The Strength of WC-Co Cemented Carbide in Relation to Structural Defects*

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The strength of cemented carbides has been investigated by many researchers in relation to the factors such as mean grain size of carbide, mean free path of binder, binder phase composition, etc. At present, however, the strength seems to remain unsolved. Thus, the present studies on the above subject are undertaken employing a new approach with an idea that the strength of cemented carbides would closely be connected with a small amount of microstructural defects in sintered specimens.

As a result, it was found that the defects which substantially correspond with micropores, coarse WC grains, and binder-pools usually acted as the source of fracture in conventional alloys, and the transverse-rupture strength of the alloy was directly controlled by the dimension, the location, etc., of the above defects. It was also found that the strength, for instance, of WC-10%Co alloy with no those defects would markedly increase up to a value higher than about 400 kg/mm² [according to the Japanese Industrial Standard (JIS)]. Therefore, more experiments would be needed to make clear the nature of the above factors in order to explain the strength of cemented carbides.

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I. Introduction

It is well known that various factors affecting the strength of WC-Co cemented carbides include the amount of the binder phase, mean grain size of carbide, grain size distribution, mean free path of binder, carbon content of the alloy (binder phase composition), sample volume, etc. The reviews concerning the effects of these factors have been published by Exner et al.⁽¹⁾ and by Suzuki⁽²⁾. Besides these factors, the stress-induced transformation of the binder phase⁽³⁾, the domain size of the binder phase⁽⁴⁾, the residual compressive stress due to grinding⁽⁵⁾⁽⁶⁾, etc., should be noted as other important ones.

However, the exact mechanism of these factors affecting the strength of cemented carbides seems to remain unsolved. In fact, the actual strength of the cemented carbides can hardly be explained by various theories regarding the strength such as proposed by Kreimer et al.⁽⁷⁾, Gurland⁽⁸⁾, and Nabarro et al.⁽⁹⁾ because only one or two of the above factors have

been incorporated into the theories.

On the other hand, it was confirmed in a recent experiment by the present authors⁽¹⁰⁾ that the transverse fracture of cemented carbides did not initiate on the tension surface, but usually in the interior of specimens, when the surface of specimens was in the ground state. Namely, the radial markings of fracture were always observed on the fracture surface and they started in a certain small area at a position very near the tension surface; and this small area showed a relatively smooth surface, corresponding to so-called white spots (a kind of fracture mirror⁽¹¹⁾ in glasses or polycrystalline ceramics). Thus, it was suggested that some microstructural defects should be contained in this spot and these defects directly controlled the strength of each specimen. Therefore, it is considered that the strength of cemented carbides can be clearly understood by examining the above defects which would act as a source of fracture. From this point of view, the present study was undertaken to make clear the characteristics of the fracture source and the strength of cemented carbides which depends upon dimensions and locations of the fracture source.

II. Specimens and Experimental Procedures

Using commercially available WC and Co

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powders as starting materials, a mixture of the powders was treated in a wet ball mill, dried, pressed, and subsequently vacuum-sintered at temperatures of 1350~1380°C for 1 hr. Two sorts of specimens, that is, the conventional WC-10%Co high carbon two-phase alloys with a mean carbide grain size of about 1.7 μ and WC-10%Co high carbon two-phase alloys with nearly the same mean grain size as above but having a small amount of coarse WC grains or binder-pools (intentionally introduced by intermixing the coarse carbide powder or by varying the sintering condition), were used in this study. The transverse-rupture strength of the former alloys was in the range of distribution of about 220~350 kg/mm² according to the Japanese Industrial Standard† (JIS H 5501) and that of the latter was about 100~320 kg/mm². It was shown by observation on fracture surfaces, as described later, that, for instance, micropores (one of defects) of less than about 120 μ in the larger axis were present in the white spot in both of the specimens. However, these pores were hardly observed on the polished cross-section by microscopy, because the number of pores in this case was extremely few compared with that of A-1 or B-1 type in ASTM Designation⁽¹²⁾. Four side surfaces of the sintered specimens were ground in parallel with their longitudinal axis by using a diamond wheel with about 70 μ diamond powder for obtaining JIS specimens.

