⊕—⊕ 30% (S)).

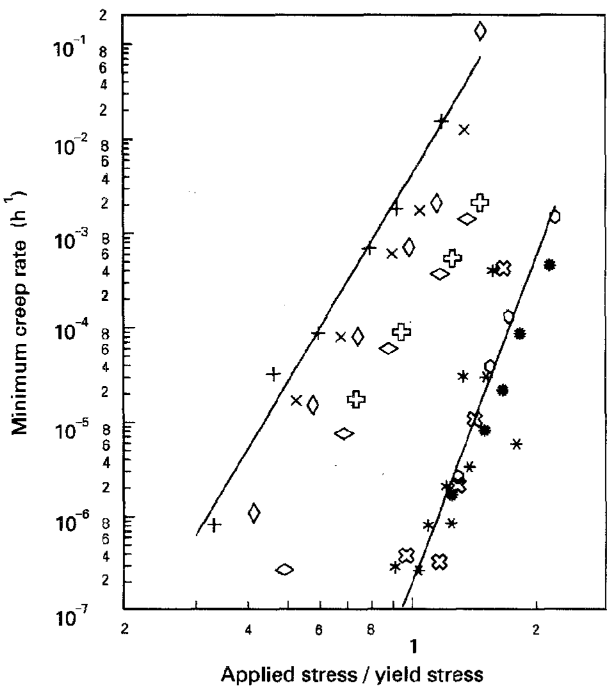

Figure 5 Variation of minimum creep rate as a function of the ratio of applied stress to yield stress at 873 K (✠—✠ 0.040 mm; ∗—∗ 0.060 mm ×—× 0.125 mm; ●—● 0.270 mm; ⬡—⬡ 0.650 mm) and 973 K (×—× 0.040 mm; +—+ 0.060 mm; ◊—◊ 0.125 mm; ◇—◇ 0.270 mm; ✚—✚ 0.650 mm) for different grain sizes.

873 K than at 973 K [24]. At 873 K creep rate was found to increase with increasing grain size in the stress range 150–260 MPa while at 973 K, creep rate was not significantly influenced by grain size in the stress range 90–140 MPa. However, at an applied stress level of 70 MPa there is a definite minimum observed in the creep rate grain size curve [24]. The dislocation substructure developed at 873 K was dislocation tangles along with uniform precipitation of $M_{23}C_6$ particles throughout the matrix. At 973 K subgrains were observed at the lower applied stress. This difference in substructure was attributed as a possible reason for the grain size dependence of creep rate [16].

Since the dislocation substructure developed at steady state, creep is more fundamental than the grain size *per se* in determining the steady-state creep rate; a plot of minimum creep rate against the substructure characterizing parameter at steady state should yield a unique relation irrespective of the grain size. Fig. 5 shows the log–log plot of the variation of minimum creep rate with σ/σ_{ys} for type 316 stainless steel over a wide range of grain sizes. The minimum creep rate decreases with decreasing σ/σ_{ys}. The data for all grain sizes at 873 K fall within a small scatter band with a slope of 13. At 973 K the data show a grain size dependence, especially for the two largest grain sizes. However, for all grain sizes a power law with an exponent of 8 is valid between creep rate and σ/σ_{ys}. No systematic variation of Norton's creep exponent with grain size was observed. The values of Norton's creep exponent (i.e. slope of log creep rate and log σ) reported was in the range 10–14 at 873 K and in the range 6–8 at 973 K.

At 873 K, the dislocation substructure contained mainly dislocation tangles irrespective of the grain size. The relative strength of the dislocation substructure depends also on the precipitation of carbides in the matrix and on dislocations. At 873 K uniform precipitation of fine and cuboidal carbide particles throughout the austenite matrix was observed. The size of the carbides was found to increase as the applied stress is decreased and as the test temperature is increased. The substructure developed at 973 K were mainly cells or subgrains [24]. Then, the relative strength of the substructure at this temperature will depend on the cell or subgrain size and may also be influenced by the carbide precipitation. This could be a possible reason for the grain size dependence observed at this temperature.

