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Newtonian flow process in polycrystalline silicon carbides: diffusional creep or Harper–Dorn creep?

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Some previous studies on hot-pressed and sintered SiC polycrystalline materials have been re-examined. Mechanical data and microstructures strongly suggest that the Newtonian creep behaviour observed in these SiC materials was induced by a dislocation process operating in Harper–Dorn creep, rather than by diffusional creep as concluded before. The supporting evidence for this suggestion includes extensive development of dislocation substructures, no dependence of creep rate upon grain size, and the measured creep rates being far faster than those predicted by the model of diffusion creep, but consistent with those estimated by the model of Harper–Dorn creep.

1. Introduction

A thorough knowledge of creep characteristics of materials is required for their use in engineering applications. Experimental studies of creep processes on crystalline materials have shown that, at low and intermediate stresses, the creep rate, $\dot{\varepsilon}$, is generally represented by

$$\dot{\varepsilon} = \frac{AGbD}{kT}\left(\frac{b}{d}\right)^p\left(\frac{\sigma}{G}\right)^n \qquad (1)$$

where D is the diffusion coefficient, G is the shear modulus, $\boldsymbol{b}$ is the Burgers vector, k is the Boltzmann's constant, T is the absolute temperature, d is the grain size, σ is the applied stress, and p, n and A are dimensionless constants.

At intermediate modulus-compensated stresses, power-law dislocation creep is the dominant mechanism giving rise to $n = 3$–5, $p = 0$ and $D = D_L$, where D_L is the lattice diffusivity. However, at low modulus-compensated stresses, there is often a transition to Newtonian viscous flow with $n = 1$. Three distinct mechanisms could result in this behaviour. Two of them involve the stress-directed flow of vacancies from free surfaces and grain boundaries under a tensile stress to those under relative compression via the crystal lattice (Nabarro–Herring, N–H, creep [1, 2]) or via the grain boundary (Coble creep [3]). For N–H creep, $D = D_L$, $p = 2$, but for Coble creep, $D = D_b$ (where D_b is the grain-boundary diffusivity) and $p = 3$. The third mechanism concerns dislocation generation, multiplication and motion within individual grains (Harper–Dorn, H–D, creep [4]). For this creep, generally, $D = D_L$, and $p = 0$.

While theory and experimental data on diffusional creep have been well documented for a few decades, knowledge about H–D creep was expanded only recently. Since the original report [4], H–D creep has been observed not only in aluminium and its alloys [5] and a few other metals (lead [6], tin [6], β-cobalt [7], α-titanium [8], α-iron [9], α-zirconium [10], but also in ceramics (CaO [11, 12], MgO [13], UO_2 [14], $Mn_{0.5}Zn_{0.5}Fe_2O_4$ [15], NaCl [16]), in perovskites ($KZnF_3$ [17], $KTaO_3$ [18]) and in silicates (quartz [19], olivine [20], feldspar [21]). Originally, H–D creep was reported for aluminium, lead and tin at $T/T_m > 0.9$ (where T_m is the melting temperature). The results on β-cobalt, α-iron, α-zirconium, α-titanium, UO_2, CaO and SiO_2 show that H–D creep also operates at $T/T_m \approx 0.5$.

In contrast with that in power-law dislocation creep, the dislocation density in H–D creep is independent of the applied stress [5, 19], but is determined by the magnitude of the Peierls stress of the crystal [22]. Based on experimental data, it was shown that the transition from power-law creep to H–D creep may occur at the stress level of the Peierls stress [23]. It was demonstrated that the internal stress model of H–D creep [24] may be an appropriate model [25]. Instead, it was illustrated that the rate-controlling flow mechanism may be dislocation climb under saturated conditions not only in aluminium [26] but also in most other materials for which H–D creep has been observed [27].

Because silicon carbide (SiC) possesses excellent resistance to thermal shock and corrosion, high fracture strength and superior creep resistance, it is a perennial candidate material in high-stress and high-temperature structural applications. In many of these applications, creep processes play an important role in the physical and even chemical properties of this material. Thus, creep of SiC has been a subject of many investigations.

Previous studies have shown that polycrystalline SiC materials exhibit a Newtonian creep behaviour

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over a wide range of experimental conditions, and suggested that the cause was the operation of diffusional creep. It is the objective of this paper to illustrate the importance of H–D creep in SiC which has a very high Peierls stress and to provide an alternative and more reasonable explanation for the Newtonian behaviour observed.