After the transverse-rupture strength test, the fracture surface in each specimen was examined; and the dimensions of the white spot, and the dimensions and the location of the defects were first measured by using an optical microscope or a scanning electron microscope if necessary. Here, the location of the defects was expressed by the two quantities; that is, the distance from the tension surface to the defect (Δt) and the distance from the vertical plane through span center to the defect (Δl) as schematically shown in Fig. 1. Next, the relations between the trans-

Fig. 1 Schematic drawing of location of the source of fracture, Δt ; distance from the tension surface to the defect, Δl ; distance from the vertical plane through the span center to the defect. ($\times 5$)

verse-rupture strength (σ_m) and the dimensions of white spots or of structural defects, the relation between the dimensions of white spots and structural defects, and the relations between two sorts of dimensions as above and the external stress (σ_d) acting on the defects at the moment of fracture were investigated in detail to clarify the role of the defects in the observed strength of WC-Co cemented carbides.

III. Experimental Results

An example of the macrostructures of the white spots developed on the transverse-ruptured surface of the alloy is shown in Photo. 1. The examination of these macrostructures revealed that the white spot corresponded to a comparatively flat surface with

Photo. 1 An example of macrostructures of white spots observed on transverse-ruptured surface of WC-10%Co alloy. ($\times 5$)

† Dimensions of specimens—4 $\times$ 8 $\times$ 24 mm; Span—20 mm; Specimens were loaded with a cemented carbide bar of 2 mm ϕ diameter at the center of the span. The transverse-rupture strength determined according to JIS is usually lower than the value obtained by ASTM Standard and the former value is about 0.9 times the latter.

Photo. 2 Microstructures of three kinds of defects as source of fracture. (a) micropore (b) coarse WC grain (c) binder-pool. ($\times 400$)

a small area, from which radial markings of fracture started, and, in addition, it was generally on the plane perpendicular to the tension surface of the specimen.

The microstructural defects were always observed in the white spot as expected, as illustrated in Photo. 2. There appeared in general three kinds of defects, i.e., micropores (a), coarse WC grain (b), and binder-pool (c). In this case, coarse WC grain and binder-pool fractured transgranularly. However, as the grain size of coarse WC decreased towards the mean grain size of the specimen (about 1.7μ), the interfacial fracture across WC/WC or WC/binder naturally increased by a small amount. As to the micropores, step patterns were observed on the inner surface as shown in Photo. 3(b), which indicates that the pore had actually existed during the sintering stage.

In this study, the dimension of white spots was conveniently expressed by their diameters because they were usually almost circular or semicircular when they were very near the surface; the dimensions of the three kinds of defects were expressed by the length of the major axis ($2a$), assuming that the shape of defects on the fracture surface was approxi-

Photo. 3 Scanning electron microstructures of white spot (containing a pore) and its neighborhood (a), and inner surface of pore (b). White spot is comparatively flat surface, enclosed by arrows. (a) $\times 20$ (b) $\times 1800$

Fig. 2 Relations between transverse-rupture strength and the diameter of white spots.

mated to an ellipse. In rare cases of interfacial fractures, the interfacial crack length (substantially approximate to the dimension of WC grain) was taken as the dimension of the defect.

The relation between σ_m and the diameter of white spot, and that between σ_m and the dimension of defects are shown in Fig. 2 and Fig. 3, respectively. Figure 4 shows the distribution of σ_m as a function of the location of defects. Figures 2~4 show that no close correlations were observed between σ_m and these variables. Figure 5 shows that the relations between the

Fig. 7 Relations between σ_d and dimensions of three kinds of defects, showing σ_d has a close correlation with $2a$.

shown in Figs. 2, 3, and 4 respectively, would be reasonable, when it is considered that σ_m is a function of three variables, i.e., the dimension of defects (or white spots), Δt , and Δl as shown in eq. (4) derived below. Accordingly, the results of Figs. 5 and 6, especially that of Fig. 7, would be important among the results obtained during this study.