The yield strength of the coarsest grain size ($\sigma_{0.65}$) is taken as the strength of the reference state and (F_{ys}) was calculated as $(\sigma_d/\sigma_{0.65})^m$ where σ_d is the yield strength of the material having a grain size d and with $m = 0.4$ at 973 K and $m = 0.2$ at 873 K. Fig. 6 shows a unique relation between minimum creep rate with (σ/F_{ys}) where σ is the applied stress for type 316 stainless steel tested at the two temperatures. The minimum creep rate decreases with decreasing (σ/F_{ys}). The data for all grain sizes at 873 K and 973 K fall within a small scatter band but have different slopes. However, for all the grain sizes, a power law between creep rate and σ/F_{ys} is valid at 873 and 973 K with an exponent of 13 and 8 respectively. The value of the exponent when creep rate is plotted against the substructure characterizing parameter is found to be more or less equal to Norton's creep exponent. A value of the creep exponent, $n > 5$, suggests that a climb-controlled dislocation creep mechanism is operative. In the present case the value is found to be 12 ± 2 at 873 K for a dislocation tangled substructure and 7 ± 2 at 973 K for a substructure

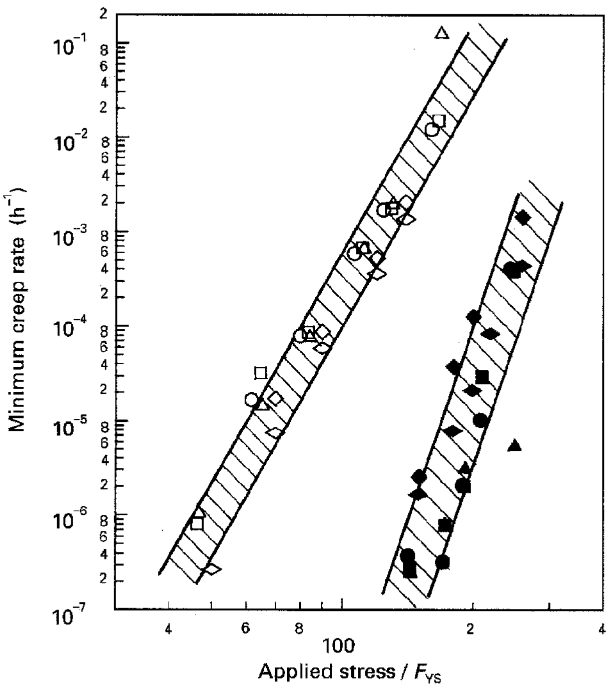

Figure 6 Dependence of minimum creep rate with the ratio of (σ/F_{ys}) at 873 K (●—● 0.040 mm; ■—■ 0.060 mm; ▲—▲ 0.125 mm; ◆—◆ 0.270 mm; ◆—◆ 0.650 mm) and 973 K (○—○ 0.040 mm; □—□ 0.060 mm; △—△ 0.125 mm; ◇—◇ 0.270 mm; ◇—◇ 0.650 mm) for different grain sizes.

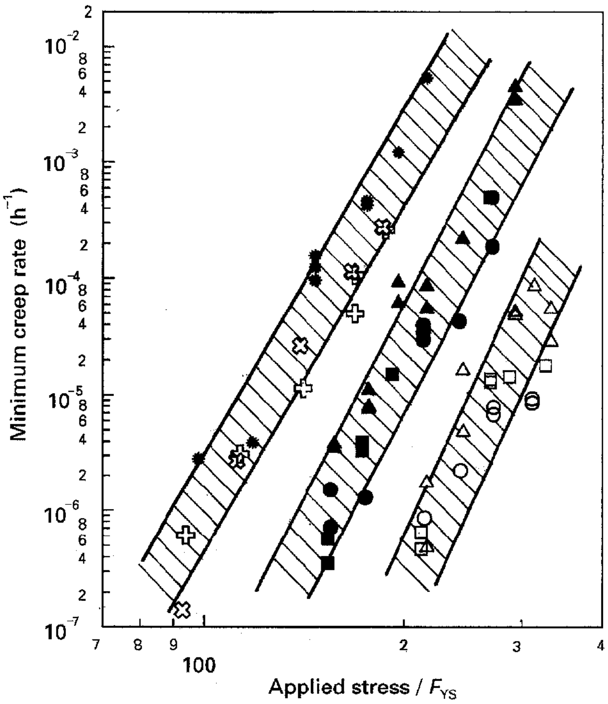

Figure 7 Dependence of minimum creep rate with the ratio of (σ/F_{YS}) at 823 (○ Heat A; □ Heat B; △ Heat C), 873 (● Heat A; ■ Heat B; ▲ Heat C) and 923 K (✣ Heat A; ✣ Heat B; * Heat C) for the three heats of type 316 stainless steel.

characterized by dislocation cells. We postulate that the value of the exponent will be of the order of 4 at still higher temperature if the dislocation substructure is characterized by subgrains.

