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Effect of Grain Size, Grain Shape and Subgrain Size on High Temperature Creep Behaviour

G. MALAKONDAIAH & P. RAMA RAO
Defence Metallurgical Research Laboratory, Hyderabad - 500258

Abstract. Several strengthening machanisms, namely solute strengthening, precipitation and fine particle strengthening, grain size as well as shape control and substructure strengthening, have been utilised in the deve'opment of creep resistant alloys. Of these, the effect of polycrystal grain size, grain shape and subgrain size on viscous as well as power-law (climb) creep processes are considered here. Viscous creep processes include Nabarro-Herring (N-H), Coble and Harper-Dorn (H-D) creep. N-H and Coble creep processes are strongly influenced by grain size whereas creep rates under H-D creep are independent of grain size. Though it is usually assumed that power-law creep is independent of grain size, in practice it is not so. Several investigations have revealed that power-law creep does depend on grain size. There are detailed and well understood results pertaining to the role of grain size in viscous creep. On the other hand, subgrain size has been seen to have effect only on the climb creep behaviour which is yet to be clearly understood.

1. Introduction

It is a privilege to have this opportunity to contribute to the special issue of the Defence Science Journal being brought out in honour of Dr. Raja Ramanna who has just completed his sixtieth year. As a pioneer of the Atomic Energy Programme, Dr. Ramanna has unflinchingly encouraged attention to original research in the related basic sciences as essential to the development of technology. Dr. Ramanna has been involved, as former Director, Bhabha Atomic Research Centre, as former Scientific Adviser to Raksha Mantri and as Chairman, Atomic Energy Commission, at the very helm of organisations indigenously developing modern technologies. In engineering metallurgy, the most important scientific development in recent times has been the understanding of alloy behaviour in terms of internal structure. High temperature creep of components involving time dependent deformation is a phenomenon of much concern in several technologies where long-term reliable service is vital, e.g. nuclear reactors and aircraft engines. The subject matter of this paper pertains to the influence that microstructural features such as polycrystal grain size,

grain shape and subgrain size have on high temperature creep behaviour of metallic materials.

At stresses less than about 10^{-5} G, where G is the shear modulus, and temperatures greater than about 0.4 T_m, where T_m is the absolute melting temperature, linear viscous creep processes, namely Nabarro-Herring (N-H), Coble or Harper-Dorn (H-D) creep dominate the flow[1]. Nabarro-Herring[2,3] and Coble[4] creep processes consider the deformation arising due to transport of matter from grain boundaries subjected to compression to those subjected to tension either through the lattice (N-H) or through the grain boundaries (Coble). N-H and Coble creep processes are generally referred to as diffusional creep processes. On the other hand, the underlying mechanism of H-D[5] creep is believed to be based on dislocation activity[6-8] which is yet to be clearly understood. With further increase in stress, i.e. in the stress range 10^{-5} to 10^{-3} G, the flow process changes from viscous to dislocation climb controlled power-law creep. Of these creep processes, climb creep is the one which has been widely studied[9]. It is only during the last two decades that the technological importance of viscous creep has been realised, especially in view of the fact that the extrapolation techniques to predict long term service behaviour based on short term laboratory creep tests lead to erroneous results. This is because the extrapolation techniques do not account for the contributions to strain from viscous creep processes that dominate at low stresses.

The effect of polycrystal grain size on the steady state climb creep has been studied by a number of investigators. It is generally observed that the creep rate at grain sizes (refers to grain diameter, d) less than about 100 μm increases with decrease in grain size[10-12]. Experimental results at coarse grain sizes are conflicting. Grain size independent creep behaviour[13-17] as well as strong grain size dependence of creep rate[10-12] have been reported. On the other hand, theory as well as experimental results unequivocally suggest that the polycrystal grain size strongly influences diffusional creep processes[1].

In a recent analysis, Burton and Knowles[18] have shown that diffusional creep represents an asymptotic limit of climb creep when the mean subgrain size (λ) approaches the grain diameter (d) at low stresses. Since λ depends on stress (this aspect will be further discussed in a later section) and d is stress independent, behaviour over the stress range encompassing diffusional (N-H creep) as well as power-law creep can be described by a critical dimension L_c as

$$\dot{\epsilon} kT/D_v \sigma b = (b/L_c)^2 \tag{1}$$

where $\dot{\epsilon}$ is the creep rate, σ the applied stress, D_v the lattice diffusivity, b the Burgers Vector and kT has the usual meaning. L_c takes the value of λ at higher stresses i. e. $\sigma > Gb/d$ and the value d under the opposite condition. This emphasises the importance of subgrain size under climb creep conditions.

Having noticed the importance of grain size under diffusional as well as climb creep conditions and subgrain size under climb creep conditions, the influence of grain size as well as subgrain size on steady state creep rate is considered in greater detail in the following Sections. The effects of grain shape on steady state creep behaviour and substructure on transient creep under viscous as well as power-law creep conditions are also discussed.