2. Previous experimental observations

Four different SiC polycrystalline materials were investigated in bending or compression: (1) reaction bonded (RB), (2) chemically vapour deposited (CVD), (3) sintered (SIN) and (4) hot pressed (HP). The grain size of the materials varied from 2–65 μm, the temperature from 1273–2473 K (0.41–0.8 T_m), and the stress from 5.7–1700 MPa. Table I lists experimental conditions and results and the suggested controlling deformation mechanism(s).

Creep curves in almost all experiments show a transient stage prior to a steady state. In tests at the lowest temperatures ever applied (0.41–0.48 T_m), a steady state was actually not clearly reached and the maximum creep strain achieved without failure was $\sim 0.17\%$ [38].

Creep parameters were determined for steady-state creep. For the RB SiC [28], the stress exponent, n, was 5.7 and the activation energy, Q, was 711 kJ mol^{-1}. It was suggested that the primary deformation mechanism was dislocation glide plus climb controlled by climb. For the CVD SiC [29], Q was 175 kJ mol^{-1} and n was 2.3 at $T < 1923$ K, but was 3.7 at $T \geq 1923$ K. It was concluded that the mechanism was dislocation glide controlled by the Peierls stress at $T < 1923$ K, but controlled by climb at $T > 1923$ K.

The HP SiC exhibited a Newtonian creep behaviour at temperatures between 0.41 and 0.80 T_m and stresses even up to 1700 MPa. Q was between 170 and 305 kJ mol^{-1}. Experiments [35] showed that this Newtonian creep changed to non-Newtonian creep ($n \gg 1$) as the stress was increased to above a critical value, σ_c. The value of σ_c decreased with an increase in temperature. The observation of a low value of Q led to a conclusion that Newtonian creep of this SiC was controlled by Coble creep [35–37, 40].

Creep of the SIN SiC is characterized by a value of n between 1 and 2 and a value of Q between 735 and 914 kJ mol^{-1} above a critical temperature, T_c ($= 1920$ K), but by a value of Q between 338 and 541 kJ mol^{-1} below T_c.

In order to determine the grain-size dependence of the creep rate, sintered α-SiC was refired at high temperatures, resulting in materials with three different average grain sizes, 3.5, 4.9 and 7.5 μm, respectively [31]. Grains in all three materials were equiaxed as well as rod-shaped. The aspect ratios of the rod-shaped grains in the three materials were ~ 5, ~ 9 and ~ 15, respectively. The refired materials had a large number of rod-shaped grains with high aspect ratios. The grain-size exponent, p, was found to be around 4 above T_c, but about 2.5 below T_c.

Detailed microstructural studies were conducted to characterize the intergranular and intragranular features of the SIN SiC before and after deformation. The following brief summary is mainly based on the study of Lane *et al.* [30].

Electron microscopic observations [30, 32] showed that there was no amorphous or crystalline second phase in grain boundaries of the SIN SiC before and after creep testing as often detected in sintered silicon nitrides. Intentional examination of numerous grain boundaries in any crept sample revealed no evidence for grain-boundary corrugation or pore formation or the occurrence of triple-point voids [30].

Most grains in the material prior to testing were free of line defects with few containing stacking faults. Within some grains, particles were observed with a few associated dislocations.

The most obvious change in the crept samples was the large increase in the number of stacking faults which were observed in almost every grain [30]. Close examination showed that the stacking faults were composed of high densities of slip bands.

Additional features observed throughout the crept samples were the substantial forests of tangled dislocations and large particles located within some grains [30]. More than half of the grains contained tangled dislocations. Some of them were essentially filled while others had only a small number of clusters. The particles were a mixture of graphite and carbon, and were accompanied by a high density of single dislocations which were apparently generated at the interface.

The less-common features were climbing dislocations and B_4C precipitates. Dislocation climb was observed most frequently in samples deformed to high strains, but only occasionally in those deformed to the initial stage of transient creep. The precipitates, 13–25 nm diameter, interacted with gliding dislocations. Precipitation and the interaction occurred in transient creep as well as in steady-state creep. Owing to its occurrence also in annealed-only samples [31], the precipitation might not be related to deformation.

Extensive observations showed that, although the activation energy is different for creep at $T > T_c$ (1920 K) from that at $T < T_c$, there was no significant microstructural difference among the samples crept above and below this transition temperature [30].

