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Intrinsic Strength of Crystals and Defromability of Ceramics

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Crystal plasticity is primarily determined by the Peierls stress and the number of slip systems operative. According to the recent investigation of the present authors, the experimental value of the Peierls stress $\tau_{\rm P}$ is in good correlation with the crystal-geometrical factor h/b of the slip system (h being the spacing of slip planes and b the magnitude of the Burgers vector), and the magnitude of $\tau_{\rm P}$ is roughly given by the classical theory of Peierls-Nabarro. For the crystals with relatively low $\tau_{\rm P}$ the maximum temperature T_0 of the Peierls mechanism is determined by the theory based on the smooth kink model, while for the crystals with large $\tau_{\rm P}$, covalent crystals, T_0 is given by the abrupt kink model. To deform crystals, especially polycrystals, it is necessary to operate five independent slip systems (von Mises) Thus, the intrinsic strength of crystals is essentially determined by the crystal geometry, and one can derive a general rule to predict the plastic deformability of crystals, if only the crystal structure in known. Following the general rule, the possibility of ductile ceramics is argued, and the deformability of high- T_c superconducting ceramics is investigated, as an example.

§1. Introduction

Ceramics are hard and brittle, in general. The reason has, however, not been discussed on the basis of the theory of dislocations or crystal plasticity. The strength of crystalline solids is determined by the resistance to plastic deformation which occurs uaually by the glide motion of dislocations. The Peierls stress τ_P is the intrinsic resistance to dislocation glide on a particular slip plane, which is due to lattice periodicity. Thus, the Peierls stress τ_P is an essential quantity to discuss plasticity or strength of crystals, unless extrinsic hardening is dominant.

The present authors^{1,2)} have recently found that the Peierls stress τ_P is in good correlation with a crystalgeometrical factor suggested by the classical theory of Peierls³⁾ and Nabarro,^{4,5)} as far as the order of magnitude. Following this fact, one can predict the magnitude of τ_P for a given particular crystal and a given slip system from the crystal-geometrical consideration.

Another important factor for the discussion of the plasticity of crystals, especially of polycrystals, is the number of independent slip systems, which concerns the requirement of the von Mises condition.⁶⁾ The number of independent slip systems in a given crystal structure is also given by the consideration of crystal geometry only.

This paper is an argument about the plsticity of crystalline solids, preferentially ceramics, from the crystallographic view points. In sections 2-4, it will be realized that the intrinsic strength of crystals is essentially determined from the crystallographic consideration only, and that one can predict the condition under which a given crystal is easy to deform plastically. The general procedure to predict the deformability of crystals is presented in §5. In §6 the plasticity of ceramic or non-metallic materials is discussed and the possibility of ductile ceramics is examined. The final section will provide an application to high- T_c superconducting ceramics, as an example.

§2. Peierls Stress and Crystal Geometry

Diamond is very hard, while copper is easy to deform plastically. This is considered to be due to the difference of the Peierls stress τ_P between them, and τ_P of diamond is supposed to be about 10⁵ times as large as that of f.c.c. metals. This big difference of τ_P is often attributed to the nature of atomic bonding, i.e. covalent bonding or metallic bonding. However, in the past, no critical argument was made about the essential factor which determines τ_P of various kinds of crystals. At the present time we have the experimental values of τ_P for many types of crystals, covering metals, ionic compounds, semiconductors and ceramics. Thus, it should be possible to discuss a universal rule which may account for the difference of τ_P among the various crystal systems.

Peierls³⁾ and Nabarro^{4,5)} estimated the Peierls stress τ_P without considering the details of the interatomic bonding, as

$$\frac{\tau_{\rm P}}{G} \sim \frac{2}{1-\nu} \exp\left(-\frac{2\pi}{1-\nu}\frac{h}{b}\right),\tag{1}$$

where G is the shear modulous, v the Poisson's ratio, h the spacing between atomic layers parallel to the slip plane, and b the magnitude of the Burgers vector **b**. The quantity h/b is given if only we know the crystal structure and the slip system. Thus, we can regard h/b as the "crystal-geometrical factor" of the slip system. This classical model of Peierls and Nabarro was followed by re-examinations ans modifications, which give similar (exponential) dependence of τ_P/G on h/b with different numerical factors.⁷⁻⁹ These classical theories suggests that the peierls stress τ_P is strongly dependent on crystal geometry.