Therefore, these results will be discussed in the following. It can be seen in Photo. 3(a) that the faint radial markings exist also in the white spot and are linked with the clear markings observed in the outside area. It is suggested that the fracture mode in cemented carbides is a radial fracture from the beginning and not a fibrous one as observed in high strength steels. Thus, the smooth area of the white spot would be formed during the period when the initial crack propagation proceeded at a relatively low velocity⁽¹¹⁾. Therefore, the result of Fig. 5 that the diameter of white spots increased with increasing dimension of the defects, is considered to be due to the fact that the velocity did not rapidly increase in the case of large defects, because the σ_d decreased with increasing dimension of defects as shown in Fig. 7. The result of Fig. 6 will be justified by considering the combined results of Figs. 5 and 7. The idea that the diameter of white spots would not

correspond with the so-called critical crack length in brittle fracture will be discussed later.

The result of Fig. 7 that σ_d decreased with increasing dimension of defects, was subsequently examined. As for pores, assuming that the stress concentration factor at the end of the pore is $(1 + 2\sqrt{a/\rho})^{\dagger(13)}$, (where ρ is the effective radius of the curvature of the inner surface of the pore), the criterion for the fracture of the sound matrix of WC-10%Co alloy would be expressed by the following equation (where σ_0 is the strength of the sound matrix around the pore),

$$\sigma_d(1 + 2\sqrt{a/\rho}) = \sigma_0 \quad (2)$$

or,

$$\sigma_d^{-1} = \sigma_0^{-1} + 2\sigma_0^{-1}\rho^{-1/2}\sqrt{a} \quad (2')$$

Therefore, when Fig. 7 is replotted by the relation between σ_d^{-1} and $\sqrt{a}$, a linear relationship will be expected.

But in the case of fracture due to coarse WC grain or binder-pool, some considerations would be needed for replotting the results in the same form, because the stress concentrations⁽¹⁴⁾ due to the difference in shear modulus between WC-10%Co matrix ($G = 25000 \text{ kg/mm}^2$ ⁽¹⁵⁾) and those defects ($G_{WC} = 30400 \text{ kg/mm}^2$, $G_{Co} = 7900 \text{ kg/mm}^2$) should be kept at different values in each defect, independently of dimensions. However, the correlations in Fig. 7 were nearly in the same way in three kinds of defects as mentioned above, showing the stress concentration factor of WC grain or binder-pool should be equal to that of the pore. Therefore, it may be reasonable to assume that coarse WC grain or binder-pool is plastically deformed under the stress lower than σ_d , and as the result, a crack with the dimension of $2a$ is formed through WC grain or binder-pool, and it will play the same role as pores. This is the reason why the coarse WC grain or binder-pool are called a kind of defects in this paper.

From these considerations, the relations between σ_d^{-1} and $\sqrt{a}$ are plotted as shown in

[†] The defect will be expressed in three dimensions. Then, the stress concentration factor is not considered established in such a simple form; but for the first attempt, it is assumed to be a two-dimensional factor for convenience.

Fig. 8 Relations between σ_d^{-1} and $\sqrt{a}$. The data points encircled by a broken line are excluded in the estimation of the straight line.

Fig. 8†. It is clear that the linear relationship is held for each defect as expected, and it is approximately expressed by the following equation:

$$\sigma_d^{-1} = 1.2 \times 10^{-3} + 26 \times 10^{-3} \sqrt{a} \quad (3)$$

or

$$\sigma_d = 830 / (1 + 2\sqrt{a/8.5 \times 10^{-3}}) \quad (3')$$

(where the unit of a is mm).

From this equation the values of σ_0 and ρ are

† The data points enclosed by a broken line in Fig. 8 were mainly related to the fracture of WC grains (with $2a \approx 2 \sim 7 \mu$) having generally existed in the layer up to about 50μ from the tension surface of the specimen; and these points were considered to be influenced in part by the residual tensile stress introduced by the surface grinding⁽⁵⁾⁽⁶⁾. Therefore, these points were excluded from the estimation of the straight line using the method of least squares. The value of σ_d^{-1} was in the range of $(1.0 \sim 1.4) \times 10^{-3} \text{ kg}^{-1}/\text{mm}^2$ for micropore and binder-pool according to the axial ratios, and it was in the range of $(1.0 \sim 1.2) \times 10^{-3} \text{ kg}^{-1}/\text{mm}^2$ for WC grain. These results showed that the effects of three kinds of defects were nearly identical as mentioned above, and the straight line in the figure was calculated from all data relating to three kinds of defects.

known to be about 830 kg/mm^2 and 8.5μ , respectively.

