Yield Strength of Diamond

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The temperature dependence of the yield stress of diamond is predicted from existing data of crystals with diamond or diamondlike structure: silicon, germanium, and silicon carbide.

Diamond is the hardest material, but little is known about its plastic properties such as the yield strength [1]. For example, it is not well known at what temperature it can be deformed on a macroscopic scale. The highly covalent crystals Si, Ge, and SiC are hard to deform at room temperature and deformable only above several hundred degrees Celsius [2–8]. The temperature dependences of the critical shear stresses τ_c for $\{111\}\langle 1\bar{1}0\rangle$ slip in Si [4,5], Ge [7], and for $\{0001\}\langle 11\bar{2}0\rangle$ slip in $6H$-SiC [8] are shown in Fig. 1. They were measured at strain rates of the order of 10^{-4} s^{-1}. Under atmospheric pressure plastic deformation of Si is possible at $T > 700$ K where τ_c is lower than 100 MPa [2–4]. Under confining pressure Castaing et al. [5] deformed Si down to 500 K and observed a steep increase of τ_c to about 1 GPa. Germanium [7] behaves similarly. The hexagonal polytype $6H$-SiC has a layer structure consistent with the zinc-blende structure. Slip occurs on the $\{0001\}\langle 11\bar{2}0\rangle$ basal slip system [8], which is equivalent to the $\{111\}\langle 1\bar{1}0\rangle$ slip system in cubic zinc-blende.

The $\tau_c(T)$ relations of the three crystals are replotted in Fig. 2 in reduced units, τ_c/G vs kT/Gb^3, where k is the Boltzmann constant, b is the length of the Burgers vector of perfect dislocations, and G is the shear modulus, which is taken to be $G = (c_{11} - c_{12} + c_{44})/3$ for Si and Ge, and $G = c_{44}$ for $6H$-SiC. The values of b and G obtained from the elastic constants c_{ij} [9–12] are given in Table I, including the values for diamond.

The strain rate $\dot{\varepsilon}$ obeys a rate equation of Arrhenius type,

$$\dot{\varepsilon} = \dot{\varepsilon}_0 \exp[-\Delta H(\tau)/kT], \qquad (1)$$

where $\dot{\varepsilon}_0$ is insensitive to τ and T, and $\Delta H(\tau)$ is the activation energy of the process determining the mobility of the dislocations under the applied stress τ. At constant $\dot{\varepsilon}$ Eq. (1) gives the relation between τ and T as

$$\Delta H(\tau) = \alpha kT, \qquad (2)$$

with a constant α, which is usually about 25. There is no doubt that the dislocation mobility in the crystals under investigation is determined by overcoming a Peierls potential due to the lattice periodicity [13,14]. For the

Peierls process $\Delta H(\tau)$ is generally expressed in the form

$$\Delta H(\tau) = Gb^3 f(\tau/\tau_P), \qquad (3)$$

where τ_P is the Peierls stress and the function f represents the details of the process; kink pair formation of kink migration in dissociated partials or in nondissociated perfect dislocations. Combining Eqs. (2) and (3),

$$f(\tau/\tau_P) = \alpha kT/Gb^3. \qquad (4)$$

Figure 2 implies that the processes determining $f(\tau/\tau_P)$ in the three crystals are the same with the same magnitude of τ_P/G and constant α. Because of their covalency, the Si, Ge, and $6H$-SiC data coincide better than the bcc ones [15–17], which scatter by a factor of 2. This fact indicates homologous behavior and suggests that diamond conforms to it as well. In Si and Ge glide dislocations are dissociated into Shockley partials [14]. Dislocations in diamond are dissociated in the same way
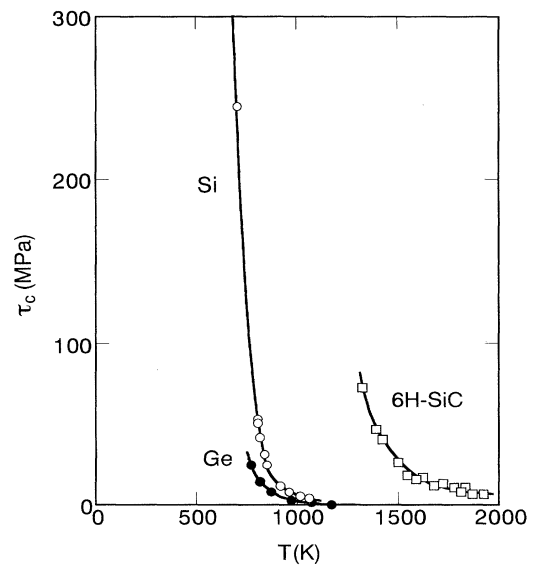