With the development of electron computers, accurate calculations of τ_P have been performed for b.c.c. metals and ionic crystals of NaCl-type, and compared with experimental values. However, since the computer simulation is possible only when the interatomic forces and the crystal structure are given, it is effective for the com-

parison of τ_P values among the crystals with same crystal structure, but ineffective for understanding the difference of τ_P between diamond and copper. The relation of τ_P and crystal structure have never been discussed, probablly because the experimental values of τ_P were known only for a limited number of crystal systems.

The Peierls stress τ_P is the stress necessary to move dislocations at the absolute zero. If we can measure the critical shear stress τ_c of high purity crystals near 0 K, we can determine $\tau_{\rm P}$. Actually, $\tau_{\rm P}$ values of b.c.c. metals,¹⁰⁻¹³⁾ B2 compounds^{14,15} and for {110} slip in ionic crystals of NaCl-type¹⁶⁻¹⁸⁾ have been determined accurately by the deformation tests at low temperatures. Even when plastic deformation is difficult at sufficiently low temperatures, $\tau_{\rm P}$ can be estimated in some degree of accuracy by the extrapolation of the temperature dependence of the critical shear stress τ_c with the aid of the theory of Peierls mechanism.¹⁹⁻²¹⁾ The $\tau_{\rm P}$ values for {100} slip in ionic crystals of NaCl-type^{18,22,23)} and semiconductors of zincblende type^{24,26)} have been deduced in this way. The crude estimates of τ_P for Si, SiO₂ and Al₂O₃ are still possible from the data of the temperature dependence of the yield strength at high temperatures.²⁷⁻²⁹⁾ The critical shear stress of f.c.c. metals,^{30,31)} basal slip in Mg³²⁾ and alkalihalides of CsCl type^{33,34} are small (<1 MPa) and only weakly dependent on temperature even around 4.2 K. They are not controlled by the Peierls mechanism, but by some extrinsic resistance to dislocation motion, for example by impurities. In these cases, the critical shear stress near 0 K should give an upper limit of τ_P . The τ_P values of bcc ³He and that for non-basal slip in hcp ⁴He were determined by a unique technique for the deformation of solid helium.^{36,37)} The present authors^{1,2)} have examined the correlation between these experimental values of $\tau_{\rm P}$ and the crystal-geometrical factor h/b of the slip systems. The result is shown in Fig. 1.

The values of h/b must be determined considering the following remarks. When the dislocations split into partial dislocations on the slip plane, b is taken to be the value for the partials. Thus, for f.c.c, metals and basal slip in Mg (h.c.p.) the Shockley partials are considered, i.e. $b=a/6\langle 112 \rangle$. For ionic crystals of NaCl type, we assume that dislocations of both primary $\{110\}$ and secondary {100} slip are not dissociated. In CsBr and CsI, (100) {011} dislocations can most probably split into two half dislocations,³⁵⁾ i.e. $b=a/2\langle 100 \rangle$. For Si and zincblende structure, splitting into Shockley partials is considered. In these semiconductors, furthermore, the value of h is different depending on whether dislocations can glide at glide set or at shuffle set. In Fig. 1 the value of h/b for semiconductors is taked to be that for Shockley partials gliding at glide set (0.353), but if they glide at shuffle set, h/b=1.06. The choise of h/b for SiO₂ and Al₂O₃ is not simple, because their crystal structures are complicated (a kind of hexagonal) and the possibility of dislocation splitting is not clear. In Fig. 1, we assumed that for SiO_2 and $Al_2O_3 h$ is the widest spacing of atomic layers perpendicular to c-axis and b the shortest translational vector.

