Machinability and Microstructure of $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ Ceramics

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$(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ セラミックスの機械加工性と微細構造

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The relative grinding energy (RGE) and mechanical properties of $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics were measured. The $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics have much larger RGE than other ceramics, such as $\alpha\text{-Fe}_2\text{O}_3$ and ZnFe_2O_4 ceramics with mechanical strength, hardness and K_{IC} similar to those of $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics. Addition of a small amount of SiO_2 to $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics does not influence the mechanical properties appreciably but lowers the RGE to a great extent. These results were discussed on the basis of the modulus of rupture (MOR) of these ceramics under bending and compressive conditions.

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1. Introduction

In recent years, machinability of ceramics has been widely investigated.¹⁾⁻³⁾ The ceramics intensively studied were engineering ceramics such as SiC , Si_3N_4 and ZrO_2 which have large mechanical strengths, fracture toughness and hardness.

However, there has been no detailed study about the machinability of soft ceramics, because these ceramics, like as MgO , ferrites etc., were thought that they could be machined easily. In present study, we investigated the machinability of $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics. They have been interested as a useful substrate for metallic thin films because of its large thermal expansion.

We measured the relative grinding energy (RGE) at down grinding and the other mechanical properties. RGE of $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics was compared with those of $\alpha\text{-Fe}_2\text{O}_3$ and ZnFe_2O_4 , whose mechanical strength, hardness and K_{IC} are almost equal. RGE of $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics with an addition of 0.1 wt% SiO_2 was also measured. Correlation of RGE and the other mechanical properties and microstructure are discussed.

2. Experimental procedure

Sample preparation $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics were prepared from NiO and MgO powders. Two kinds of powder were weighed, ball-milled with an ethanol and calcined at 1000°C for 2 h in air. After that, the calcined powder was ball-milled again, dried, granulated and molded under

29.4 MPa (300 kgf/cm^2). The green compact was hot-pressed at several temperatures for 2 h under 29.4 MPa in air. The grain size of the sintered body was controlled by changing the hot-pressing temperature. $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics with an addition of 0.1 wt% SiO_2 , $\alpha\text{-Fe}_2\text{O}_3$ and ZnFe_2O_4 ceramics were prepared by the similar method as mentioned above.

The sintered bodies were lapped by SiC abrasives. The density was measured by the Archimedes method and the average grain size was measured by using a scanning electron microscope.

Machinability Machinability of the ceramics was evaluated by determining a relative grinding energy. The grinding experiments were performed using a 150 grid metal bonded diamond wheel, 10 cm diameter and 0.4 mm thickness, operating at 3000 rpm with cooling water.

Grinding was carried out with the cutting face of wheel moving in the same direction as the workpiece (down grinding). Test piece were ground at feed rate of 1 cm/min. The depth and length of cutting were 3 and 30 mm, respectively (low feed rate and long cutting depth). The consumption of electric power during grinding, E , was measured. The relative grinding energy, RGE, was calculated from the equation.⁴⁾

$$\text{RGE} = \frac{E}{\{(W_0 - W)/d\}} \quad (1)$$

where W_0 and W are the weights of the test piece before and after grinding respectively and d is the

density of sample. This measurement was performed five times and average value was calculated.

After cutting, the grinding surface was observed by a scanning electron microscope.

Mechanical properties A three-point bending strength, σ_b and Young's modulus, E_b , were measured with a test piece of $3 \times 4 \times 20$ mm by using a mechanical strength meter (Yonekura Co., Ltd.). The crosshead speed was 0.5 mm/min. A modulus of rupture under bending condition, MOR(b), was calculated from the equation.⁵⁾

$$\text{MOR(b)} = \frac{\sigma_b^2}{2E_b} \quad (2)$$

Similarly, a compressive strength, σ_c , and stress strain ratio, E_c , were measured with a test piece of $2 \times 2 \times 1$ mm. A modulus of rupture under compressive condition, MOR(c), was calculated from the equation.

$$\text{MOR(c)} = \frac{\sigma_c^2}{2E_c} \quad (3)$$

Vickers hardness, H_v was measured under a pressing of 29.4 MPa (300 kgf/cm²) for 10 s. The testing was performed with a Vickers hardness meter (Akashi Co., Ltd.). One surface of the test piece was lapped to a smooth surface as a mirror with diamond abrasives. Fracture toughness, K_{IC} , was obtained by the indentation microfracture method and it was calculated from the following equation.^{6), 7)}

$$K_{IC} = 0.203 H_v a^{1/2} (c/a)^{-3/2} \quad (4)$$

where a is an indentation impressive radius and c is a crack length.

All these measurements were performed ten times and average values were used.