Based on creep data and microstructures, it was concluded that the deformation of the SIN SiC was controlled by N–H creep at $T > T_c$, but by Coble creep at $T < T_c$ [30–32].

3. Evaluation of Newtonian flow process in polycrystalline SiC

Previous studies show that the creep kinetics in polycrystalline SiC depend on the applied stress and temperature and also, more strongly, on the process used to develop the material and the resulting microstructure.

The RB and the CVD SiC exhibited a power-law creep behaviour. As concluded previously [28, 29], these two SiC materials were probably controlled by

TABLE I Experimental conditions and results on polycrystalline materials of SiC

Reference	Impurities (p.p.m. or %)	Theoretical density (%)	Developing process	Grain size (μm)	T (K)	T/T_m	Stress (MPa)	Testing technique	Stress exponent	Activation energy (kJ mol^{-1})	Proposed mechanism
Carter *et al.* [28]	< 1%		RB	10–100	1848–1923	0.60–0.62	110–220	Compression	5.7	711	Climb-controlled dislocation glide
Carter *et al.* [29]	~ 500	98.3	CVD		1573–2023	0.51–0.65	69–220	Compression	2.3–3.7	175	Dislocation glide controlled by the Peirels stress or by climb
Lane *et al.* [30]	< 200	98.4	SIN	3.7	1670–2073	0.54–0.69	138–414	Compression	1.44–1.71	338–434 (< 1920 K) 802–914 (> 1920 K)	Coble creep N–H creep
Nixon and Davis [31]			SIN	3.5–7.5	1770–2020	0.57–0.65	35–373	Compression	1.4–2.5 (< 1920 K) 1.0–1.4 (> 1920 K)	387–541 (< 1920 K) 838–877 (> 1920 K)	Diffusional creep plus dislocation glide
Tanaka and Inomata [32]	0.6	95.3	SIN	2–5	1893–2033	0.61–0.66	63–267	Compression	1.3–2	735–790	Diffusional creep
Grathwohl *et al.* [33]	0.25%–0.34%				1743–1933	0.56–0.62	100–190		1	796	N–H creep
Prochazka and Smith [34]	Pure		HP	2–3	1723–1823	0.56–0.59	206.9	Bending		732.6	
Djemel *et al.* [35]	< 2%	94–96	SIN	3.5–5	1573–1773	0.51–0.57	500–1700	Compression	1	170–300	Coble creep
Farnsworth and Coble [36]	~ 2.5%	> 99	HP	26	2173–2473	0.7–0.8	20–200	Bending	~ 1	305	Coble creep
Francis and Coble [37]	2.5%	95–99	HP	9.8–27.7	2248–2393	073–0.77	5.7–143	Bending	~ 1	305	Coble creep
Marshall and Jones [38]	< 1000	92	SIN	10	1273–1473	0.41–0.48	207–496	Bending	1–2	230	
Seltzer [39]			HP		1523–1748	0.49–0.56	55–344	Compression	0.9		
Krishnan and Notis [40]	Various	97		65	1573–1673	0.51–0.54	34.5–86.2	Bending	0.9	145	Coble creep

intragranular dislocation processes operating in conventional power-law creep.

By contrast, the HP and the SIN SiC showed a Newtonian creep behaviour over wide ranges of experimental conditions. This behaviour was unlikely to be due to grain-boundary sliding creep involving rotation of individual grains, as already pointed out [30–32]. The reason is that this creep process is accommodated by (a) viscous flow of an amorphous grain-boundary phase, or (b) formation of grain-boundary cavities, or (c) the occurrence of triple-point folds [41]. However, no amorphous grain-boundary phase, nor triple-point voids nor folds or grain-boundary porosity were observed. Instead, it was suggested that the Newtonian behaviour was induced by diffusional creep [30–32].

Common in all previous studies on SiC, H–D creep was never mentioned and the possibility of this creep becoming rate-controlling under certain conditions was never considered. However, recent analyses of data on metals, ceramics and silicates consistently showed that the Peierls stress of a crystal is an important parameter in determining H–D creep behaviour. Namely, a high Peierls stress leads to a high dislocation density in this creep [22], a high stress marking the transition from this creep to power-law creep [23], and a small transition grain size from this creep to diffusional creep [42]. These results suggest that SiC could be controlled by H–D creep under wide experimental conditions because of its high Peierls stress.