The meanings of the above values were next examined. The σ_0 has already been defined as the strength of the matrix free from defects. The ideal strength of WC and that of the binder phase should be about 4600 and 1200 kg/mm^2 , respectively, if the ideal strength of WC could also be approximated by $0.15G^{(16)}$. Then, the above value of 830 kg/mm^2 for σ_0 suggested that the strength of cemented carbides was brought about by the binder phase dispersion-hardened by WC particles. On the other hand, the value of ρ seemed to be considerably larger than expected, but it would be still reasonable within the accuracy of the data in view of a slight plastic deformation that might take place at the edge of defects.

In connection with Fig. 8, it was confirmed that a linear relationship was also held between σ_d^{-1} and the square root of the half of the diameter of a white spot. However, the corresponding value of σ_0 obtained in this case was negative, supporting the above idea that the diameter of the white spot would have no bearing on the critical crack length.

Combining eqs. (1) and (3'), the following equation was derived:

$$\sigma_m = \{830 / (1 + 2\sqrt{a/8.5 \times 10^{-3}})\} \{t / (t - 2\Delta t)\} \times \{l / (l - 2\Delta l)\}. \quad (4)$$

From this equation, some interesting matters would easily be seen. For example, when the distribution of defects was known in a series of specimens having a fixed dimension of defects, the deviation and the average value of σ_m would be evaluated. That is, when the distribution of defects was assumed to be in simple cubic array having spacing of λ , the maximum value of σ_m would be obtained at $\Delta t = \lambda$ and $\Delta l = \lambda/2$, and the minimum would be at $\Delta t = \Delta l = 0$. The values of σ_m calculated from the above equation using various values of $2a$ and λ are shown in Fig. 9. For the simplest example, when a series of specimens containing only the defect with $2a = 20 \mu$ and simple cubic distribution of $\lambda = 0.2 \text{ mm}$, the average value of σ_m would be about 280 kg/mm^2 , which was commonly obtained in the case of $2a = 40 \mu$ and $\lambda = 0.8 \text{ mm}$, although the range of deviation was different

Fig. 9 σ_m calculated as functions of $2a$ and λ according to eq. (4). λ ; spacing of defects, distribution of which is assumed to be of simple cubic array.

in each case, that is, about $260 \sim 300 \text{ kg/mm}^2$ in the former case and $200 \sim 350 \text{ kg/mm}^2$ in the latter case. In order to calculate the actual strength of the specimen, it is natural that the true distributions of dimensions of defects should be taken into consideration. Anyhow, it is clear that the strength sharply increases with decreasing dimension and increasing spacing of defects.

To be more concrete, the transverse-rupture strength should rise to a value higher than at least about 400 kg/mm^2 (according to JIS), for instance, when the dimension of defects was decreased to the value almost equal to 1.7μ (mean grain size of carbide). On the other hand, it is possible to estimate the strength of the transverse-ruptured specimen by substituting the measured values of defects on the fracture surface for (a , Δt , and Δl) in eq. (4).

Considering that the strength of the matrix (σ_0) and the radius of curvature (ρ) in cemented carbide should be functions of the mean grain size and contiguity of WC, the amount of binder, carbon content (the binder phase composition), etc., the various straight lines with various values of σ_0 and ρ would be obtained according to these factors. Therefore, the analysis of those straight lines would help to comprehend the various factors affecting the strength in addition to the theory of the strength of cemented carbides. This subject is now under investigation by the present authors. Weibull's equation⁽¹⁷⁾ is in this sense considered to be somewhat abstract, because the dimensions and

the distribution of visible defects are not introduced into his theory.

Furthermore, the difference between the transverse-rupture strength and the tensile strength of cemented carbides would be explained. In the case of $2a = 40\mu$ and $\lambda = 0.8 \text{ mm}$, the average value of σ_m was about 280 kg/mm^2 as mentioned above; however, the tensile strength was assumed to be about 200 kg/mm^2 corresponding to the minimum value of σ_m . Experimental results⁽¹⁸⁾ of the present authors support this presumption.