Figure 1 shows that τ_P/G of various crystal structures is in good correlation with the crystal-geometrical factor h/b and the order of magnitude of τ_P/G is given by the relation of eq. (1) by Peierls and Nabarro. Only the primary slip system ({110} slip) in NaCl structure is exceptional. If $\langle 1\overline{10} \rangle$ dislocations in NaCl structure splits into two half dislocations on {110} slip plane, then h/b=1.0 and the experimental values of τ_P/G come just bove the straight line in Fig. 1. The possibility of the split-



Fig. 1. τ_P/G vs h/b. NaCl(1) and NaCl(2) denote $\langle 01\overline{1} \rangle \{011\}$ and $\langle 01\overline{1} \rangle \{100\}$ slip systems, respectively, in NaCl type ionic crystals. Solid line indicates eq. (1) with $\nu = 0.3$. (Takeuchi and Suzuki²)

for $\tau \simeq \tau_{\rm P}$.

ting of $\langle 1\bar{1}0 \rangle \{110\}$ dislocations in NaCl was argued by Haasen, ³⁸⁾ but the calculation of stacking fault energy³⁹⁾ and the computer simulations of the dislocation core structure^{40,41)} oppose to the splitting. The core structure of dislocations in NaCl structure and the magnitude of $\tau_{\rm P}$ are mysterious.

The large τ_P of semiconductors or some engineering ceramics has often been attributed to the large covalency of atomic bonding. However, Fig. 1 suggests that the covalent bonding itself is not necessarily direct reason for their large τ_P , and that small h/b is rather essential. We can regard that τ_P is determined by the crystal structure and the geometry of slip system, at least as an order of magnitude. The nature of atomic bonding should affect the magnitude of G, and details of interatomic potential relate to the difference of τ_P/G within one order of magnitude among the crystals with same structure.

It must be noted here that Fig. 1 does not necessarily confirm the validity of the classical theory of Peierls-Nabarro³⁻⁵⁾ nad also in its modifications made by some workers,⁷⁻⁹⁾ $\tau_{\rm P}$ was calculated from the changes of misfit energy accompanying the motion of rigid dislocations under zero applied stress. It is not clear whether such procedure gives true $\tau_{\rm P}$, the minimum stress necessary to move the dislocation. More appropriate treatment is desired to give theoretical basis to the relation of $\tau_{\rm P}/G - h/b$.

§3. Critical Temperature of Peierls Mechanism

At finite temperatures dislocations can overcome the Peierls potential by the nucleation of kink pairs with the aid of thermal energy, thus plastic deformation occurs under the stress smaller than $\tau_{\rm P}$. This mechanism is usually called the Peierls mechanism. The critical shear stress $\tau_{\rm c}$ determined by the Peierls mechanism is usually exhibits strong temperature dependence, and there exists a maximum temperature T_0 at which τ_c becomes zero. Above this T_0 the observed flow stress is determined by the extrinsic resistance to dislocation motion, and it shows usually weak temperature dependence as compared with that in the Peierls region. Therefore, the critical temperature T_0 is a practical measure of the deformability of the crystal. To be noted is that the Peierls stress $\tau_{\rm P}$, thus the critical temperature T_0 , is given for each individual slip system, different for different slip system, even in a same crystal.

The thermally activated motion of dislocations surmounting the Peierls potential can be devided into two regimes: the smooth kink regime and the abrupt kink regime. The smooth kink approximation is applicable to the crystals with relatively low τ_P , such as b.c.c. metals and ionic crystals. The abrupt kink approximation is applicable to the crystals with large τ_P , such as semiconductors.