2. Influence of Grain Size on Creep Behaviour

2.1 *Viscous Creep Processes*

Nabarro-Herring[2,3] and Coble[4] theories assume grain boundaries to be perfect sources and sinks for vacancies and the diffusion to be the rate controlling processs. The resulting strain rates have been derived to be

$$\dot{\epsilon}_{NH} = A_{NH}\ \Omega\ D_v\ \sigma/d^2\ kT \tag{2}$$

$$\dot{\epsilon}_{Co} = A_{Co}\ w\Omega\ D_B\ \sigma/d^3 kT \tag{3}$$

where *N-H* and *Co* represent N-H and Coble creep processes respectively. A is a dimensionless constant, Ω the atomic volume, w the effective grain boundary width and D_B the grain boundary self diffusivity. Equations (2) and (3) indicate that creep rates under N-H and Coble creep conditions are inversely proportional to square and cube of the grain size respectively. It means a ten-fold increase in grain size will decrease N-H creep rate by hundred-fold whereas the corresponding reduction in creep rate under Coble creep conditions will be thousand-fold. Grain coarsening thus imparts significant creep resistance to a material under low stress service conditions where diffusional creep processes dominate the flow. Strong grain size dependence of diffusional creep rates, in accordance with the theoretical predictions, has been reported by several investigators[1]. Typical plots revealing inverse square and inverse cube grain size dependence of creep rate under N-H and Coble creep conditions are shown[19] in Figs. 1 and 2 respectively.

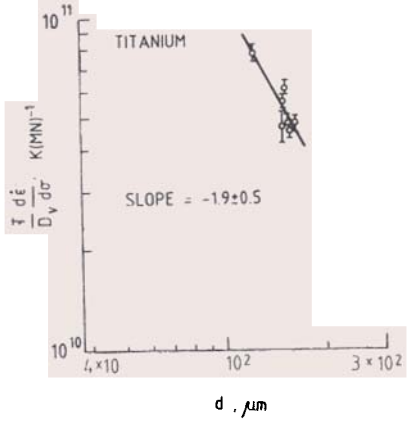

Figure 1. Temperature and lattice diffusivity compensated creep rate per unit stress as a function of grain size.

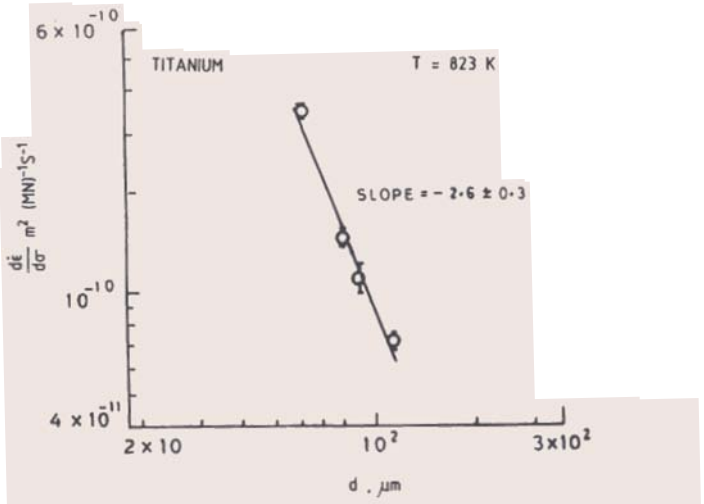

Figure 2 Creep rate per unit stress as a function of grain size.

As mentioned earlier, N-H[2,3] and Coble[4] creep models assume grain boundaries to be perfect sources and sinks for vacancies which may not be always true. If such a limitation occurs, the vacancy concentration enhancement factor exp ($\sigma\Omega/kT$), assumed under ideal conditions, will not apply. Similarly, the vacancy concentration near grain boundaries parallel to the stress will not be the thermal equilibrium value as assumed under ideal conditions. Whichever of these processes is rate controlling, the rate of emission or absorption of vacancies at a grain boundary should be proportional to its area. Under these conditions, creep rate has been shown[20] to be inversely proportional to the grain size. However, no experimental evidence on metallic systems in support of this prediction exists in the literature.

Significant threshold stress (σ_0), stress below which no detectable creep occurs, has been observed experimentally[1,21,22] though N-H and Coble theories do not predict (equations 2 and 3). Further, threshold stress has been found to vary inversely with the grain size. This implies that refinement of grain size leads to decreased creep rates by reducing the effective stress ($\sigma - \sigma_0$). This behaviour is contradictory to the predictions of N-H and Coble creep theories which assume grain boundaries to be perfect sources and sinks for vacancies. Threshold stress thus offers a means to strengthen the material against diffusional creep. However, our present understanding of the origin of threshold stress is not complete. Though several models have been proposed[1,21,22] to explain the experimental results, no single model can satisfactorily explain all the features.

Recently Ruano and Sherby[23,24] have disputed the available evidence on Coble creep in single phase metallic polycrystals where significant threshold stress has been

observed. Based on a reanalysis of the published data, grain boundary sliding accommodated by slip has been suggested as the principal mechanism of plastic flow rather than Coble creep associated with a threshold stress as concluded by earlier investigators. The suggested mechanism[23,24] also predicts an inverse cube grain size dependence of creep rate in a manner similar to that of Coble[4] theory. However, Sritharan and Jones[25,26] have rejected the hypothesis put forward by Sherby[23,24].