In the following sections, it will be evaluated which creep was more likely to be responsible for the Newtonian behaviour observed in SiC; diffusional creep or H–D creep? The evaluation will proceed according to experimental observations on (1) creep curves, (2) microstructures, (3) creep activation energies, (4) grain-size effects and (5) creep rates.

3.1. Creep curves

Because diffusional creep is a viscous flow, a steady state is anticipated immediately after the application of a stress in a creep experiment, or after the imposition of a strain rate in a constant strain-rate test. The flow continues to high strains without fracturing. However, SiC specimens showed a clear transient stage and a decohesion stage leading to fracturing preceding and following the steady state stage, respectively.

3.2. Microstructures

The suggestion that the HP SiC was controlled by diffusional creep is mainly based on an early belief that dislocation movement was difficult in SiC due to its high Peierls stress.

Farnsworth and Coble [36] stressed single crystals of α-SiC in bending to 276 MPa at 2400 K for 10 h. TEM confirmed the generation but reluctant movement of a few dislocations on non-basal $\{4\bar{4}01\}$ and $\{1\bar{1}01\}$ planes and a lack of dislocation effects on $\{0001\}$ planes. Similarly, no plastic flow was observed in a study on α-SiC single crystals subjected to four-point bending up to 2473 K [43].

In contrast, in other studies, also using bending tests, considerable plastic deformation occurred in single crystals of SiC at 2273–2373 K [44–46] and a high density of dislocations with slip on the basal plane having a vector of $\langle 11\bar{2}0\rangle$ was developed [47]. Very recently, single crystals of SiC were deformed in compression at orientations parallel to and at 45° to [0001] [48]. Deformation of the 45 °C orientation was dominated by basal slip. The steady state creep rates of above $10^{-7}\,s^{-1}$ were measured at temperatures even as low as 800 °C. The deformation studies on RB, CVD and SIN polycrystalline SiC also revealed very active dislocation motion during experiments [28–30, 32].

Thus, it now appears possible that dislocation process can be activated and SiC can deform by dislocation creep under experimental conditions.

Actually, TEM studies on the SIN SiC showing a Newtonian behaviour revealed extensive development of dislocation substructures during deformation [30, 32]. These substructures include widespread stacking faults and slip bands, ubiquitous forests of tangled dislocations, dislocation generation at the interface of particles, obvious dislocation climb in highly strained specimens and interaction of precipitates with gliding dislocations. Most of these dislocation substructures were not observed in the materials before experiments. Further, there was also evidence that the substructures were more widely and better developed in steady-state creep than in transient creep [30].

Because strain is mainly induced by diffusion of atoms from one place to another in diffusional creep, but by dislocation generation and motion in H–D creep, the extensive development and evolution of dislocation substructures is indicative of a control of H–D creep.

The dislocation density is dependent upon, and thus is determined by the applied stress in power-law creep, whereas it is independent of the applied stress, but is proportional to the magnitude of the Peierls stress of the crystal in H–D creep [22]. The Peierls stress of SiC is larger than those for quartz and other crystalline materials for which H–D creep has been observed. Inspection of the transmission electron micrographs taken from SiC specimens which exhibited a Newtonian creep behaviour (Figs 8c, 9a and 11 in [30]) shows that the dislocation density is higher than that observed in quartz [19].

One may consider that the dislocation substructures were formed to accommodate grain-boundary sliding, as modelled [49]. Lane *et al.* [30], however, argued that this was not the case in the SiC material investigated.

3.3. Creep activation energies

Further evidence cited for a control of Coble creep in the HP SiC was the experimental observation that the activation energy for creep was significantly smaller than that for lattice self-diffusion.

The observed activation energy ($\sim 300\,\mathrm{kJ\,mol^{-1}}$) for Newtonian creep in the HP SiC is certainly much smaller than that for lattice self-diffusion and also smaller than that for grain-boundary diffusion of carbon. Diffusion studies on both single [50, 51] and polycrystals [52, 53] of SiC suggested that the activation energy for lattice diffusion of both silicon and carbon ranges from 715–912 kJ mol^{-1}. The activation energy for grain-boundary diffusion of carbon was measured to be 563 kJ mol^{-1} [52].