(1) Smooth kink model

According to the theories of Peierls mechanism based on the line tension model of dislocations,¹⁹⁻²¹⁾ the activation enthalpy $H_{\rm kp}$ for a kink pair formation is given as a function of stress τ , like a curve in Fig. 2. The total activation energy $H_0=H_{\rm kp}$ ($\tau=0$), which is equal to twice the energy of an isolate kink, is given by

Fig. 2. Relation between stress τ and activation enthalpy for kink pair formation $H_{\rm kp}$, according to the theories based on smooth kink model (Ref. 19–21). $\tau_{\rm P}$ is the Peierls stress. H_0 is the total activation

energy given by eq. (2). Broken line indicates the asymptotic relation

$$H_0 = 2\alpha d \sqrt{b d\tau_{\rm P} T_{\rm L}}, \qquad (2)$$

where *d* is the period of the Peierls potential, $T_{\rm L}$ the line tension of the disloation and the numerical factor α expressing the shape of the Peierls potential is about 0.5. The estimate of $T_{\rm L}$ should be made by the theory of anisotropic elasticity, but usually $T_{\rm L} \sim Gb^2$ is assumed. Under a constant strain rate $H_{\rm kp}(\tau)/kT$ =constant (~30, usually), thus Fig. 2 expresses the temperature dependence of the flow stress τ . The critical temperature T_0 at which τ =0 is defined by $H_{\rm kp}(\tau)/kT$ = $H_0/kT \sim 30$. Using eq. (2), we have

$$T_0 \simeq H_0 / 30 \text{ k} = \alpha d \sqrt{b} d\tau_{\rm P} \overline{T_{\rm L}} / 15 \text{ k}.$$
(3)

It has been shown that the plasticity of pure b.c.c. metals and ionic crystals of NaCl type satisfies the relation of eq. (2) or eq. (3).²⁾

(2) Abrupt kink model

When $\tau_{\rm P}$ is large and the smooth kink model is not applicable, we must consider the abrupt kink model. In this case the energy of an isolate kink is roughly equal to the energy of a segment of dislocation core of a length d,⁴²⁾ and we can express the energy as

$$E_{k} = \gamma K b^{2} d, \qquad (4)$$

where the constant g is about 0.1 and K is the energy factor of the dislocation in the kink direction. The formation energy of a kink pair is then given by

$$H_0 \simeq 2E_k = 2\gamma K b^2 d, \tag{5}$$

and under a constant strain rate, we have

$$T_0 \simeq H_0 / 30 \,\mathrm{k} \simeq \gamma K b^2 d / 15 \,\mathrm{k}.$$
 (6)

Takeuchi *et al.*²⁾ investigated experimental values of H_0 for various semiconductors and showed that eq. (5) is satisfied with $\gamma = 0.1$. The energy factor K should be estimated by anisotropic elasticity theory, but often approximated by the shear modulous G.

Castaing *et al.*²⁷⁾ performed the plastic deformation of Si under hydrostatic pressures and measured τ_c down to



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about 500 K. In Fig. 3 are compared the temperature dependence of the critical shear stress τ_c of Si with τ_c for $\langle 1\bar{1}0 \rangle \{110\}$ slip in MgO. The data of MgO is well represented by the theory of the smooth kink model $(\tau_P/G \sim 10^{-2})$, while τ_c of Si exhibits the temperature dependence much steeper than the prediction from the smooth kink model and experimental T_0 (≈ 800 K) is too small to satisfy eq. (3).

The condition for the applicability of the smooth kink model or the abrupt kink model is not clear, but for $\tau_P/G \ge 10^{-1}$ the abrupt kink model is surely applicable. In the case of the abrupt kink, we need not know the value of τ_P to predict the deformability, since T_0 of eq. (6) is independent of τ_P and the temperture dependence of τ_c is very steep.

§4. Number of Independent Slip Systems

The plastic deformability is closely related to the number of slip systems which are actually operative or possible to be activated. The activity of each individual slip system is determined by the magnitude of τ_P of the slip system. Therefore, when the number of slip systems with low τ_P is limited, the slip systems with large τ_P are necessary to operate, thus the crystal exhibits relatively poor deformability.

To deform a crystal into any desired form, it is necessary to operate five independent slip systems, as pointed out first by von Mises.⁶⁾ The plastic deformation of policrystals or sintered materials is impossible unless the von Mises condition is satisfied. The independent number of the slip systems in typical crystal structures was investigated by Groves *et al.*,⁴³⁾ as the results shown in Table I.