Unlike N-H and Coble creep processes, H-D creep is characterised by grain size independent creep rate [19] (Fig. 3). Though this creep mechanism has been identi-

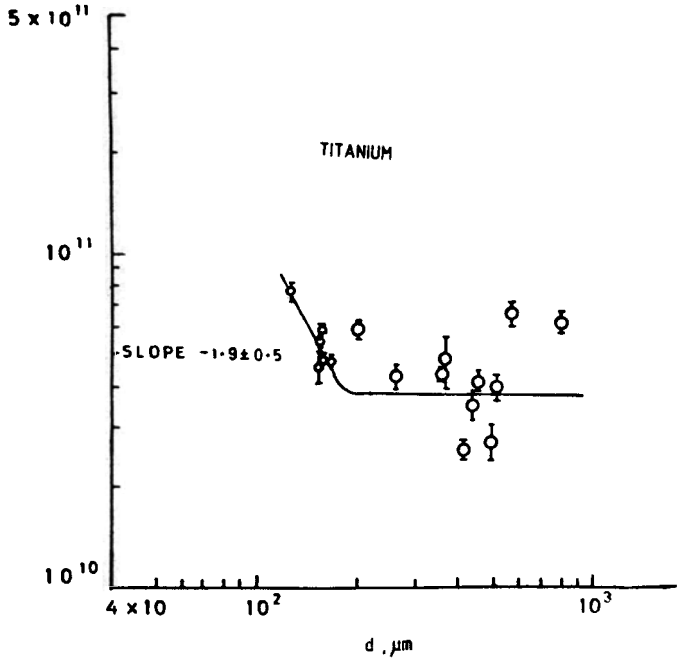

Figure 3. Temperature and lattice diffusivity compensated creep rate per unit stress as a function of grain size.

fied as early as 1957 by Harper and Dorn[5] during experiments on coarse grained aluminium, studies until recently were confined to temperatures close to the melting point[7] ($> 0.9\ T_m$). The authors' recent work on creep of alpha-titanium[19] and beta-cobalt[27] at low stresses has clearly demonstrated that H-D creep can dominate the flow at temperatures as low as $0.5\ T_m$, the temperature range of practical importance. Very recently Fiala et al[28] have reported similar observations during experiments on iron. The operation of H-D creep at coarse grain sizes thus sets a limit to the attainment of creep resistance through grain coarsening route for low stress applications. Inspite of its practical importance, studies pertaining to H-D creep todate are limited[7,8] and the underlying mechanism remains yet to be understood clearly.

2.2 *Climb Creep*

Though it has been reported[12,29] that the transient creep strain increases with decreasing grain size, attention was all the time focussed on the influence of grain size variation on the steady state creep rate. Accordingly the effect of grain size on steady state creep rate is considered in detail.

The constitutive equation for power-law creep is

$$\dot{\epsilon}_c = A_c\,(D_v\,Gb/kT)\,(\sigma/G)^n \tag{4}$$

where A_c is a dimensionless constant and n the stress exponent. It does not include the grain size term and so it is generally assumed that the climb creep is independent of grain size. However, in practice it is not so. Experimental observations concerning the effects of variations in grain size on the climb creep behaviour are not identical and often conflicting[2,9]. Several early investigators[10–12] have found that the steady state creep rate decreases with increasing grain size upto some optimum grain size ($\sim$ 100 μm) and then increases with a further increase in grain size. Sherby[30] and Feltham et al[31,32] have found that even for smaller grain sizes $\dot{\epsilon}_c$ does not always decrease with an increase in grain size but may be proportional to the square of the grain size. On this basis, the general equation for creep of pure metals and alloys, with some reservations, has been shown[30] to be given by

$$\dot{\epsilon}_c = S_c\,d^2\,D_v\,(\sigma/E)^5 \tag{5}$$

where S_c in a constant and E the Young's modulus. Alternatively, it has been reported by others[13–17] that $\dot{\epsilon}_c$ is essentially independent of grain size above about 100 μm while for smaller grain sizes $\dot{\epsilon}_c$ increases with decreasing grain size. Typical results exhibiting such behaviour obtained on Cr-Mn-N austenitic stainless steel are shown in Fig. 4.

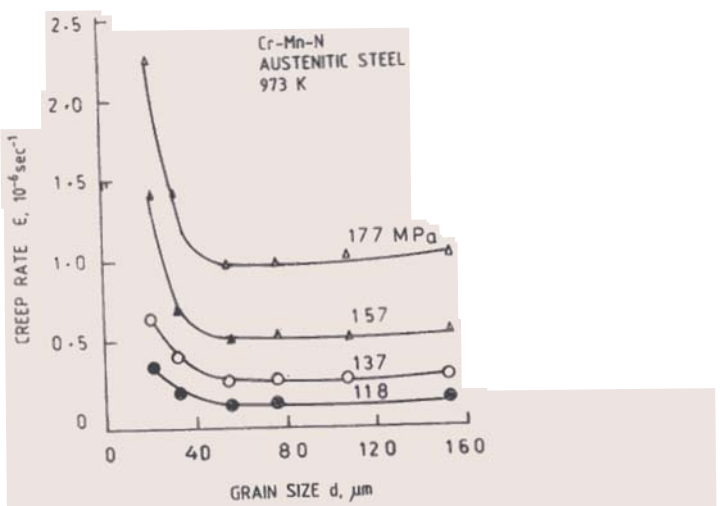

Figure 4. Grain size dependence of steady state climb creep rate.