However, it cannot be definitely concluded that Newtonian creep in the HP SiC which had an activation energy much smaller than that for lattice diffusion was controlled by Coble creep. The reasons are: (1) the activation energy for power-law dislocation creep in single crystals [48, 54] and polycrystals [29] of SiC were also ~300 kJ mol^{-1}, and (2) the activation energy for creep of the HP SiC was around 300 kJ mol^{-1}, even at very high temperatures ($T = 0.7$–$0.8\ T_m$) [36, 37] where creep is usually controlled by lattice diffusion.

Thus, it appears that an unspecified dislocation process could also have a creep activation energy much smaller than that for lattice diffusion.

3.4. Grain-size effects

A grain-size dependence of creep rate was observed in the study on the SIN SiC materials with different average grain sizes [31]. The grain size exponent, p, was found to be around 4 above T_c, but to be about 2.5 below T_c. This observation may suggest a control of diffusional creep. However, there are a few reasons to oppose this suggestion.

1. As discussed [31], measurements of the grain-size dependence were not reliable, because the theory relating grain size to steady state creep rate originally assumed equiaxed grains. The SiC samples used to study the effect of grain size did not satisfy this criterion. Instead, these materials contained grains of different shapes and sizes. During diffusional creep, different grain shapes give rise to different lengths of diffusion paths. Materials have an identical average grain size but with different grain-size distributions may deform at different rates at the same temperature and stress [55].

2. The effect of grain-size distribution on creep of polycrystalline materials was numerically modelled [56]. It was shown that in the case where the smaller grains deform by diffusional creep, but the larger grains deform by power-law dislocation creep, a small number of small grains may result in a Newtonian creep behaviour if the distribution of stress is uniform over the bulk specimen, and that a small number of large grains may give rise to a power-law creep behaviour if the distribution of strain is uniform between differing sized grains. In the uniform stress case, at a given average grain size, a given grain-size distribution and in a given range of stress, as the temperature increases, the number of grains deforming by diffusional creep will increase. The consequence is that the observed creep behaviour becomes closer to that for pure diffusional creep (e.g. $n \approx 1$). However, this result is not consistent with the observation on the SIN SiC with different grain sizes. For example, the material with the largest average grain size exhibited a behaviour with $n = 1.39$ at a lower temperature of 1820 K, but showed a behaviour with $n = 2.46$ at a higher temperature of 2020 K [31].

3. Theories predict that N–H creep with a grain-size exponent $p = 2$ is dominant at higher temperatures, while Coble creep with $p = 3$ is dominant at lower temperatures. That is, at given stresses and grain sizes, with increasing temperature, the value of p is expected to decrease. However, the observed p value for SiC increased from 2.5 to 4 as the temperature was increased above T_c.

Therefore, based on the above considerations, it seems that the grain-size dependence observed in the SIN SiC [31] may not be directly related to diffusional creep.

Although the grain-size dependence was not systematically studied and there are no reliable data, it may be illustrative to plot all available data on SiC materials with grain sizes from 2–65 μm for which Newtonian creep has been observed. Fig. 1 shows such a plot. All data were recalculated to a temperature of 2020 K using activation energies determined in experiments on individual materials. Data on both the HP and the SIN SiC were included.

It is shown that most datum points fall in a narrow range of $\dot{\varepsilon}/\sigma$ (less than one order of magnitude). If possible effects of impurities and porosity on creep rate are neglected, it may be considered that there is no evidence for an inverse dependence of creep rate upon grain size, as predicted by diffusional creep. The HP SiC with the largest grain size (65 μm) actually deformed at strain rates much faster than all other SiC materials with smaller grain sizes.

Differing from that in diffusional creep, theoretically, there should be no grain-size dependence in H–D creep as in power-law dislocation creep due to the operation of a dislocation process(es) within