In f.c.c. metals and b.c.c. metals the von Mises condition is satisfied by only one slip system: $\langle 1\bar{1}0 \rangle \{111\}$ and $\langle 1\bar{1}1 \rangle \{110\}$ slip system, respectively, of which independent number is five. This is one reason for the relatively high deformability of these metals. The slip systems of diamond structure and of zincblende structure are the same as that of f.c.c. metals, but these crystals possess poor deformability because of their large τ_P and high T_0 . In h.c.p. metals and ionic crystals, the von Mises condition cannot be fulfilled by only one slip system which has the smallest τ_P (primary slip system), thus the slip system with larger τ_P (secondary or higher order slip system) must operate to satisfy the von Mises condition.

Table I. Number of independent slip systems in typical crystal structures (Groves and Kelly⁴³).

crystal structure	slip s	independent number	
	<01 <u>1</u> >	{111}	5
h.c.c.	(111)	{011}	5
h.c.p.	$\langle 11\overline{2}0\rangle$	{0001}	2)
	$\langle 11\overline{2}0\rangle$	{1100}	2 } 4
	$\langle 11\overline{2}0\rangle$	{1101}	4)
diamond }	$\langle 01\overline{1}\rangle$	{111}	5
NaCl	$\langle 01\overline{1}\rangle$	{011}	2
11401	(011)	{100}	3
CsCl	<100>	{011}	3



Fig. 3. Temperature dependence of critical shear stress τ_c of Si (Ref. 27), compared with that for $\langle 110 \rangle \{001\}$ slip in MgO (Ref. 23).

It must be noted that the primary slip system and the secondary (or higher order) slip system are not always independent of each other. If the independent number of the primary slip system is n_1 , and that of secondary is n_2 , then the cumulative independent number integrating both slip systems is

$$n_{1+2} \leq n_1 + n_2.$$
 (7)

As seen in Table I, the equality holds for NaCl structure, while the inequality for h.c.p. metals. In h.c.p. structure, especially, the independent number of the slip system with $b = \langle 11\overline{2}0 \rangle$ is only four, as the cumulative number, thus another slip system is necessary, which has a component of b not paralell to the basal plane.

As a general rule, when the crystal has high crystallographic symmetry, it has a large number of independent slip systems which have low crystallographic indexes. On the contrary, in the crystals of low symmetry, the independent number of each individual slip system is limited.

§5. Recepé for the Prediction of Deformability

Following the arguments presented in sections 1–4, we can predict the plastic deformability of crystalline solids in some degree, if only we know the crystal structure and the shear modulus G. To be done is to find five slip systems which are independent of each other and have small T_0 . For one particular slip system, the independent number n and the crystal-geometrical factor h/b are given by the crystal geometry, then we can estimate τ_P/G from eq. (1) and the critical temperature T_0 from eq. (3) or eq. (6), depending on the magnitude of the estimated τ_P/G .

It is only in limited cases that the von Mises condition is satisfied by only one type of slip system, so that we must examine two or more types of slip systems so as to

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Fig. 4. Scheme for determination of T^* . $T_0^{(i)}$ is the critical temperature and n_i is the independent number of i-th slip system, n_i^c is the cumulative independent number up to i-th slip system.

find five independent slip systems. The examination should be made from the slip system of low crystallographic indexes which probably gives large h/b, thus small T_0 . Then, we must calculate the cumulative independent number starting from the slip system of the lowest T_0 in the order of the magnitude of T_0 untill it exceeds five, as shown in Fig. 3. In this way we can find T_0 of the fifth independent slip system. This T_0 , which will be refered to as T^* , is the temperature above which the crystal is easy to deform plastically.

The whole procedure can be summerized as follows:

(1) Consider the slip systems with low crystallographic indexes, and determine the crystal-geometrical factor h/b for each slip system. The possibil-ty of dislocation splitting must be examined.

(2) Estimate τ_P/G from eq. (1) for every slip systems and also for splitted partial dislocations.

(3) If $\tau_P/G < 10^{-1}$, then T_0 is estimated from eq. (3) for the smooth kink model.