Garofalo et al[12] have attributed the observed increase in creep rate with decreasing grain size below a certain optimum grain size to the action of grain boundaries as dislocation sources or sinks. Barrett et al[13] have questioned this proposition as it requires the mobile dislocation density to vary with the grain size. This is because the measurements by Barrett et al[33] on Fe-3 per cent Si alloy, where the grain size was varied from 50 to 300 μm, have shown the dislocation density within subgrains to be independent of grain size. Based on the measurments of the grain boundary sliding contribution to the total strain, Barrett et al[13] have attributed the increase in creep rate with decrease in grain size to the increased importance of grain boundary sliding as a deformation mechanism as the grain size is decreased. However, the underlying sliding mechanism has not been identified. As per the models proposed by Crossman and Asbhy[34] and Chen and Argon[35], grain boundary sliding can enhance the applied stress by a constant factor at strain rates below a critical value or equivalently below a critical grain size and thereby cause the creep rate to increase with decreasing grain size. Very recently, Tsang-Tse Fang and Linga Murty[36] have proposed to use the grain boundary sliding mechanism, based on the climb-controlled triple point fold formation, suggested by Gifkins[37] and showed an excellent agreement with the experimental results.

On the other hand, Evans [38] has attempted to explain the observed increase in creep rate with decreasing grain size by assuming the operation of Coble creep at fine grain sizes and climb creep at coarse grain sizes, which has been questioned by Burton[39] on the basis of stress dependence of the creep rate. If diffusional creep dominates at fine grain sizes, the value of the stress exponent should change accordingly from a value corresponding to the power-law creep to unity. Experimental results, however, indicate that the stress dependence of creep rate remains strong at all the grain sizes examined. It is pertinent to refer here to our recent findings related to the creep of alpha-titanium[40]. Increase in creep rate at fine grain sizes has been shown to be due to the contribution from diffusional creep processes (Fig. 5). The important point of distinction between the earlier observations and those on alpha-titanium, however, lies in the stress at which the creep tests have been performed, which is low and falls in the range below $10^{-4}G$. Figure 5 represents creep rate at a constant stress of 1.3 MN/m^2 ($\sigma/G = 5.3 \times 10^{-5}$) plotted against grain size. It is evident from the experimental creep mechanism map inserted in Fig. 5 that at $\sigma/G = 5.3 \times 10^{-5}$ (indicated by a broken line) climb creep dominates beyond a grain size of $\sim$ 100 μm, the region in which $\dot{\epsilon}$ is independent of grain size. On the other hand, at d less than about 100 μm, the region in which $\dot{\epsilon}$ increases with decreasing d, diffusional creep processes, namely N-H or Coble creep, dominate the flow.

Increase in creep rate with increasing grain size beyond certain optimum value has been attributed[13,30] to grain boundaries acting as barriers to the dislocation movement. On the other hand, the contradictory observation that the creep rate is independent of grain size has been explained[13] on the basis that subgrain boundaries, whose size is independent of grain size, act as barriers to the movement of dislocations

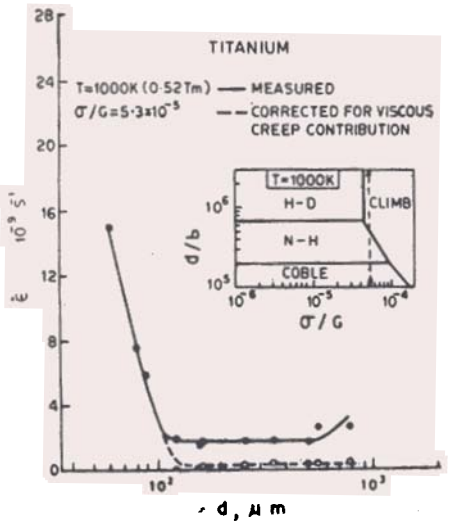

Figure 5. Steady state creep rate as a function of grain size at constant temperature (1000 K) and stress (1.3 MN/m^2) for titanium.

in addition to grain boundaries. At grain sizes coarser than 100 μm, the majority of boundary area belongs to the subgrains and hence the observed grain size independent behaviour of the creep rate.

It is evident from the foregoing that power-law creep does depend on grain size. Very recently Takahashi and Yamane[41], based on the experimental results on 25 Wt. per cent Cr-20 Wt. per cent Ni austenitic steel, have suggested that the power-law creep could be subdivided into three characteristics of P_1, P_2 and P_3. Constitutive equations for P_1, P_2 and P_3 are given by

$$\dot{\epsilon}_{p_1} = A_1 (D_v Gb/kT) (\sigma/G)^4 (b/d) \tag{6}$$

$$\dot{\epsilon}_{p_2} = A_2 (D_v Gb/kT) (\sigma/G)^5 (b/d)^2 \tag{7}$$

$$\dot{\epsilon}_{p_3} = A_3 (D_v Gb/kT) ((\sigma-\sigma_c)/G)^4 \tag{8}$$

where A is a constant and σ_c is the strain arrest stresss. P_1 is observed for the fine grained specimens ($d < 60\ \mu m$) and P_2 and P_3 for $d > 100\ \mu m$. P_3 is essentially independent of grain size and is considered to be strongly associated with the formation of dislocation substructure. The range of grain size from 60 to 100 μm can be regarded as the transition region from P_1 to P_3. It may be added here that P_1, P_2 and P_3 are not independent but only alternative mechanisms to each other.