4.3. Yield strength modified by heat to heat variation

Now we look at the results from three heats of type 316 austenitic stainless steel creep-tested at 823, 873 and 923 K at various stress levels [25, 26]. The three heats were different mainly in the composition variation of minor elements and grain size (the linear intercept grain sizes of the heats A, B and C were 0.040 mm, 0.060 mm and 0.080 mm respectively). These differences resulted in a variation in the yield strength at the test temperature between the heats. The weakest heat is taken as the reference state. The minimum creep rate plotted against the parameter σ/F_{ys} is shown in Fig. 7. The data points for a particular temperature for all the three heats group together. The slopes of the plot are similar to the Norton's creep exponent reported for the three heats at the test temperatures (see Table III). The substructural features observed in the material at three test temperatures were mainly dislocation tangles, though a tendency for cell formation is observed at very low stresses at 923 K.

4.4. Yield strength modified by heat treatment

Heat treatments of ferritic steels change the strength and the microstructure of the material. Creep data as well as hot tensile data generated on a

TABLE III Value of stress index for various heats at different temperatures [17]

Temperature (K)	Creep exponent		
	Heat A	Heat B	Heat C
823	9	9	11
873	11	10	10
923	8	8	8

1Cr–0.3Mo–0.25V steel, with a wide range of microstructure prepared from virgin state as well as on the material in the service-exposed condition are available in the literature [8]. At the creep temperature of 823 K, the yield strength ratio was maximum for a 100% bainitic structure (RHV1 (H_v 210)) and decreases with 80% ferrrite + 20% bainitic (RHV2 (H_v 190)), 80% ferrite + 20% tempered bainitic (RHV3 (H_v 160)) and 90% ferrite + 10% tempered bainitic (SE (H_v 152)) structures. The microstructure of the virgin material was 90% ferrite + 10% bainitic structure with hardness, H_v 156. Variation of the yield strength ratio with respect to the virgin state of the material at different temperatures is shown in Fig. 8. The yield strength ratio increases with increase in temperature initially and then decreases.

The variation of minimum creep rate with the substructure characterizing parameter σ/σ_{ys} and the parameter σ/F_{ys} at the creep temperature is shown in Figs 9 and 10 respectively. When plotted against the substructure characterizing parameter, the data for identical microstructure (comparable initial hardness and yield strength) fall in a single curve with a slope 9.5. This could imply that the substructure which governs the creep rate could be an identical one. In the

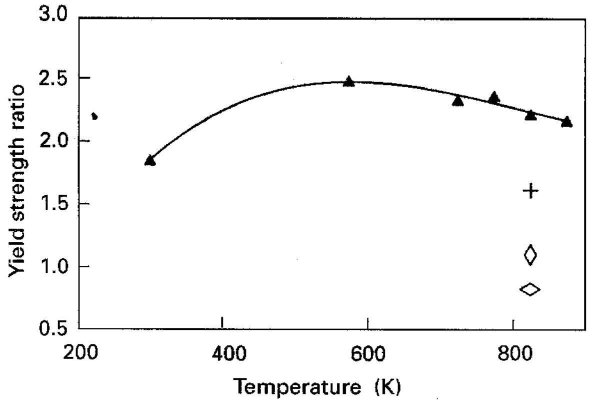

Figure 8 Dependence of yield strength ratio with temperature for a Cr–Mo–V steel with different microstructure (▲ RHV1; + RHV2; ◊ RHV3; ◇ SE) [8].