(4) If $\tau_P/G \gtrsim 10^{-1}$, then T_0 is estimated from eq. (6) for the abrupt kink model.

(5) Arrange the slip systems in the order of the magnitude of T_0 .

(6) Accumulate the independent number of the slip systems in order starting from the primary slip system untill it becomes five or exceeds five.

(7) The T_0 of the fifth independent slip system, defined as T^* , is a measure for the deformability of the crystal.

In this procedure, the uncertainty of the dislocation splitting causes large ambiguity to the prediction of T^* , so that this recepé is not almighty, but a kind of guide to grasp the situation.

§6. Possibility of Ductile Ceramics.

Ductility or plastic deformabily is one of the important requirments for engineering materials. We can regard that the 5000 years history of metal technology has been based on the deformability or machinability of f.c.c. and b.c.c. metals. As described in sections 2–4, ductility or deformability of crystals is essentially determined by the crystal structure and its symmetry. When the crystal has high crystallographic symmetry, then it could have a slip system with large h/b, thus small τ_P/G , and the independent number of this slip system could also be large. On the contrary, in the crystals of low symmetry the independent number of the primary slip system should be limeted (<5), thus the secondary slip system which has large τ_P/G must be activated to deform the crystals.

As far as the von Mises condition, only $\langle 1\overline{10} \rangle |111 \rangle$ and $\langle 1\overline{11} \rangle \{011\}$ slip systems in cubic crystals can satisfy this condition. The deformability of non-cubic crystals is, therefore, determined by the activity of the secondary or the higher order slip systems which has larger τ_P than the primary slip system. Furthermore, in crystals of low symmetry h/b might be small even for the primary slip system. The crystal structure of ceramics are generally complicated and poor in symmetry. It is, therefore, quite natural that ceramics of hexagonal type such as α -Al₂O₃ and α -SiO₂ are hard and brittle.

In ionic crystals of NaCl type (cubic structure), $\langle 1\bar{1}0 \rangle$ {110} and $\langle 1\bar{1}0 \rangle$ {001} slip are usually the primary and the secondary slip systems, respectively. If $\langle 1\bar{1}0 \rangle$ {111} slip is possible in NaCl structure, the crystal should have high ductility, but there is no evidence that {111} slip occurs in alkali-halides of NaCl type. However, in AgCl $\langle 1\bar{1}0 \rangle$ {111} slip is much easier than {110} or {001} slip at low temperatures, and the estimated τ_P (=45 MPa) for {111} slip agrees with the prediction from eq. (1), provided that the dislocations split into two Shockley partials on {111} plane.⁴⁴ The T_0 for {111} slip in AgCl is about 100 K, thus it is quite easy to deform AgCl polycrystals at room temperature. The reason why {111} slip is operative in AgCl, but not in alkali-halides, is an interesting problem.

In ionic crystals of CsCl type (CsBr and CsI), $\langle 100 \rangle$ {001} slip has small τ_P , but its independent number is three. The secondary slip system is not known. If $\langle 11\overline{1} \rangle$ {011} dislocations in CsCl structure split into two halves as in β -CuZn, we can expect high ductility, but it seems impossible because such splitting brings ions of same sign face to face.

Ceramics are generally not mono-atomic but are compounds of two or more kinds of elements. In most monoatomic crystals, except diamond lattice, the Burgers vector **b** is the vector from one atom to its nearest neighbour atoms. In compounds, contrary, nearest neighbour atoms are usually of different kinds. The Burgers vector should be the vector between the second or third neighbour atoms. Therefore, **b** in compound crystals is relatively large, even in cubic crystals. This is one reason for small h/b of ceramics, and diamond lattice as well.

As a conclusion, dactile ceramics can be possible, when the crystal has a cubic or nearly cubic symmetry and h/b is relatively large, even though not the largest. The splitting of $\langle 1\overline{10} \rangle \{111\}$ or $\langle 1\overline{11} \rangle \{110\}$ dislocation is preferable to gain large h/b. These requirements for ductile ceramics are rather severe. It seems considerably hard to find ceramic materials which are ductile around the room temperature, even though they can possibly be ductile at elevated temperatures. There are some ceramics of cubic structure, such as perovskite, but they are still not sufficiently ductile, as shown in the next section.