3. Influence of Grain Shape on Creep Behaviour

It has been shown in Sections 2.1 and 2.2 that grain size has significant effect on diffusional as well as climb creep. Here we consider the effect of grain shape.

Recently, Nix[42] has analysed N-H and Coble creep processes for the case of three dimensional grains of length L, width W and thickness H. The general expressions thus derived for N-H and Coble creep processes are

$$\dot{\epsilon}_{NH} = \frac{12D_v\Omega\sigma}{S^2kT}(R_1R_2)^{-2/3}\left(\frac{R_1^2 + R_2^2}{1 + R_1^2 + R_2^2}\right) \qquad (9)$$

$$\dot{\epsilon}_{CO} = \frac{12wD_B\Omega\sigma}{S^3kT} \frac{\left(\frac{1}{R_1} + \frac{1}{R_2}\right)\left(\frac{R_2}{R_1} + \frac{R_1}{R_2}\right)}{\left(\frac{1}{R_1} + R_1 + \frac{1}{R_2} + R_2 + \frac{R_2}{R_1} + \frac{R_1}{R_2}\right)} \qquad (10)$$

where S is the grain size defined as $S = (LWH)^{1/3}$ and R_1 and R_2, defined as $R_1 = L/H$ and $R_2 = L/W$, are grain aspect ratios. The applied tensile stress σ is assumed to act on the solid parallel to its length. Equations (9) and (10) allow one to predict the diffusional creep rate for grains of any length, width and thickness.

Nix[42] has also examined grains with square cross-section, the results of which are reproduced in Fig. 6. For grains with square cross-section, $R_1 = R_2 = R$. It is interesting to examine the two limiting cases of $R \to 0$ and $R \to \infty$. As $R \to \infty$, $\dot{\epsilon}_{NH} \sim R^{-4/3}$ i.e the creep rate decreases at large grain aspect ratios. $\dot{\epsilon}_{NH} \sim R^{2/3}$ as $R \to 0$ which also indicates that the creep rate should be very low for small grain aspect ratios. Similarly when R is very large, Coble creep rate is low i.e. $\dot{\epsilon}_{Co} \sim R^{-2}$ as $R \to \infty$. However, for small grain aspect ratios ($R \to 0$), the Coble creep rate approaches a constant value ($\dot{\epsilon}_{Co} = 24wD_B\Omega/S^3kT$). This is somewhat surprising in that unlike

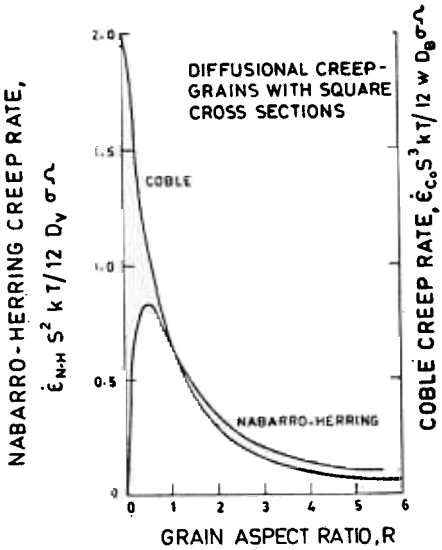

Figure 6. Effect of grain aspect ratio on Nabarro-Herring and Coble creep for grains with square cross-section.

all of the other cases considered here, the creep rate does not fall to zero at very small grain aspect ratios. All these limiting cases are shown in Fig. 6. It is evident from Fig. 6 that diffusional creep rate (Coble as well as N-H) is low at higher grain aspect ratios. This is because the diffusional distances required for diffusional creep are very large at higher grain aspect ratios. This aspect is taken advantage of while developing materials for high temperature applications e.g. directional solidification of Nickel-based superalloys for turbine blades.

Not only the grain size but also the grain shape influences climb creep behaviour. The effect of grain shape in terms of grain aspect ratio, defined as L/W, on creep behaviour of dispersion strengthened nickel alloys[43] is shown in Fig. 7. Physical interpretation[44] of the behaviour shown in Fig. 7 is that grain boundary sliding (GBS) plays an important role in climb creep. When most of the grain boundaries are parallel to the stress axis, i.e., when there is a highly elongated microstructure, there is, on the average, a low resolved shear stress on the boundaries. Since shear stress is the driving force for GBS, this minimises the overall amount of sliding. Maximum sliding would occur when the aspect ratio is unity (equiaxed grains). However, our present state of understanding of the effect of grain aspect ratio on climb creep is far from satisfactory. For example, based on the existing climb creep models, one would expect subgrain size (as will be shown in Section 6) and not the grain dimensions or grain aspect ratio which should affect the creep rate.