From these results, it is clear that the strength of cemented carbides cannot be fully understood through the studies on the conventional various factors mentioned, because the strength has been confirmed to be directly controlled by a small amount of defects which are generally not observed on a usually polished surface prepared for the microstructure. It is also concluded that the presumption of the distribution or the mean value of the strength would be possible when the dimensions and the distribution of defects are given. In other words, the present authors obtained a method to strengthen the cemented carbide including cermets and ceramics, and an indication to develop the theory of the strength.

V. Conclusions

The transverse-rupture strength (σ_m) of WC-10%Co cemented carbide with a mean carbide grain size of 1.7μ was studied in relation to the microstructural defects which generally appeared in the so-called white spot on the fractured surface. The results obtained were as follows.

(1) It was confirmed that the microstructural defect such as micropore, coarse WC grain or binder-pool existed nearly at the center of the white spot (a kind of fracture mirror in ceramics), and that the defect generally acted as a source of fracture. The white spot was considered to correspond with the radial fracture zone with crack propagation of low velocity.

(2) The value of σ_m did not closely correlate with the dimensions of the white spot and defects or with their locations, but the calculated

external stress (σ_d) acted on the white spot or the defect at the moment of the fracture closely correlated with their dimensions. The relation between σ_d and the dimension of defects was considered to be of special interest.

(3) The effects of the dimension and location of defects on σ_m were expressed by eq. (4), which is common in three kinds of defects, showing that the strength of cemented carbide is directly controlled by such microstructural defects. Thus, when the spacing and dimension of defects are known, the deviation and the mean value of σ_m for a series of specimens would be evaluated. With decreasing defect size and increasing defect spacing, the value of σ_m of cemented carbide would show a marked increase up to a value higher than about 400 kg/mm² (according to JIS) in the specimen used here. The value of σ_m of a fractured specimen would be calculated using the measured values for the dimension and the location of the defects on the fracture surface.

(4) Various results so far reported for the strength of cemented carbides from the viewpoint that the strength is affected by conventional various factors are considered to be still ambiguous; therefore, further examination of these factors should be required.

REFERENCES

- (1) H. E. Exner and J. Gurland: *Powder Met.*, **13** (1970), 13.
- (2) H. Suzuki: *Bull. Japan Inst. Metals*, **11** (1972), 125.
- (3) H. Suzuki, T. Yamamoto and H. Sakanoue: *J. Japan Inst. Met.*, **32** (1968), 993.
- (4) H. Suzuki, K. Hayashi and Y. Fuke: *J. Japan Soc. Powder and Powder Met.*, **19** (1972), 346.
- (5) P. O. Snell and E. Pärnama: *Planseeberichte für P/M*, **21** (1973), 271.
- (6) H. Suzuki and K. Hayashi: To be published.
- (7) G. S. Kreimer and N. A. Alekseyeva: *Phys. Met. and Metallog.*, **13** (1962), 117.
- (8) J. Gurland: *Trans. Met. Soc., AIME*, **227** (1963), 1146.
- (9) F. R. N. Nabarro and S. Bartolucci: *Conference Preprints of International Conference on Strength of Metals and Alloys held in Tokyo, (1967)*, p. 374.
- (10) H. Suzuki and K. Hayashi: *Trans. JIM*, **10** (1969), 360.
- (11) R. C. Bradt, D. P. H. Hasselman and F. F. Lange: *Fracture Mechanics of Ceramics*, Plenum Press, New York. (1973), 309.
- (12) ASTM Designation: B276-1954.
- (13) A. S. Tetelman and A. J. McEvily, Jr.: *Fracture of Structural Materials*, John Wiley and Sons, Inc. (1967), 48.
- (14) R. H. Edwards: *J. Appl. Mech.*, **18** (1951), 19.
- (15) H. Doi, Y. Fujiwara, K. Miyake et al.: *Met. Trans.*, **1** (1970), 19.
- (16) J. J. Burke et al.: *Strengthening Mechanisms*, Syracuse Univ. Press. (1966), p. 3.
- (17) W. Weibull: *Ingenios Veteskaps Akad. Handl.*, **153** (1939), 151.
- (18) H. Suzuki and K. Hayashi: To be published.