§7. Formability of High- T_c Superconducting Ceramics.

For the engineering application of high- T_c superconductiong ceramics, their formability is one of the important problems. Some trials have been made to produce

wires of superconducting ceramics, but there have been made no discussion about the plasticity of such materials from the viewpoint of the theory of crystal plasticity. In this section we will try to predict the deformability of high- T_c superconductors of perovskite type, YBa₂Cu₃O_{7-x}, following the recepé presented in §5.

We will first consider the simple perovskite structure (Fig. 5), ignoring the layered structure and oxigen failures in YBCO superconductors. The slip systems in perovskite lattice examined here are listed in Table II. The spacing of {001} planes is the widest and $a\langle 100\rangle$ is the shortest translational vector, thus $a\langle 100\rangle$ {001} slip has the largest h/b, which gives $\tau_P/G \sim 3 \times 10^{-2}$. Since the independent number of this slip system is only 3, another slip system must operate. The spacing of {011} planes is the scond widest and that of {111} is the third. On {011} planes, $\mathbf{b} = a\langle 100\rangle$ is the shortest and $\mathbf{b} = a\langle 01\overline{1}\rangle$.

For $a\langle 01\overline{1} \rangle$ dislocations we must consider the possibility of splitting into partials. In BaTiO₃ the splitting of $a\langle 01\overline{1} \rangle$ dislocaions on $\{011\}$ planes into two half dislocations has been observed by TEM using weak beam method.⁴⁵⁾ For partial dislocations of this type, h/b=0.5, being equal to that for perfect $a\langle 100\rangle \{001\}$ dislocations. On $\{111\}$ planes, $a\langle 011\rangle$ dislocations can probably split into two Shockley partials, $a/3\langle 1\overline{2}1\rangle$ and $a/3\langle \overline{11}2\rangle$, as the splitting of $a/2\langle 01\overline{1}\rangle$ dislocations in f.c.c. metals. The splitting of $a\langle 100\rangle$ dislocations seems difficult on any plane.

The results of the examination are given in Table II. The lattice constant *a* is taken to be 0.385 nm, the length of the *a*-axis of YBa₂Cu₃O_{7-x}.⁴⁶⁾ For the evaluation of eqs. (3) and (6), we assume the conventional relations $T_{\rm L} \simeq Gb^2$ and $K \simeq G$, and G=40 GPa of sintered YBa₂Cu₃O_{7-x}.⁴⁶⁾ The period of the Peierls potential valley *d* in eqs. (3) and (6) is taken to be equal to *b*, the magnitude of the Burgers vector. In the last column of Table II the equation applied to estimate T_0 is indicated.

The temperature dependences of the critical shear stress τ_c predicted from the T_0 values in Table II are drawn in Fig. 6, schematically. If $\langle 01\overline{1} \rangle \langle 011 \rangle$ disloations split into two halves, as observed in BaTiO₃,⁴⁵ it gives the lowest T_0 , but the independent number of this slip system is only 2. The second smallest T_0 is that of $\langle 100 \rangle \langle 001 \rangle$





Fig. 5. Perovskite structure. (a) $\langle 100 \rangle \{011\}$ slip system. (b) Burgers vectors on $\{111\}$ plane; $\mathbf{b} = a \langle 01\overline{1} \rangle$ for perfect dislocation, and $\mathbf{b}_1 = \frac{1}{3}a \langle 11\overline{2} \rangle$ and $\mathbf{b}_2 = \frac{1}{3}a \langle \overline{121} \rangle$ for Shockley partial disloations.