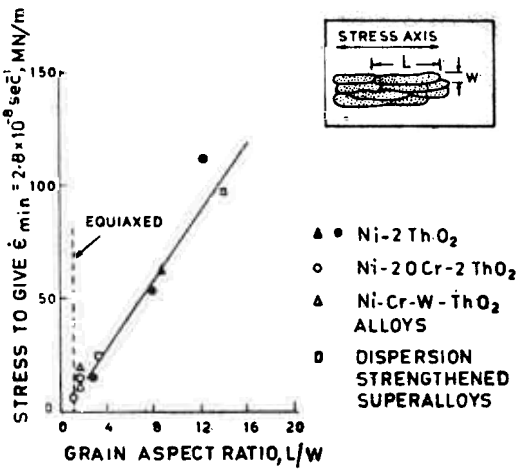

Figure 7. The effect of grain aspect ratio on creep strength of dispersion strengthened nickel alloys at 1373K.

4. Grain Size Effects in Intermetallic Alloys

Intermetallic alloys based on iron, nickel or titanium are considered to be futuristic structural materials for elevated temperature applications. We would, therefore, expect the creep properties of these alloys to be well documented and analysed. It

is, however, surprising to note[45] that the present understanding of the processes controlling creep deformation in these solids is woefully inadequate. There have been very few studies in which creep properties have been reported[45–47]. Currently, attention[48,19] is diverted mostly to those aspects related to room temperature ductility and fabricability of these alloys. In these studies[48,49] greatly intriguing results have been obtained. Intrinsic as well as extrinsic grain boundary embrittlement has been observed, so also enhancement of ductility through segregation of dopants such as boron to grain boundaries. Furthermore, the action of such dopants has been found[49] to be dependant on composition of Ni_3 *Al*. These grain boundary aspects should have profound influence on high temperature creep of Ni_3Al. However, the limited studies on creep indicate that the creep behaviour of single Crystal[47] Ni_3 (*Al, Ti*) as well as polycrystalline[46] Ni_3 *Al* obeys power-law relationship with stress exponent of nearly 3. There is thus ample scope for original research in the case of Ni_3Al and other such promising high temperature intermetallic alloys.

Influence of Subgrain Size on Creep Behaviour

Subgrain formation and development dominate the microstructural changes that occur during creep (climb) flow of coarse grained polycrystalline metals[50]. Subgrains form during the primary stage and fully develop at the transition between the primary and the secondary stage. Jerkins and Meller[51] were among the first to observe subgrains formed during creep of iron in 1935. Material with high SFE, e.g. Al and α-Feare subject to marked subgrain formation which start early in the primary stage[52,53]. On the other hand, materials with a low SFE show less tendency to develop subgrains. For instance, in α-brass subgrains were not observed in the primary and secondary stage[54]. However, the specific influence of subgrains on the creep rate has been ignored for nearly three decades. Subgrain formation was often considered to be simply a product of the creep process and was regarded as neither beneficial nor detrimental to the creep resistance of the material. Substructure can also be pre-induced. Cold work followed by recovery annealing or warm working leads to the formation of subgrain structure. It is only during the last two decades that the importance of pre-induced substructure has been realised.

Let us first consider the effect of substructure on transient creep. Though the classical theories of N-H[2,3] and Coble[4] do not predict occurrence of a transient stage during which creep rate decreases with time, prior to the attainment of steady state, transient stage has almost always been observed under diffusional creep conditions[1,21]. Transient creep stage has been observed under H-D creep conditions too[22]. In a recent that analysis, the authors[55] have shown dislocation activity as the cause of transient creep under viscous creep conditions. Creep tests have been performed at a constant temperature of 1046-1049K on alpha-titanium of nearly the same grain size (d = 149 and 153 μm) but obtained via different thermomechanical treatments, the results of which study, reported earlier[55], are reproduced in Fig. 8. It is seen from