FIG. 1. Temperature dependence of the critical shear stress $\tau_c(T)$ for $\{111\}\langle 1\bar{1}0\rangle$ slip in Si [4,5], Ge [7], and for basal $\{0001\}\langle 11\bar{2}0\rangle$ slip in $6H$-SiC [8].
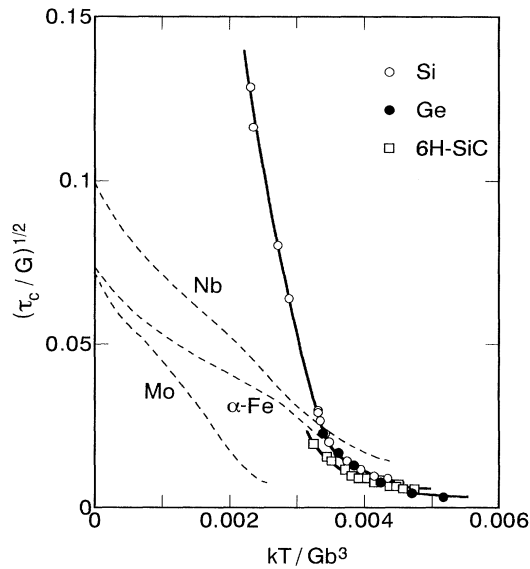

FIG. 2. $\tau_c(T)$ of Si, Ge, and 6H-SiC in reduced scales. The scaling of τ_c/G is taken as $(\tau_c/G)^{1/2}$ so as to enlarge the low stress region and shorten the high stress region. Broken lines are the data of bcc transition metals α-Fe [13], Nb [14], and Mo [15].
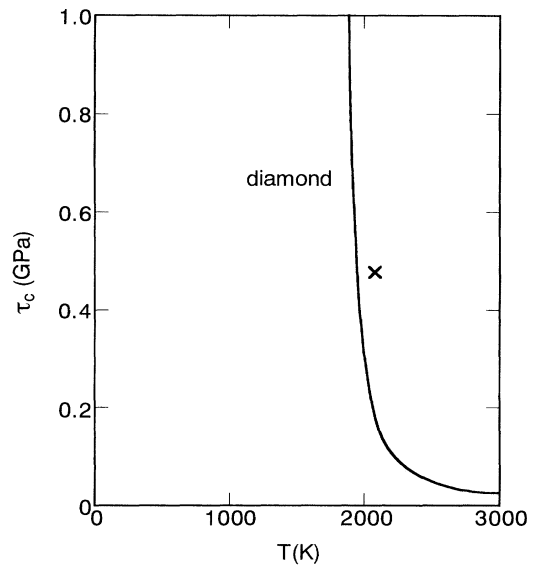

FIG. 3. Predicted temperature dependence of the critical shear stress for $\{111\}\langle 1\bar{1}0\rangle$ slip in diamond. Evans and Wild [20] detected plastic bending of type II diamond platelets at stresses higher than the point ×.

[18]. Also in SiC they should be largely dissociated because it has a low stacking fault energy consistent with the existence of polytypes [19]. Thus we conclude that in all these crystals dislocation mobility is controlled by the same intrinsic process and the τ_c/G vs kT/Gb^3 relation of diamond should obey the universal curve in Fig. 2. With the values of b for the $\{111\}\langle 1\bar{1}0\rangle$ slip system and G given in Table I and assuming the same value of α, we obtain for diamond the $\tau_c(T)$ of Fig. 3. Of course, our results refer to the intrinsic deformation process of type II diamond (without precipitated second phase) and not to type I diamond (where nitrogen platelets cause precipitation hardening). We consider the intrinsic yield stress, without work hardening of diamond.

In three point bending of type II diamond plates at 2073 K, Evans and Wild [20] detected appreciable plastic bending above a resolved shear stress of 0.5 GPa. Their data point is shown in Fig. 3. Similar experiments were carried out by Trefilov, Milman, and Grigoriev [21]. Although there is some inconsistency between their bending and indentation tests, they nevertheless indicate

the onset of plastic deformation around 2000 K at stresses of about 1 GPa, the range predicted by us.

One concludes that the deformation processes active in diamond are the same as those in the homologous semiconductors Si, Ge, and SiC.

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TABLE I. The length of the Burgers vector b and the shear modulus G as defined in the text.

		Si	Ge	6H-SiC	Diamond
b	(nm)	0.384	0.400	0.308	0.252
G	(GPa)	60.5	48.9	192	510

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