Table II. Slip systems in perovskite structure. Independent number n, crystal-geometrical factor h/b, ratio of Peierls stress $\tau_{\rm P}$ to shear modulus G and critical temperature T_0 estimated by the indicated equation.

slip plane	b/a	n	h/b	$ au_{ m P}/G$	<i>T</i> ₀ (K)	eq.
{001}	<100>	3	0.50	3×10^{-2}	950	(3)
(001)	(110)	3	0.29	2×10^{-1}	3140	(6)
{011}	<100>	2	0.35	1×10^{-1}	1560	(6)
	$\langle 01\overline{1}\rangle$	2	0.25	3×10^{-1}	3140	(6)
{111}	$\frac{1}{2}\langle 01\overline{1}\rangle^{*}$	2	0.50	3×10^{-2}	610	(3)
	2(011)	5	0.20	5×10^{-1}	3140	(6)
	$\frac{1}{3}\langle 11\overline{2}\rangle^{**}$	5	0.35	1×10^{-1}	1090	(6)

*) half dislocation.

**) Shockley partial dislocation.



Fig. 6. Temperature dependences of the critical shear stress τ_c of four slip systems in perovskite structure. *n* is the independent number of individual slip system. Solid lines denotes perfect dislocations and broken line dissociated partials.

slip systems. The cumulative independent number of $\langle 01\bar{1} \rangle \{011\}$ and $\langle 100 \rangle \{001\}$ slip systems is still 3, less than the requirment of the von Mises condition. Therefore, the operation of $\frac{1}{3}\langle 11\bar{2} \rangle \{111\}$ slip is necessary to satisfy the von Mises condition, and it is sufficient as well, since the independent number of this slip system is 5. After all, T_0 for $\frac{1}{3}\langle 11\bar{2} \rangle \{111\}$ slip is T^* ($\approx 800^\circ$ C) for YBa₂Cu₃O₉ with simple perovskite structure, regardless the possibility of the splitting of $\langle 01\bar{1} \rangle \{011\}$ dislocations. If $\langle 01\bar{1} \rangle \{111\}$ dislocations do not split into Shockley partials, then $\langle 100 \rangle \{011\}$ slip should be activated, and T^* be much higher than 1000°C, being higher than the melting point.

Superconducting YBa₂Cu₃O_{7-x} is not of simple perovskite type, but has a three-layered structure containing oxigen failures. These modulations of the structure will dissolve the degeneracy of T_0 among the slip systems of same type. For example, T_0 for [100](001) slip should be slightly different from that for [001](100) slip. However, such changes should not alter the estimate of T^* seriously. We conclude here that 800°C or higher temperature is necessary to make plastic forming of YBa₂Cu₃O_{7-x} or other superconducting ceramics of perovskite type.

Finally, we report the results of deformation tests of $YBa_2Cu_3O_{7-x}$. The specimens were produced by Vacuum Metallurgical Co. Ltd. following the usual sintering procedure. The density of the specimens is about 74% of the ideal density. The compression tests of the specimens of

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Fig. 7. Results of deformation tests of sintered YBa₂Cu₃O_{7-x}. (a) Stress(σ)-strain(ε) relations. (b) Maximum stress σ_{max} of σ - ε relations against temperature *T*.

 $3 \times 3 \times 5$ mm³ were performed in air at the strain rate of 10^{-3} sec⁻¹. Figure 7(a) shows the stress-strain relations observed at several temperatures. Below 700°C, specimens break without any indication of yielding. Between 700 and 900°C, a maximum appears and failure occurs just after the maximum. At 930°C, resistance to deformation becomes small and the specimens deform up to several tens per cent. Quite interesting is that above 950°C the specimens deform even up to 100% by only very small stress (<0.1 MPa). The muximum value σ_{max} of the stress-strain relation decreases steeply with increasing the temperature, as shown in Fig. 7(b), and 930°C is the critical temperquire at which the resistance to deformation almost vanishes. It is not clear that the observed maxima of stress-strain curves correspond to true plastic yielding due to dislocation glide. However, if so, the agreement of the predicted $T^*(800^{\circ}C)$ with the observed critical temperature (930°C) is satisfactory, and it implies that the recepé presented in the preceding section is quite prospective.

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