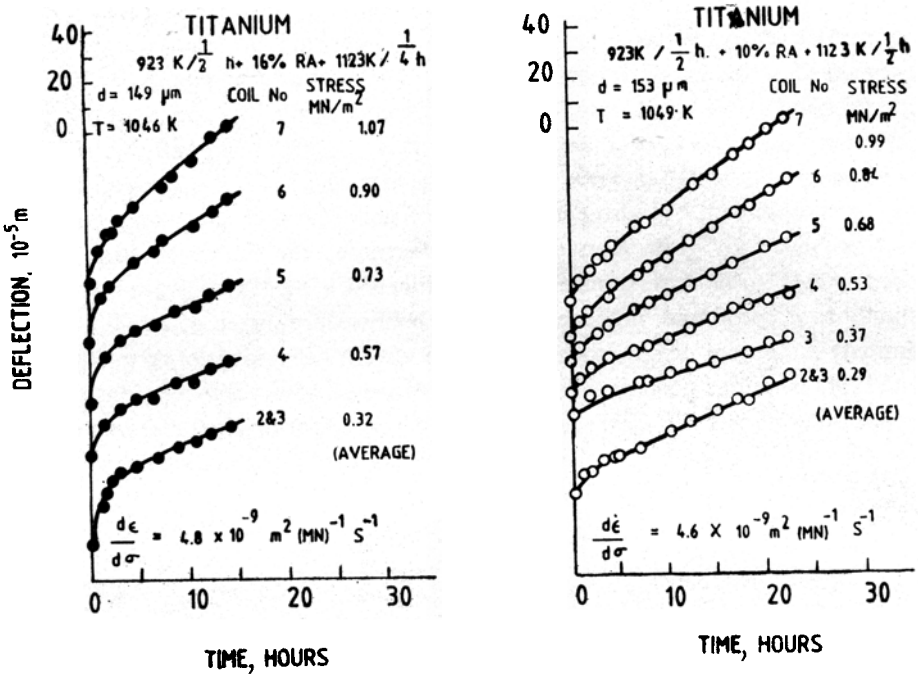

Figure 8. Deflection-time curves for alpha-titanium. RA refers to reduction in cross sectional area by cold drawing.

Fig.8 that the steady state creep rate, strongly sensitive to grain size, is nearly independent of prior thermomechanical history. On the other hand, transient strains are markedly different. This effect has been attributed[55] to difference in contributions to transient creep from a mechanism based on dislocation activity since the prior dislocation substructure, dependent on the extent of cold work and the degree of recovery, was different. The occurrence of a transient stage under climb creep conditions, on the other hand, is a phenomenon which is well known and extensively studied. It is well understood in terms of substructural changes, i.e., formation and attainment of a stable subgrain structure[56]. It is therefore evident from the foregoing that substructure plays a role during transient stage under viscous as well as power-law creep conditions.

Having discussed the effect of substructure on transient creep, we now consider how the substructure influences steady state creep behaviour. It has been shown earlier, vide Fig.8, that the substructure does not influence the steady state behaviour under diffusional creep conditions. However, it may be mentioned here that although the lattice substructure (dislocation density or distribution) does not influence steady state diffusional creep rates, creep rates do depend on the density and mobility of grain boundary dislocations which act as sources and sinks for matter[57]. There are reports that subgrains form during H-D creep[7]. Since our present understanding of

H-D creep is not clear[7], the discussion on the influence of substructure will not be extended to this particular creep mechanism. The influence of substructure on climb creep, the subject of prime interest to many investigators, will now be discussed in greater detail.

The importance of substructure in climb creep is still the subject of considerable controversy in the literature. The majority of theories proposed to describe the creep process do not include subgrain size in the rate equation. An attempt has been made by Robinson and Sherby[38] earlier to incorporate the subgrain size in the empirical rate equation after analysing all of the creep data on polycrystalline tungsten. Recently, Sherby et al[59] reanalysed the data available in the literature for several materials and proposed an equation of the form

$$\dot{\epsilon}_i = B\ (D_{eff}/b^2)\ (\lambda/b)^p\ (\sigma/E)^N \tag{11}$$

where $\dot{\epsilon}_i$ is the instantaneous strain rate, B the structure dependent constant ($\sim 1.5 \times 10^9$ for high SFE materials), and D_{eff} the effective diffusivity. For aluminium it was found[50] that $p = 3$ and $N = 8$. Equation (11) predicts a thousand-fold decrease in creep rate by a ten-fold decrease in subgrain size for aluminium.

While developing Eqn. (11), it was implicitly assumed[59] that transient period following a stress reduction, performed in the steady state region, must be accompanied by subgrain growth. Subgrain growth continues until a value consistent with that observed for steady state deformation at the reduced stress is obtained. This assumption has been criticized by several investigators[60,61]. Experiments conducted on copper by Perker and Wilshire[60] failed to show subgrain growth following a stress reduction. Mitra and McLean[62] and Pontikis and Poirier[61] also reported similar behaviour. However, recent data reported by Ferreira and Stang[63] and Soliman et al[64] clearly demonstrate that subgrain growth occurs in high purity aluminium in support of the assumptions of Sherby et al[59]. Similar observations of subgrain growth following stress reduction have been reported by Blum et al[65] for *Al*-11 per cent *Zn* and Eggeler and Blum[66] for *NaCl*. Very recently Ferreira and Stang[67] have reported experimental data on high purity aluminium in support of the concepts developed by Sherby et al vide Eqn. (11).

Control and stabilisation of subgrain structure leads to significant improvement in creep strength of a given material. If a refined subgrain structure is stabilised by inert hard particles, one can expect a material with very high creep resistance. Evidence supporting this hypothesis has been obtained from tests on thoria dispersed (TD) nickel[68]. TD nickel in the fully annealed state does not exhibit high creep resistance. When it is cold worked, however, it becomes very strong at high temperature. One possible explanation is that cold work followed by annealing creates the subgrain network that the thoria particles stabilise.