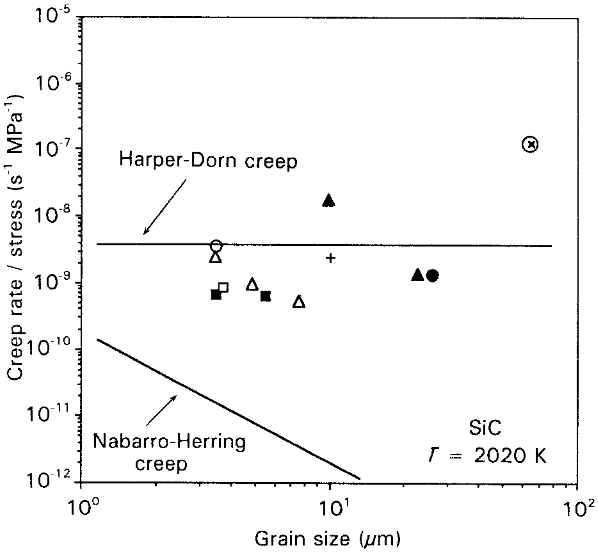

Figure 1 Plot of Newtonian creep data on SiC polycrystalline materials. It is shown that there is no grain-size dependence and the measured creep rates are drastically faster than those predicted by the model of Nabarro–Herring creep, but are in reasonable agreement with those predicted by the model of Harper–Dorn creep. (●) [36], (HP); (▲) [37], (HP); (■) [35], (HP); (⊗) [40]; (△) [31], (SIN); (□) [30], (SIN); (○) [32], (SIN); (+) [38], (SIN).

grains. The available data plotted in Fig. 1 suggest that the observed Newtonian creep in the HP and the SIN SiC is probably independent of the grain size over the grain-size range covered (2–65 μm). This observation supports the suggestion that these SiC materials may be controlled by H–D creep.

3.5. Creep rates

To make a reasonable suggestion for a control of diffusional creep or H–D creep, one also needs to consider the difference between the strain rates measured experimentally and those predicted by theoretical models. This difference is estimated as follows.

The rate equation for N–H diffusional creep may be written as ([1, 2])

$$\dot{\varepsilon} = \frac{14\Omega\sigma}{kTd^2}D_{\mathrm{L}} \qquad (2)$$

where Ω is the molecular volume ($\approx 0.7\boldsymbol{b}^3$ [57]). For the application of this model, D_{L} in Equation 2 should be the diffusion coefficient of the slower moving species in compounds [57].

Fig. 2 shows a plot of the variations of lattice diffusion rates with temperature of both silicon and carbon in both single and polycrystals of SiC. The results are (a) the activation energy is approximately the same for lattice diffusion of silicon and carbon, (b) silicon is the slower diffusion species in SiC, and (c) the diffusion rate of SiC is about two to three times slower in polycrystals than in single crystals at a temperature of 2020 K.

The diffusion data for silicon in the polycrystalline material of SiC is chosen for predicting the N–H creep rate ($D_{\mathrm{L}} = 2.18 \times 10^{-20}\ \mathrm{m^2\,s^{-1}}$ at 2020 K). The predicted rate (actually the ratio of $\dot{\varepsilon}/\sigma$) at $T = 2020$ K is presented also in Fig. 1 for comparison. It is obvious that the predicted rate is > 1.5 orders of magnitude smaller than the observed rate at a given stress and grain size. The difference increases drastically with an increase in the experimental grain size. For example, the difference is up to three orders at a grain size of 10 μm.

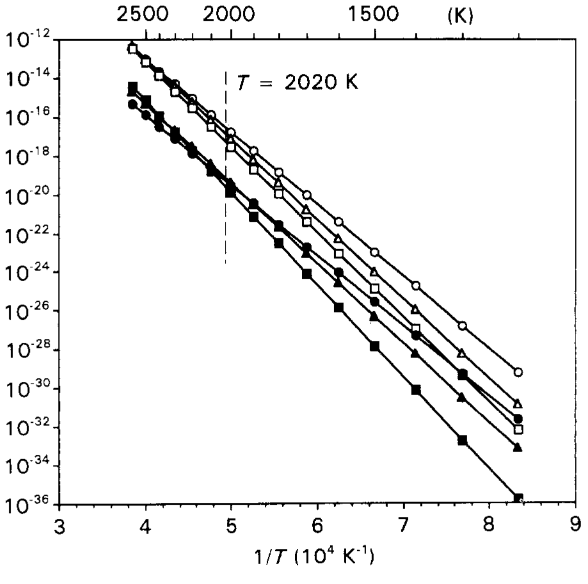

Figure 2 (○, △, □) Carbon and (●, ▲, ■) silicon self-diffusion rates in single and polycrystals of SiC. Silicon is the slower diffusion species. Pure single crystal: (○) [50], (●) [51], Nitrogen-doped single crystal; (△) [50], (▲) [51]. Dense polycrystalline: (□) [52], (■) 53.