3. Results and discussion

All data obtained in this study are shown in Table 1. There are four compositions; $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics, $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics

Fig. 1. Correlation of relative grinding energy (RGE) with inverse square root of grain size ($1/\sqrt{D}$).

with 0.1 wt% of SiO_2 , $\alpha\text{-Fe}_2\text{O}_3$ ceramics and ZnFe_2O_4 ceramics. The relative densities of these samples are almost equal to 100% except $\alpha\text{-Fe}_2\text{O}_3$ ceramics sintered at 900°C.

Figure 1 shows the relative grinding energy, RGE, of these samples as a function of average grain size. It is understood that RGE is proportional to inverse square root of grain size. Similar result was obtained by Rice and Speronello.²⁾ They interpreted the result in terms of Petch's relation. It is also clear in Fig. 1 that RGE of $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics without additives is much larger than that of $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics with an addition of 0.1 wt% SiO_2 .

Figure 2 shows RGE measured on different ceramics plotted as a function of hardness, H_v . Hardness of $\alpha\text{-Fe}_2\text{O}_3$ and ZnFe_2O_4 ceramics are larger than 6.3 GPa and their RGEs increase with the H_v s. For $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics with SiO_2 or

Table 1. Several properties of the ceramics.

Material	Sintering (°C)	Density g/cm³	%	Grain (μm)	σ_b (GPa)	E_b (GPa)	MOR(b) (MJ/m²)	σ_c (GPa)	E_c (GPa)	MOR(c) (MJ/m²)	H_v (GPa)	K_{IC} (MJ/m³/²)	R. G. E. (GJ/m³)
4NiO-MgO	1300	6.08	98.7	3.2	0.179	72.6	0.221	0.829	50.9	6.75	5.70	2.06	12.71
	1350	6.09	98.9	13.4	0.172	80.4	0.184	0.640	37.1	5.52	5.86	1.70	8.36
	1400	6.09	98.9	45.3	0.078	27.4	0.111	0.307	24.4	1.95	6.20	—	2.32
4NiO-MgO +SiO₂	1300	6.12	99.5	4.2	0.183	117.7	0.142	0.961	25.7	17.37	5.94	1.72	1.34
	1350	6.12	99.4	14.7	0.141	143.2	0.069	0.781	28.6	10.66	6.23	1.53	0.62
$\alpha\text{-Fe}_2\text{O}_3$	900	5.06	96.3	0.9	0.253	168.7	0.190	0.856	42.5	8.62	11.68	2.19	5.00
	1000	5.25	99.8	3.2	0.339	177.5	0.324	1.718	75.2	19.62	9.80	2.52	2.14
	1100	5.18	98.6	10.4	0.167	155.9	0.089	1.019	39.6	13.11	8.94	1.62	1.46
	1200	5.17	98.4	49.8	0.022	27.5	0.009	0.113	6.2	1.03	6.27	1.20	0.14
ZnFe₂O₄	1100	5.30	99.6	1.3	0.198	154.9	0.127	1.262	52.9	15.05	6.96	1.78	1.29
	1200	5.30	99.5	2.6	0.140	153.0	0.064	0.976	42.2	11.29	6.35	1.39	0.37
	1300	5.31	99.8	19.5	0.145	151.0	0.070	0.942	41.2	10.77	6.31	1.52	0.29

Fig. 2. Correlation of relative grinding energy (RGE) with Vickers hardness, H_v , of samples.

not, RGE decrease with H_v s below 6.3 GPa. It is understood that the addition of SiO_2 to $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics lowers RGE to a great extent but scarcely influences on H_v .

Figure 3 shows RGE measured on different ceramics plotted as a function of modulus of rupture under bending condition, MOR(b). MOR(b) is an elastic energy until fracture hap-

Fig. 3. Relative grinding energy (RGE) of several ceramics plotted as a function of the modulus of rupture under bending condition, MOR(b) of the samples.

pens under bending condition. For all samples except $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics without additives and $\alpha\text{-Fe}_2\text{O}_3$ ceramics sintered at 900°C , RGEs increase with MOR(b). This evaluation method was proposed by Gielisse et al.⁵⁾ In general, ceramic materials are brittle, so they are broken by tensile stress. RGE of such material is appraised by MOR (b). Therefore, we concluded that grinding of these ceramics except $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics without additive and $\alpha\text{-Fe}_2\text{O}_3$ ceramics sintered at 900°C are mainly caused by bending (tensile) fracture process.

$\alpha\text{-Fe}_2\text{O}_3$ ceramics sintered at 900°C are considered to have pores in fairly large quantities compared with other ceramics because its density is lower than the theoretical value. A pore becomes a fracture origin. Therefore, in spite of its small grains, $\alpha\text{-Fe}_2\text{O}_3$ ceramics sintered at 900°C has smaller bending strength and MOR (b) than sintered at 1000°C does. On the other hand, RGE is the energy consumption of material removal process and it does not strongly depend on the small pores contained in the test piece substantially. Therefore, as shown in Fig. 3, RGE of $\alpha\text{-Fe}_2\text{O}_3$ ceramics sintered at 900°C deviates from the smooth curve.