The major difficulty associated with subgrain refinement is that subgrains usually are not stable and grow especially under the influence of stress. Several investigations

have revealed that the subgrain size is a unique function of the flow stress experienced by the sample during creep. Typical results on *Al*-1*Mn*-1 *Mg* reported recently by Blum et al[69] are shown in Fig 9. The data presented in Fig. 9 can be described by a straight line corresponding to

$$\lambda = AbG/\sigma \tag{12}$$

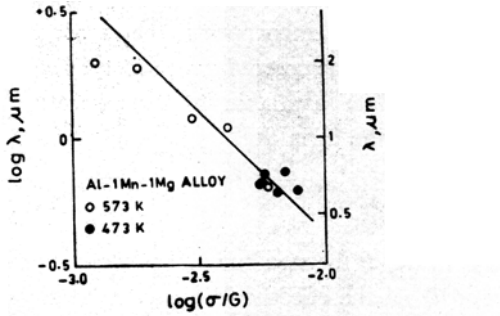

Figure 9. Stress dependence of subgrain size.

where A is a constant equal to 14. Equation (12) further suggests that λ is independent of temperature which has been verified experimentally by Blum et al[69].

Blum et al[69] have shown that the substructure hardening effect depends on the difference between the initial subgrain size λ_0 and the final (steady state subgrain size) subgrain size λ_S. Under conditions where $\lambda_s \approx \lambda_o$, substructural hardening is prominent. On the other hand, under conditions of low stress and high temperature where $\lambda_s >> \lambda_o$, there is virtually no hardening effect. This effect has been demonstrated by Blum et al[69] on the basis of the results obtained on *Al*-1*Mn*-1*Mg* and the same is reproduced in Fig. 10. According to Eqn. (11) given value, by substituting $p = 3$, the ratio of the initial creep rate ($\dot{\epsilon}_o$) to the steady state creep rate ($\dot{\epsilon}_s$) is given by

$$\dot{\epsilon}_s/\dot{\epsilon}_0 = (\lambda_S/\lambda_0)^3 \tag{13}$$

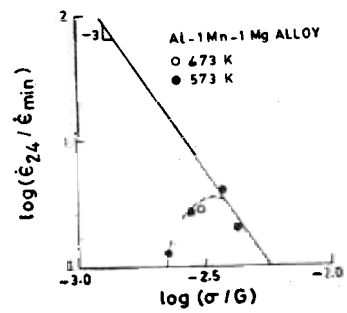

Figure 10. Double-logarithmic plot of $\dot{\epsilon}_{24}/\dot{\epsilon}_{m\,n}$ as a function of normalised stress. Solid line is drawn with a slope of −3 as per Eqn. 14.

Combining this with Eqn. (12) leads to

$$\dot{\epsilon}_S/\dot{\epsilon}_0 = (14b/\lambda_0)^3 \ (\sigma/G)^{-3} \tag{14}$$

To check the usefulness of Eqn. 14, $\dot{\epsilon}_S$ and $\dot{\epsilon}_0$ have been approximated to $\dot{\epsilon}_{24}$, the strain rate at $\epsilon = 0.24$, and $\dot{\epsilon}_{min}$ respectively and the ratio of $\dot{\epsilon}_{24}$ and $\dot{\epsilon}_{min}$ is plotted against σ/G on log-log basis (Fig. 10). For the sake of comparison, predicted line as per Eqn. (14) is also included. The data presented in Fig. 10 reveal that there is indead a limited range of stress in which Eqn. (14) is a useful approximation. Under conditions where $\lambda_S > \lambda_0$, Eqn. (14) fails completely in predicting $\dot{\epsilon}_{24}/\dot{\epsilon}_{min}$. Instead of increasing steeply with decreasing stress, $\dot{\epsilon}_{24}/\dot{\epsilon}_{min}$ decreases. This emphasises the importance of the difference between λ_0 and λ_S rather than the magnitude of λ_0 for substructure strengthening.

6. Summary

Influence of grain size, grain shape as well as subgrain size on viscous (Nabarro-Herring, Coble and Harper-Dorn Creep) and power-law creep processes has been considered. The present status of our understanding of these aspects can be summarised as follows:

1. Grain size strongly influences diffusional creep processes. Creep rate varies inversely as square and cube of grain size under N-H and Coble creep conditions respectively. On the other hand, steady state creep rate under H-D creep conditions is independent of grain size. Though it is generally assumed that power-law creep is independent of grain size, it is not so in practice. Experimental observations related to the effect of grain size on climb creep are varied and often conflicting.

2. Grain shape influences diffusional as well as climb creep processes. Long elongated grain structure, oriented parallel to the stress axis, leads to strengthening under diffusional as well as power-law creep conditions.

3. Substructure plays a role during the transient stage of viscous as well as power-law creep processes.

4. Refinement and stabilisation of subgrain size leads to significant strengthening under climb creep conditions. Subgrain size, unless stabilised, varies inversely with stress. Under such conditions, it is the difference between the starting (λ_0) and the steady state subgrain size (λ_S) that is important but not the prior subgrain size. Subgrain strengthening is significant if $\lambda_o \sim \lambda_S$. On the other hand, if $\lambda_S >> \lambda_o$ there is virtually no hardening effect.

5. The role of grain boundaries in virgin and doped intermetallic alloys, a new class of solids currently attracting attention for technological applications, is intriguing, which situation offers fascinating opportunities for new and pathfinding research.

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