Similar comparison in the creep rate between the model of Coble creep and the experimental observation cannot be meaningfully made because the grain-boundary diffusion coefficient on the slower moving species of silicon has not yet been determined.

The rate-controlling flow process in H–D creep was illustrated to be dislocation climb under saturated conditions in most materials for which this creep has been observed [27]. A rate equation was suggested to permit a prediction of the H–D creep rate.

$$\dot{\varepsilon} = A_{\mathrm{HD}}\frac{D_{\mathrm{L}}G\boldsymbol{b}}{kT}\left(\frac{\sigma}{G}\right) \qquad (3)$$

with A_{HD} given by

$$A_{\mathrm{HD}} = \frac{1.4\,\pi(\tau_{\mathrm{p}}/G)^2}{-\ln(\tau_{\mathrm{p}}/G)} \qquad (4)$$

where τ_{p} is the Peierls stress of a crystal.

In order to predict the H–D creep rate for SiC, it is necessary to know its value of τ_{p}. In practice, τ_{p} is defined as the minimum shear stress required to cause dislocation glide in the crystal lattice. Although the value of τ_{p} is not known with a high degree of precision, it may be estimated approximately from the theoretical relationship [58]

$$\tau_{\mathrm{p}} = \frac{2G}{1-\nu}\exp\left(-\frac{2\pi\eta}{\boldsymbol{b}}\right) \qquad (5)$$

with η given by

$$\eta = \frac{(3-2\nu)}{4(1-\nu)}L \qquad (6)$$

where ν is Poisson's ratio and L is the distance between the atomic planes.

In estimating τ_{p} for SiC, L is taken as the distance between the most closely packed atomic planes in the lattice (one-sixth of the unit lattice dimension in the c-axis orientation), $\boldsymbol{b}$ as the atomic distance in the most closely packed direction on the most closely packed planes (taken as the unit lattice dimension in the a-axis orientation).

With $\nu = 0.17$ [59] and $L/\boldsymbol{b} = 0.817$, from Equation 5, $\tau_{\mathrm{p}}/G = 3.91 \times 10^{-2}$. Such determined value of τ_{p}/G represents the lower limit of the true value of SiC at $T = 0$ K. Owing to thermal vibration of the atoms, Dietze [60] estimated that the value of τ_{p}/G would decrease by a factor of ~13 between 0 K and the melting point. Thus, for SiC, $\tau_{\mathrm{p}}/G \approx 4.61 \times 10^{-3}$ at $T = 2020$ K.

With this Peierls stress and $D_{\mathrm{L}} = 2.18 \times 10^{-20}$ for the diffusion of silicon at 2020 K, the predicted H–D creep rate for SiC is also included in Fig. 1. Inspection shows that the difference between the predicted rate and the experimentally observed rates is less than one order of magnitude for all SiC materials, except for that having the largest grain size (65 μm).

The above comparison between the strain rates measured experimentally and those predicted by theoretical models strongly suggests that the Newtonian creep of the HP and the SIN SiC polycrystalline materials can be described by H–D creep much better than by N–H creep.

Therefore, based on the foregoing considerations on creep curves, dislocation substructures, creep activation energies, grain-size dependencies and the measured creep rates, it may be concluded that the Newtonian creep behaviour observed in polycrystalline materials of SiC was more likely to be induced by H–D creep than by diffusional creep.

In the study on a SIN and two HP SiC materials [35], a Newtonian creep behaviour was observed at stresses even up to 1700 MPa and the transition stress from Newtonian creep at lower stresses to non-Newtonian creep at higher stresses decreased with an increase in temperature. For example, the HP SiC with a grain size of 5 μm showed the transition at 1650 MPa at 1603 K, but at 740 MPa at 1763 K. These observations may be because SiC has a very high Peierls stress and the Peierls stress decreases with increasing temperature.

4. Conclusion

Some previous studies on HP and SIN SiC polycrystalline materials have been re-examined. Mechanical data and microstructures suggest that the Newtonian creep behaviour observed in these SiC materials were induced by a dislocation process operating in Harper–Dorn creep, rather than by diffusional creep as concluded before. The supporting evidence for this suggestion is that (1) extensive dislocation substructures was developed, (2) there was no indication that the observed Newtonian creep depended on grain size, and (3) the measured creep rates were far smaller than that predicted by the model of diffusion creep, but were within one order of magnitude of that predicted by the model of H–D creep.

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