But the densities of $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics without additive is almost equal to 100% and their bending strengths are not smaller than those of $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics with additives. Their main difference is Young's modulus, E_b . So, there is other reason why $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics without additive deviate from the smooth curve

Fig. 4. Relative grinding energy (RGE) of several ceramics plotted as a function of K_{Ic} of the samples.

in Fig. 3.

Figure 4 shows RGE measured on different ceramics plotted as a function of K_{IC} . This figure looks like Fig. 3 exactly. In this study, K_{IC} is directly proportional to MOR(b).

In Fig. 5, RGEs are plotted on the graph with MOR(c) on ordinate and MOR(b) on abscissa. MOR(c) is an elastic energy until fracture happens under compressive condition. In this figure we could make a distribution of RGE on it. It is understood that RGE increase with MOR(b) under constant MOR(c) condition. But the sample which has large MOR(b) and large MOR(c), for example α -Fe₂O₃ ceramics sintered at 1000°C, has small RGE, 2.1 GJ/m³. In general, ceramics are brittle, so grinding process is almost due to brittle fracture, even if the plastically flowed regions are observed on the grinding surface.^(3),8) In fact, as shown in Fig. 3, RGEs of ZnFe₂O₄ and α -Fe₂O₃ ceramics only depend on MOR(b). In these ceramics, MOR(c) increase with MOR(b). Therefore, grinding is caused mainly by tensile fracture.

However, for (Ni_{0.8}Mg_{0.2})O ceramics with relatively large MOR(b) and small MOR(c) compared with other ceramics, compressive fracture is considered to occur in grinding process. These process can be explained as shown schematically in Fig. 6. When a diamond abrasive touches the test piece, elastic deformation at first takes

Fig. 5. The distribution chart of relative grinding energy (RGE).

Fig. 6. Schematic diagrams showing the grinding process of (a) α -Fe₂O₃ and ZnFe₂O₄ ceramics, (b) (Ni_{0.8}Mg_{0.2})O ceramics.

place. After that, for usual ceramics, tensile fracture occurs toward the grinding direction (Fig. 6(a)). In this case, grinding chips are easily ejected from the test piece. However, if the sample has large MOR(b) and small MOR(c), compressive fracture takes place to the opposite direction (Fig. 6(b)). In this case, the down grinding wheel make grinding chips to move to the back edge of the contact area between the grinding wheel and the test piece. Therefore, grinding chips are crushed and fill up the diamond abrasives. These effect increase RGE.

It is proved in Fig. 7 which shows the grinding surface of (Ni_{0.8}Mg_{0.2})O ceramics. The plastically flowed regions are observed on the grinding surface of (Ni_{0.8}Mg_{0.2})O ceramics without additives (Fig. 7(a)). These regions are the result of crushed chips.

In Fig. 7(b), the grinding surface of (Ni_{0.8}Mg_{0.2})O ceramics with an addition of 0.1 wt% SiO₂ is shown. The plastically flowed regions are apparently smaller than in Fig. 7(a).

As mentioned before, the addition of small amount of SiO₂ to (Ni_{0.8}Mg_{0.2})O ceramics makes the Young's modulus larger (Table. 1). This ceramics has much larger MOR(c) than the same ceramics without additive does. Therefore, this ceramics does not show compressive fracture in grinding.

As shown in Table.1, the densities of (Ni_{0.8}Mg_{0.2})O ceramics with SiO₂ additives are little larger than (Ni_{0.8}Mg_{0.2})O ceramics without additives, so SiO₂ improves sinterability. The energy dispersion spectrum of a fracture surface of (Ni_{0.8}Mg_{0.2})O ceramics with 0.1 wt% SiO₂ suggests that SiO₂ is segregated at grain

Fig. 7. SEM photographs of the grinding surface of (a) $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics, and (b) $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics with an addition of 0.1 wt% SiO_2 .

boundaries.⁹⁾ This segregated SiO_2 also change mechanical and machining properties of the ceramics. But the reason why these SiO_2 change them is not clear in this study. It need more detailed study to solve this problem.

4. Conclusions

The "soft" ceramics like as MgO , ferrites etc. were thought that they can grind easily. But $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics, which have small MOR (c) as compared with it's MOR(b), is difficult to down grinding because of its compressive fracture process. A little addition of SiO_2 to $(\text{Ni}_{0.8}\text{Mg}_{0.2})\text{O}$ ceramics make MOR (c) larger, and it become easily to grind.

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