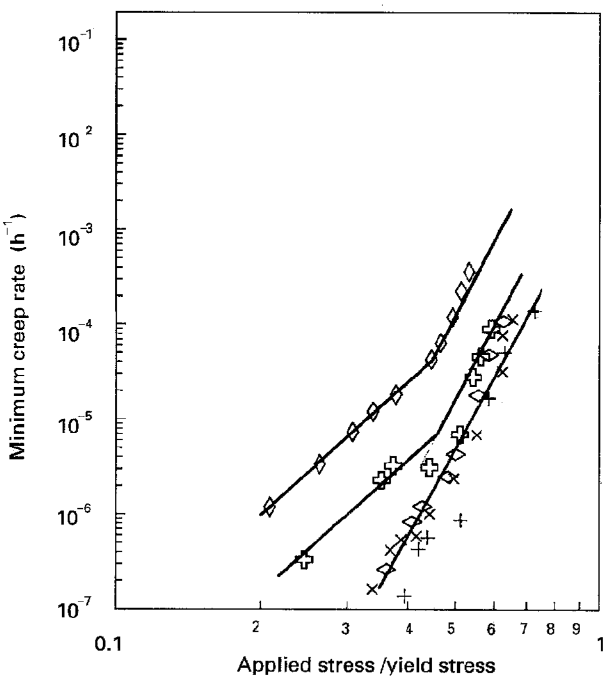

Figure 9 Variation of minimum creep rate as a function of the ratio of applied stress to yield stress for a Cr–Mo–V steel with different microstructure (× VN; + SE; ◊ RHV1; ✙ RHV2; ◇ RHV3)

case of RHV1 and RHV2, the material showed a slope of 9.5 at stress ratio above 0.4 while below the stress ratio of 0.4, the slope was 4.5. Minimum creep rate, when plotted against the parameter σ/F_{ys}, shows an almost unique correlation and the data fall in a single curve with a slope of 9.5. This could imply that the substructure which governs the creep rate could be an identical one.

From these analyses it is observed that the variation of minimum creep rate with the substructure characterizing parameter shows three regions having distinct slopes: (a) greater than 12, corresponding to a tangled dislocation substructure; (b) around 6–8 for a dislocation cell substructure; and (c) possibly around 4 for a subgrain substructure.

5. Conclusions

A parameter (σ/σ_{ys}) where σ is the applied creep stress and σ_{ys} is the yield strength of the material at the test

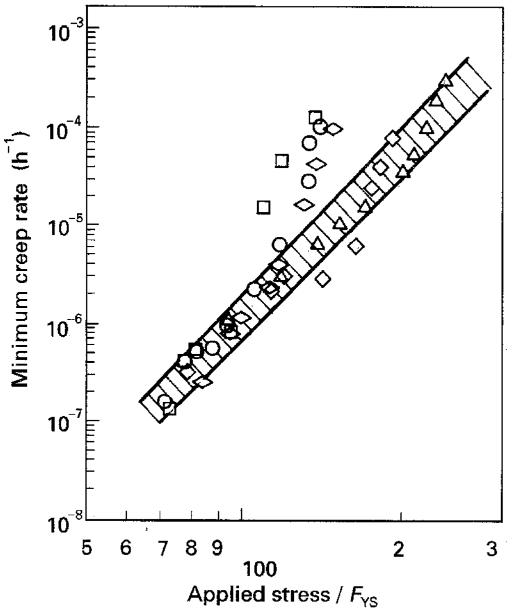

Figure 10 Variation of minimum creep rate with the ratio (σ/F_{ys}) and a Cr–Mo–V steel with different microstructure (○ VN; □ SE; △ RHV1; ◇ RHV2; ◇ RHV3) [8].

temperature, which characterizes the relative strength of the dislocation substructure is shown to correlate with the minimum creep rate under creep deformation conditions for (i) type 316 stainless steel whose yield strength is modified by grain size and through chemistry variation, (ii) type 316 LN stainless steel subjected to prior cold work and (iii) a Cr–Mo–V ferritic steel where yield strength modifications were done by heat treatment. It is postulated that the substructure developed at the creep temperature depends on this parameter and governs the minimum creep rate. Minimum creep rate when plotted against the substructure characterizing parameter yields an exponent similar to Norton's creep exponent and it is postulated that the value of this exponent could be assigned to the type of substructure developed in creep. A unique correlation is also observed between minimum creep rate and a parameter σ/F_{ys}, where σ is the applied creep stress and F_{ys} is a function of the ratio of the yield strength of a given microstructure to that of a reference microstructure (zero cold work for cold-worked materials, largest grain size when microstructural variation is through grain size and solution-annealed microstructure among heat treatments) at a given test temperature with the exponent identical to Norton's creep exponent.

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